TRANSACTIONS

of the

American

Foundrymen's Association

Proceedings of the

Fiftieth Annual Meeting

Cleveland, Ohio

May 6-10, 1946

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OFFICERS

of the

AMERICAN FOUNDRYMEN'S ASSOCIATION

(INCORPORATED)

(Fiscal Year Ending June 30, 1947)

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**Vice-President,* MAX KUNIANSKY, Lynchburg Foundry Co., Lynchburg, Va.

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Staff Officers

Headquarters, Room 1198, 222 W. Adams St., Chicago, Ill.

Secretary-Treasurer, WM. W. MALONEY

Technical Director, S. C. MASSARI

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* Members, Executive Committee.

Summary of Proceedings of the 50th Annual Meeting

GOLDEN ANNIVERSARY—the 50th Annual Convention of the American Foundrymen's Association was held in Cleveland, Ohio, May 6 to 10, 1946. The biennial Foundry and Allied Industries Show was staged by the Association in conjunction with the Convention. Exhibits were displayed in the Cleveland Public Auditorium.

Credit for the excellent reception and hospitality received by those attending is given to the Northeastern Ohio Chapter of A.F.A. under the chairmanship of A. C. Denison, Fulton Foundry & Machine Co., Inc. This committee excelled in its splendid work and hospitality as host to the large number of members and guests attending.

This exhibit broke all previous records from the standpoint of number and quality of exhibits. Six acres of exhibit space, displaying an unparalleled array of new equipment and materials for the industry, was a source of constant activity throughout the convention. The attendance of 18,000 registered members was also a record breaker, which taxed the capacity of hotels in Cleveland. The number and value of papers and discussions presented was outstanding. Many papers were of extremely pertinent coverage on current foundry problems and practices. During the five days of the meeting, over 50 sessions were held for the presentation of over 95 papers and committee reports. A summary of the sessions which were held in the meeting rooms in the Cleveland Public Auditorium, Statler and Cleveland Hotels follows:

ALUMINUM AND MAGNESIUM ROUND TABLE LUNCHEON

Monday, May 6, 12:00 P.M.

Presiding—R. E. Ward, Eclipse Pioneer Div., Bendix Aviation Corp., Bendix, N. J.

Co-Chairman—A. T. Ruppe, Bendix Products Corp., South Bend, Ind.

Report of Committee on Recommended Practices for Aluminum and Magnesium Die Castings, by J. C. Fox, Doehler-Jarvis Corp., New York.

Report of Committee on Recommended Practices for Aluminum and Magnesium Permanent Mold Castings, by A. Sugar, American Metal Co. Ltd., New York.

JOINT MEETING, PLANT EQUIPMENT AND SAFETY & HYGIENE

Monday, May 6, 2:00 P.M.

Presiding—James Thomson, Continental Foundry & Machine Co., East Chicago, Ind.

Co-Chairman—C. P. Guion, W. W. Sly Mfg. Co., Chicago.

Accident Prevention as a Function of Management, by R. R. Meigs, Liberty Mutual Insurance Co., Boston.

Mechanical Shakeouts, by J. L. Yates, Worthington Pump & Machinery Corp., Buffalo, N. Y.

LECTURE COURSE (SESSION 1)

Monday, May 6, 4:00 P.M.

Presiding—D. Basch, Almin Co., Schenectady, N. Y.

Co-Chairman—H. Brown, Aluminum Industries Inc., Cincinnati.

Subject—*Aluminum and Magnesium Foundry Control.*

Discussion Leader—Oscar Bloh m, Hills-McCanna Co., Chicago.

GRAY IRON SHOP COURSE (SESSION 1)

Monday, May 6, 4:00 P.M.

Presiding—George A. Timmons, Climax Molybdenum Co., Detroit.

Co-Chairman—P. T. Bancroft, Republic Coal & Coke Co., Moline, Ill.

Subject—*The Metallurgy of Cupola Mixes.*

Discussion Leader—K. H. Priestley, Eaton Mfg. Co., Vassar, Mich.

SAND SHOP COURSE (SESSION 1)

Monday, May 6, 8:00 P.M.

Presiding—F. S. Brewster, Dow Chemical Co., Bay City, Mich.

Subject—*Magnesium Molding Sands.*

Discussion Leader—Oscar Bloh m, Hills-McCanna Co., Chicago.

JOINT MEETING, PLANT EQUIPMENT AND SAFETY & HYGIENE

Monday, May 6, 8:00 P.M.

Presiding—James Thomson, Continental Foundry & Machine Co., East Chicago, Ind.

Co-Chairman—W. O. Vedder, Pangborn Corp., Hagerstown, Md.

Exhaust Hoods and Piping Systems, by E. A. Carsey, Kirk & Blum Mfg. Co., Cincinnati.

Dust Collection and Disposal

Centrifugal Collectors, by H. H. Dohrmann, Buell Engineering Co., New York.

Cloth Collectors, by G. A. Boesger, W. W. Sly Mfg. Co., Cleveland.

Wet Collectors, by A. S. Lundy, Claude B. Schneible Co., Detroit.

Fans and Exhausters, by Phillip Cohen, B. F. Sturtevant Co., Cleveland.

Maintenance, by K. M. Smith, Caterpillar Tractor Co., Peoria, Ill.

Note: The papers presented at this Session are also published separately as the "Symposium on Foundry Dust Control."

ALUMINUM AND MAGNESIUM

Monday, May 6, 8:00 P.M.

Presiding—Leslie Brown, Magnesium Fabricators Div., Bohn Aluminum & Brass Corp., Adrian, Mich.

Co-Chairman—G. G. Glider, Ebaloy Foundries Inc., Rockford, Ill.

Interrelation of Various Major Factors Determining the Amount of Microporosity in Magnesium Alloy Castings, by L. W. Eastwood and J. A. Davis, Battelle Memorial Institute, Columbus, Ohio.

The Effect of Mold Materials and Practices on the Quality of Magnesium Alloy Castings, by M. V. Chamberlin and J. G. Mezoff, Dow Chemical Co., Midland, Mich.

Centrifugal Castings of a Magnesium Part, by F. P. Strieter and R. J. Maenner, Dow Chemical Co., Midland, Mich.

MALLEABLE FOUNDRY PRACTICE

Tuesday, May 7, 10:00 A.M.

Presiding—A. M. Fulton, Northern Malleable Iron Co., St. Paul, Minn.

Co-Chairman—C. F. Joseph, Saginaw Malleable Iron Div., General Motors Corp., Saginaw, Mich.

Chemical Composition of Malleable Iron, by H. A. Schwartz, National Malleable & Steel Castings Co., Cleveland.

Cores for Automotive Malleable Iron Castings, by W. G. Ferrell, Auto Specialties Mfg. Co., St. Joseph, Mich.

Heat Treating Furnaces in Foundry Work, by Victor Paschkis, Columbia University, New York City.

ALUMINUM AND MAGNESIUM SESSION

Tuesday, May 7, 10:00 A.M.

Presiding—Walter Bonsack, National Smelting Co., Cleveland.

Co-Chairman—L. W. Eastwood, Battelle Memorial Institute, Columbus.

Heat Treatment of Aluminum Alloy Castings, by Walter E. Sicha and H. J. Rowe, Aluminum Co. of America, Pittsburgh, Pa.

The Effect of Certain Elements on Some Properties of an Aluminum Sand Casting Alloy, by R. A. Quadt, American Smelting & Refining Co., Barber, N. J.

Characteristics of Aluminum Casting Alloy 3 Per Cent Silicon—5 Per Cent Copper, by O. Tichy, National Smelting Co., Cleveland.

INSPECTION OF CASTINGS

Tuesday, May 7, 10:00 A.M.

Presiding—Harold W. Warner, Allis-Chalmers Mfg. Co., Milwaukee.

Co-Chairman—E. G. Leverenz, American Steel Foundries, East Chicago, Ind.

Duties and Functions of the Inspection Dept., As Seen by Management, by M. D. Johnson, Purolator Products Inc., Newark, N. J.

Inspection—The Backbone of Quality Control, by W. H. Gunselman, Samuel Greenfield Co., Inc., Buffalo, N. Y.

Non-Destructive Inspection of Castings, by C. L. Frear and R. E. Lyons, Bureau of Ships, U. S. Navy Dept., Washington, D. C.

LUNCHEON—ENGINEERING SCHOOL ALUMNI GROUPS

Tuesday, May 7, 12:00 P.M.

Presiding—F. G. Sefing, International Nickel Co., New York.

CANADIAN MEMBERS' LUNCHEON

Tuesday, May 7, 12:00 P.M.

Presiding—Joseph Suily, Sully Foundry Div., Neptune Meters, Ltd., Toronto, Ont.

NORTHEASTERN OHIO DAY

Tuesday, May 7, 1:00-10:00 P.M.

Chairman—Edw. Follman, Griffin Wheel Co., Cleveland.

INSPECTION OF CASTINGS

Tuesday, May 7, 2:00 P.M.

Presiding—E. F. Platt, Sperry Gyroscope Co., Brooklyn, N. Y.

Co-Chairman—W. H. Gunselman, Samuel Greenfield Co. Inc., Buffalo, N. Y.

Panel Discussion—*Suppliers' and Buyers' Viewpoint of Inspection in the Foundry.*

Discussion Leaders:

O. B. Sandberg, Chicago & Northwestern R. R., Chicago.
Walter Klayer, Aluminum Industries Inc., Cincinnati.
John Amoroso, Sterling Engine Co., Buffalo, N. Y.
G. Dewald, Ampco Metals, Inc., Milwaukee.
H. Wallace, American Steel Foundries, Verona, Pa.
Nelson Stuit, Grede Foundries, Inc., Milwaukee.
F. L. Bender, Chicago Hardware Foundry Co., North Chicago, Ill.

REFRACTORIES

Tuesday, May 7, 2:00 P.M.

Presiding—C. E. Bales, Ironton Fire Brick Co., Ironton, Ohio.

Co-Chairman—C. S. Reed, Chicago Retort & Fire Brick Co., Chicago.

Suggested Cupola Refractory Standardization, by R. P. Stevens, Chicago Retort & Fire Brick Co., Chicago.

Foundry Refractories—Their Properties and Application, by C. A. Brashares, Harbison-Walker Refractories Co., Pittsburgh, Pa.

Refractories in Non-Ferrous Practice, by H. R. Kohl, Algonac Foundry Co., Algonac, Mich.

The Use of Rammed Refractories in Electric Steel Foundries, by R. H. Zoller, Zoller Castings Co., Bettsville, Ohio.

MALLEABLE FOUNDRY PRACTICE

Tuesday, May 7, 2:00 P.M.

Presiding—R. J. Anderson, Belle City Malleable Iron Co., Racine, Wis.

Co-Chairman—D. I. Dobson, General Malleable Corp., Waukesha, Wis.

Malleable Foundry Core Making Practice for Pipe Fittings and Small Castings, by D. F. Sawtelle, Malleable Iron Fittings Co., Branford, Conn.

Core Sand Practice and Control in a Mechanized Malleable Iron Foundry, by Joseph J. Clark, Saginaw Malleable Iron Div., General Motors Corp., Saginaw, Mich.

ALUMINUM AND MAGNESIUM

Tuesday, May 7, 4:00 P.M.

Presiding—W. E. Martin, National Smelting Co., Cleveland.

Co-Chairman—A. W. Stolzenburg, Aluminum Co. of America, Detroit.

Committee Report—*Unsoundness in Cast Light Alloys,* by L. W. Eastwood, Battelle Memorial Institute, Columbus, Ohio. (A.F.A. Committee on Shrinkage and Porosity.)

Report of Committee on Dimensional Stability of Aluminum Castings, by Hiram Brown, Solar Precision Castings Co., Des Moines.

The Relation of Design and Metallurgical Factors to Serviceability of Magnesium Castings, by G. H. Found, Dow Chemical Co., Bay City, Mich.

GRAY IRON SHOP COURSE (SESSION 2)

Tuesday May 7, 4:00 P.M.

Presiding—T. E. Barlow, Battelle Memorial Institute, Columbus, Ohio.

Co-Chairman—W. B. McFerrin, Electro Metallurgical Co., Detroit.

Casting Defects Attributable to Metal, A.F.A. Committee on Analysis of Castings Defects.

LECTURE COURSE (SESSION 2)

Tuesday May 7, 4:00 P.M.

Presiding—W. B. McFerrin, Electro Metallurgical Co., Detroit.

Co-Chairman—W. D. McMillan, International Harvester Co., McCormick Works, Chicago.

Malleable Foundry Control, by C. F. Joseph, Saginaw Malleable Iron Div., General Motors Corp., Saginaw, Mich.

CHAPTER OFFICERS AND DIRECTORS DINNER

Tuesday, May 7, 7:00 P.M.

Presiding—A.F.A. Vice-President S. V. Wood, Minneapolis Electric Steel Castings Co., Minneapolis, Minn.

REFRACTORIES

Tuesday, May 7, 8:00 P.M.

Presiding—C. E. Bales, Ironton Fire Brick Co., Ironton, Ohio.

Co-Chairman—C. S. Reed, Chicago Retort & Fire Brick Co., Chicago.

The Lining and Patching of Cupolas, by E. J. Lally, Jr., Forest City Foundries Co., Cleveland.

Information Please on Foundry Refractories—Question and Answer Panel.

Moderator—C. S. Reed, Chicago Retort & Fire Brick Co., Chicago.

Panel Leaders:

Richard H. Stone, Vesuvius Crucible Co., Swissvale, Pa.
J. A. Bowers, American Cast Iron Pipe Co., Birmingham, Ala.
E. J. Carmody, Chas. C. Kawin Co., Chicago.
A. S. Klopf, Lester B. Knight & Associates, Inc., Chicago.

SAND SHOP COURSE (SESSION 2)

Tuesday, May 7, 8:00 P.M.

Presiding—D. F. Sawtelle, Malleable Iron Fittings Co., Branford, Conn.

Subject—*Malleable Molding Sand Control.*
Discussion Leaders:
E. C. Zirzow, National Malleable & Steel Castings Co., Cleveland.
L. Richard Spann, Eastern Malleable Iron Co., Naugatuck, Conn.

MALLEABLE FOUNDRY PRACTICE

Wednesday, May 8, 10:00 A.M.

Presiding—W. D. McMillan, International Harvester Co., McCormick Works, Chicago.
Co-Chairman—D. F. Sawtelle, Malleable Iron Fittings Co., Branford, Conn.
Core Practice in an Agricultural Malleable Foundry, by Eric Welander, Union Malleable Iron Works, East Moline, Ill.
Malleable Core Making Practice, by E. C. Zirzow, National Malleable & Steel Castings Co., Cleveland.

BRASS AND BRONZE

Wednesday, May 8, 10:00 A.M.

Presiding—D. Frank O'Connor, American Saw Mill Machine Co., Hackettstown, N. J.
Co-Chairman—Herman Smith, Federated Metals Div., American Smelting & Refining Co., Pittsburgh, Pa.
Observations on Gas Elimination in Non-Ferrous Alloy Castings, by George Dalbey, U. S. Navy Yard, Mare Island, Calif.
Correct Combustion for Crucible Furnace Melting, by A. C. Schmid, Jos. Dixon Crucible Co., Jersey City, N. J.
Conditions of Flow in Bronze Castings, by Lt. J. T. Robertson and R. G. Hardy, Naval Research Laboratory, Washington, D. C.

SAND RESEARCH

Wednesday, May 8, 10:00 A.M.

Presiding—H. F. Taylor, Massachusetts Institute of Technology, Cambridge, Mass.
Co-Chairman—J. A. Rassenfoss, American Steel Foundries, Indiana Harbor, Ind.
Sixth Annual Report of Committee on Physical Properties of Foundry Sands at Elevated Temperatures, by D. C. Williams, A.F.A. Research Fellow, Cornell University, Ithaca, N. Y.
Heat Absorption of Sand, by H. W. Dietert and Robert L. Doelman, Harry W. Dietert Co., Detroit.
Molding Sand Binders, by L. Jenicek, Prague, Czechoslovakia.
Measurement and Investigation of Gases Escaping from Oil Cores, by Vojtech Volenik, Czechoslovakia, Czechoslovak Foundrymen's Association Exchange Paper.

SYMPOSIUM ON ENGINEERING PROPERTIES OF GRAY IRON

Wednesday, May 8, 10:00 A.M.

Presiding—R. G. McElwee, Vanadium Corp. of America, Detroit.
Co-Chairman—W. E. Mahin, Armour Research Foundation, Illinois Institute of Technology, Chicago.
The Engineering Properties of Gray Iron, by W. E. Mahin, Armour Research Foundation, Chicago.
Note: This paper will be published as a separate publication.

BRASS AND BRONZE ROUND TABLE LUNCHEON

Wednesday, May 8, 12:00 P.M.

Presiding—B. A. Miller, Baldwin Locomotive Works, Cramp Brass & Iron Foundries Div., Philadelphia.
Co-Chairman—C. Louis Lane, Florence Pipe Foundry & Machine Co., Florence, N. J.
Molding Sands for Brass and Bronze, by L. B. Osborne, Hougland & Hardy, Inc., Evansville, Ind.
The Effect of Section Size and Chills on Properties of Sand-Cast Bronzes, by W. B. George and A. H. Hesse, R. Lavin & Sons, Inc., Chicago.
New Type Exothermic Powders for Making Sound Castings, by C. G. Lutts, J. P. Hickey, and Michael Bock II, Boston Naval Shipyard, Boston.

MALLEABLE ROUND TABLE LUNCHEON

Wednesday, May 8, 12:00 P.M.

Presiding—J. H. Lansing, Malleable Founders' Society, Cleveland.
Co-Chairman—A. M. Fulton, Northern Malleable Iron Co., St. Paul, Minn.; and K. H. Hamblin, Columbia Malleable Castings Corp., Columbia, Pa.
Atmosphere Indicators in Melting.
Single and Double Burner Use in Melting.
Use of Large End Test Bars.
Sulphur-Manganese Ratios and the Effect on Other Elements.

SAND RESEARCH

Wednesday, May 8, 2:00 P.M.

Presiding—B. H. Booth, Carpenter Bros., Inc., Milwaukee.
Co-Chairman—R. E. Morey, Naval Research Laboratory, Washington, D. C.
The A.F.A. Clay and Fineness Determination Can Be Quantitative, by D. C. Williams, A.F.A. Research Fellow, Cornell University, Ithaca, N. Y.
The Fineness Tests of Molding Sands, by A. I. Krynitsky and Margaret Price, Bureau of Standards, Washington, D. C.
The Prevention of Veining and Penetration on Castings Made in Synthetically Bonded Sands, by R. E. Morey and J. R. Kattus, Naval Research Laboratory, Washington, D. C.
Behavior of Molding Materials in Their Own Atmosphere at Elevated Temperatures, by H. W. Dietert, H. W. Dietert Co., Detroit.

STEEL

Wednesday, May 8, 2:00 P.M.

Presiding—H. H. Blosjo, Minneapolis Electric Steel Castings Co., Minneapolis.
Co-Chairman—Charles Locke, West Michigan Steel Foundry Co., Muskegon, Mich.
Bore Cracks in Cast Steel Valves and Fittings, by H. F. Bishop, W. H. Johnson, Naval Research Laboratory, Washington, D. C., and H. F. Taylor, Massachusetts Institute of Technology, Cambridge, Mass.
Homogenization Heat Treatment of Cast Steel, by P. C. Rosenthal and J. G. Kura, Battelle Memorial Institute, Columbus, Ohio.
Thermal Relief of Stresses in Steel Castings and Weldments, by C. R. Jelm and S. A. Herres, Watertown Arsenal, Watertown, Mass.
Committee Report on Heat Treatment of Steel Castings, by E. R. Young, Climax Molybdenum Co., Chicago.

SYMPOSIUM ON ENGINEERING PROPERTIES OF GRAY IRON

Wednesday, May 8, 2:00 P.M.

Presiding—R. G. McElwee, Vanadium Corp. of America, Detroit.
Co-Chairman—W. E. Mahin, Armour Research Foundation, Illinois Institute of Technology, Chicago.
The Ductility and Elasticity of White and Gray Irons, by R. A. Flinn and H. J. Chapin, American Brake Shoe Co., Mahwah, N. J.
Gray Cast Iron Tensile Strength, Brinell Hardness and Composition Relationships, by T. E. Barlow and C. H. Lorig, Battelle Memorial Institute, Columbus, Ohio.
Effect of Carbon Equivalent and Section Size on Physical Properties of Gray Iron, by K. Geist, Allis-Chalmers Mfg. Co., Milwaukee.
Structural Control for Wear Resistance, by F. G. Sefing, International Nickel Co., New York.
Practical Considerations Affecting the Section Sensitivity of Gray Iron, by H. C. Winte, Worthington Pump & Machinery Corp., Buffalo.

GRAY IRON SHOP COURSE (SESSION 3)

Wednesday, May 8, 2:00 P.M.

Presiding—W. A. Hambley, Allis-Chalmers Mfg. Co., Milwaukee.
Co-Chairman—George W. Anselman, Goebig Mineral Supply Co., Chicago.

Casting Defects Attributable to Metal, A.F.A. Committee on Analysis of Castings Defects.

LECTURE COURSE (SESSION 3)

Wednesday, May 8, 4:00 P.M.

Presiding—H. M. St. John, Crane Co., Chicago.
Co-Chairman—C. O. Thieme, H. Kramer & Co., Chicago.
Subject—*Brass and Bronze Foundry Control.*
Discussion Leader—D. Frank O'Connor, American Saw Mill Machinery Co., Hackettstown, N. J.

NORTHEASTERN OHIO CHAPTER, RECEPTION FOR FOREIGN VISITORS, MIDDAY CLUB, UNION COMMERCE BLDG.

Wednesday, May 8, 5:00 P.M.

Presiding—J. H. Tressler, Hickman, Williams & Co., Cleveland.
Co-Chairman—W. G. Gude, Penton Publishing Co., Cleveland.

FOUNDRY INSTRUCTORS' DINNER

Wednesday, May 8, 6:00 P.M.

FOREMAN TRAINING

Wednesday, May 8, 8:00 P.M.

Presiding—F. E. Wartgow, A.F.A., Chicago.
Engineer of Human Relations, by Steven G. Garry, Caterpillar Tractor Co., Peoria, Ill.

ENGINEERING STUDENT TRAINING

Wednesday, May 8, 8:00 P.M.

Presiding—G. J. Barker, University of Wisconsin, Madison, Wis.
Co-Chairman—C. J. Freund, University of Detroit.
Panel Discussion—*College Graduates vs. Employers.*

SAND SHOP COURSE (SESSION 3)

Wednesday, May 8, 8:00 P.M.

Presiding—E. E. Woodliff, Foundry Sand Service Engineering Co., Detroit.
Subject—*Brass and Bronze Sand Control.*
Discussion Leader—W. M. Ball, Jr., Magnus Brass Div., National Lead Co., Cincinnati.

GRAY IRON SHOP COURSE (SESSION 4)

Thursday, May 9, 8:30 A.M.

Presiding—E. J. Burke, Hanna Furnace Corp., Detroit.
Co-Chairman—P. T. Bancroft, Republic Coal & Coke Co., Moline, Ill.
Subject—*"The Metallurgy of Carbon Control in the Cupola,* by Ralph A. Clark, Electro Metallurgical Co., Chicago.

ANNUAL BUSINESS MEETING

Thursday, May 9, 9:30 A.M.

Presiding—F. J. Walls, President, American Foundrymen's Association.

President Walls called the meeting to order as the Annual Business Meeting of the American Foundrymen's Association. He then introduced T. Makemson, Secretary, Institute of British Foundrymen and Official Representative of the British Ministry of Supply in charge of cast iron production in Great Britain, also the first and only Secretary of the International Committee of Foundry Technical Associations. Mr. Makemson expressed greetings to A.F.A. on behalf of foundrymen of Europe visiting the Convention. Representatives of 19 countries outside the United States and Canada have registered at the Convention. These countries were as follows: Argentina, Australia, Belgium, Brazil, Czechoslovakia, China, Denmark, England, France, Holland, Ireland, Mexico, Norway, Russia, Scotland, South Africa, Sweden, Switzerland and the Panama Canal Zone.

President Walls then read telegrams conveying best wishes to the American Foundrymen's Association from R. Cheers, President, Institute of Australian Foundrymen, and Prof. F. Pisek, President, Czechoslovak Foundrymen's Association.

Following these preliminaries, President Walls delivered the President's Annual Address, which was followed by the report of the Secretary and the report of the Treasurer. After these reports were presented, Secretary Maloney in presenting the report on the Annual Apprentice Contest Winners said that the judges had been selected through the assistance of J. G. Goldie, Head, Foundry Division, and Frank C. Cech, Head, Patternmaking Division, Cleveland Trade School, Cleveland, both members of the Apprentice Contest Committee. The judges secured were as follows:

Gray Iron Molding
Grover C. Cole, Forest City Foundries Co., Cleveland
Frank Weisehan, Ferro Machine & Foundry Co., Cleveland
Milo Barrett, Chandler & Price Co., Cleveland

Steel Molding
Tom West, West Steel Castings Co., Cleveland
Louis Klein, Crucible Steel Castings Co., Cleveland
E. C. Zirzow, National Malleable & Steel Castings Co., Cleveland

Non-Ferrous Molding
Dan Gluntz, Gluntz Brass Foundry, Cleveland
Edward J. Metzger, Wellman Bronze & Aluminum Co., Cleveland
Fred J. Fredriksen, Westinghouse Electric Corp., Cleveland

Patternmaking
Walter Seebeck, Western Pattern Works, Cleveland
Tom Gallagher, Cleveland Punch & Shear Works, Cleveland
V. J. Sedlon, Master Pattern Co., Cleveland
George Gedeon, Aluminum Company of America, Cleveland
Edward Pierie, Motor Patterns Co., Cleveland
Pete Rettig, Rettig Pattern Co., Cleveland
Carl Hamburger, Cleveland Board of Education, Cleveland
Ray Tisdale, Service Pattern Works, Cleveland
Frank Hrabak, East End Pattern Works, Cleveland
James Young, Standard Pattern Works, Cleveland
George Gibala, Royal Pattern Works, Cleveland

Mr. Maloney announced the winners in the various classes as follows:

Gray Iron Molding Division
1st—Lawrence Kinsinger, Caterpillar Tractor Co., Peoria, Ill.
2nd—M. Tamburini, Warden King Ltd., Montreal, Que.
3rd—Harold Young, Cleveland Trade School, Cleveland

Steel Molding Division
1st—Robert Bina, Crucible Steel Casting Co., Cleveland
2nd—John Pietrzyki, Crucible Steel Casting Co., Cleveland
3rd—John Kraynak, West Steel Casting Co., Cleveland

Non-Ferrous Molding Division
1st—C. Corriveau, Montreal Technical School, Montreal, Que.
2nd—P. Blais, Montreal Technical School, Montreal, Que.
3rd—H. Deschamps, Jenkins Bros., Montreal, Que.

Patternmaking Division
1st—R. Dragon, Western Pattern Works, Montreal, Que.
2nd—John Yeich, Birdsboro Steel Foundry & Machine Co., Birdsboro, Pa.
3rd—Michael Simenic, Aluminum Co. of America, Cleveland

Following announcement of winners, Pres. Walls announced that the Association had arranged each first prize winner to attend the convention and to receive his award in person. He then called the first prize winners to the platform, expressed congratulations and presented them with engraved certificates and checks for $30.00. The second prize was $20.00 and the third prize was $10.00 in addition to the engraved certificate.

After presentation of awards the Secretary of the Association reported on nominations of Officers and Directors for the coming year and stated that no additional nominees had been received in accordance with the procedure prescribed in the Association By-laws. He therefore was instructed by Pres. Walls to cast the unanimous ballot of the membership of A.F.A. for election of the following:

President (to serve one year):
S. V. Wood, President and General Manager, Minneapolis Electric Steel Castings Co., Minneapolis.

Vice-President (to serve one year):
Max Kuniansky, Vice-President and General Manager, Lynchburg Foundry Co., Lynchburg, Va.

Directors (to serve three years):

J. E. Kolb, Pattern Shop Superintendent, Caterpillar Tractor Co., Peoria, Ill.

H. G. Lamker, Superintendent of Foundries, Wright Aeronautical Co., Paterson, N. J.

B. L. Simpson, President, National Engineering Co., Chicago.

Stowell Wasson, Manager of Chicago and Melrose Park Plants, National Malleable & Steel Castings Co., Chicago.

Director (to serve one year):

F. J. Walls, Metallurgist and Manager, Detroit Office, International Nickel Co., Detroit.

Following announcement of elections, Pres. Walls introduced the newly elected Officers and Directors present.

Following this brief introduction, Pres. Walls introduced Dr. G. H. Clamer, President, Ajax Metal Company, Philadelphia, Pa., who presented the Fourth Annual Foundation Lecture on the subject of "Test Bars for 85-5-5-5 Alloy, Their Design and Some Factors Affecting Their Design." (Dr. Clamer's lecture will be found beginning on page 1 of this volume.)

The meeting was adjourned upon completion of this lecture.

STEEL ROUND TABLE LUNCHEON

Thursday, May 9, 12:00 P.M.

Presiding—T. D. West, West Steel Casting Co., Cleveland.
Co-Chairman—E. C. Troy, Dodge Steel Co., Philadelphia.
Subject—*General Discussion on Steel Foundry Practice.*

BRASS AND BRONZE

Thursday, May 9, 2:00 P.M.

Presiding—G. K. Dreher, Ampco Metal, Inc., Milwaukee.
Co-Chairman—A. K. Higgins, Allis-Chalmers Mfg. Co., Milwaukee.

The Use of Insulating Pads and Riser Sleeves for Producing Sound Bronze Castings, by H. F. Taylor and W. C. Wick, formerly Naval Research Laboratory, Washington, D. C.

Knock-off Risers for Non-Ferrous Castings, by Stanley W. Brinson and J. A. Duma, Norfolk Navy Yard, Portsmouth, Va.

Sealing Bronze Pressure Castings Through Heat Treatment, by F. L. Riddell, Naval Research Laboratory, Washington, D. C.

PATTERNMAKING

Thursday, May 9, 2:00 P.M.

Presiding—Frank C. Cech, Cleveland Trade School, Cleveland.
Co-Chairman—V. J. Sedlon, Master Pattern Co., Cleveland.

The Use of Wood Models in Planning and Design, by G. W. Schuller, Caterpillar Tractor Co., Peoria.

Patterns as the Production Foundry Sees Them, by Guy Pealer, Elmira Foundry Co., Inc., Elmira, N. Y.

GRAY IRON

Thursday, May 9, 2:00 P.M.

Presiding—T. E. Eagan, Cooper-Bessemer Corp., Grove City, Pa.

Microstructure Related to Properties of Cast Iron, by W. E. Mahin and H. W. Lownie, Jr., formerly Westinghouse Electric Corp., East Pittsburgh, Pa.

Cupola Operation and Control, by W. W. Levi, Lynchburg Foundry Co., Radford, Va.

Technical Organization of the French Foundry, by A. Le Thomas, President, French Technical Committee of Foundry Industries—French exchange paper.

JOB EVALUATION AND TIME STUDY

Thursday, May 9, 2:00 P.M.

Presiding—Robert J. Fisher, Falk Corp., Milwaukee.
Co-Chairman—F. E. Wartgow, American Foundrymen's Association, Chicago.

What Management Should Know About Timestudy, by Phil Carroll, Jr., Maplewood, N. J.

FOUNDRY COSTS

Thursday, May 9, 4:00 P.M.

Presiding—R. L. Lee, Grede Foundries, Inc., Milwaukee.
Co-Chairman—L. M. Nesselbush, Falcon Bronze Co., Youngstown, Ohio.
Subject—*Current Foundry Problems and Fundamentals of Cost for Small Foundries.*

Meeting Foundry Competition With a Sound Cost Program, by A. E. Grover, Cleveland Heights, Ohio.

APPRENTICE TRAINING

Thursday, May 9, 4:00 P.M.

Presiding—F. C. Cech, Cleveland Trade School, Cleveland.
Co-Chairman—W. J. Hebard, Continental Foundry & Machine Co., East Chicago, Ind.
Panel Discussion:

Standard Apprentice Training, by Carl F. Haertel, Falk Corp., Milwaukee.

Accelerated Training Program, by W. J. Hebard, Continental Foundry & Machine Co., East Chicago, Ind.

Upgrading Systems, by M. J. Gregory, Peoria, Ill.

Veteran Training, by W. M. Owen, Caterpillar Tractor Co., Peoria, Ill.

General Summary, by C. J. Freund, University of Detroit, Detroit.

LECTURE COURSE (SESSION 4)

Thursday, May 9, 4:00 P.M.

Presiding—John Howe Hall, General Steel Castings Corp., Eddystone, Pa.
Co-Chairman—C. E. Sims, Battelle Memorial Institute, Columbus, Ohio.

Steel Foundry Control, by G. A. Lillieqvist, American Steel Foundries, East Chicago, Ind.

APPRENTICE TRAINING DINNER

Thursday, May 9, 6:45 P.M.

Presiding—C. W. Wade, Caterpillar Tractor Co.
Co-Chairman—John Grennan, University of Michigan, Ann Arbor, Mich.
Speaker—J. E. Goss, Brown & Sharpe Mfg. Co., Providence, R. I.—*A Means to Produce Skilled and Leadership Personnel.*

A.F.A. ALUMNI DINNER

Thursday, May 9, 7:00 P.M.

SAND SHOP COURSE (SESSION 4)

Thursday, May 9, 8:00 P.M.

Presiding—E. E. Woodliff, Foundry Sand Service Engr. Co., Detroit.
Subject—*Gray Iron Sand Control.*
Discussion Leaders:

H. C. Winte, Worthington Pump & Machinery Corp., Buffalo, N. Y.

T. W. Curry, Lynchburg Foundry, Lynchburg, Va.

Factors to Consider When Purchasing Core Sand, by H. Louette, Warden King, Ltd., Montreal, A. E. Murton and S. H. Fairfield, Bureau of Mines, Ottawa, Can.

JOB EVALUATION AND TIME STUDY

Thursday, May 9, 8:00 P.M.

Presiding—Robert J. Fisher, Falk Corp., Milwaukee.
Co-Chairman—F. E. Wartgow, American Foundrymen's Association, Chicago.
Round Table Discussion on Time Study Practices.
Discussion Leaders:

Phil Carroll, Jr., Maplewood, N. J.

J. C. Lewis, Associated Engineers, Inc., Fort Wayne, Ind.

APPRENTICE TRAINING

Thursday, May 9, 8:00 P.M.

Presiding—C. W. Wade, Caterpillar Tractor Co., Peoria, Ill.
Co-Chairman—John Grennan, University of Michigan, Ann Arbor, Mich.

Passing On the Know-How, by R. L. Lee, General Motors Corp., Detroit.

SAND SHOP COURSE (SESSION 5)

Friday, May 10, 9:30 A.M.

Presiding—R. H. Jacoby, Key Co., East St. Louis, Ill.
Subject—*Steel Foundry Sand Control.*
Discussion Leader—E. E. Woodliff, Foundry Sand Service Engineering Co., Detroit.

GRAY IRON

Friday, May 10, 10:00 A.M.

Presiding—H. Bornstein, Deere & Co., Moline, Ill.
Co-Chairman—V. A. Crosby, Climax Molybdenum Co., Detroit.
Effect of Heat Treatment on the Endurance Limit of Alloyed Gray Cast Iron, by T. E. Eagan, Cooper-Bessemer Corp., Grove City, Pa.
Engineering Properties of Heat Treated Cast Irons, by J. S. Vanick, International Nickel Co., New York.
Discussion of Charts—*Cupola Operations Handbook,* by R. G. McElwee, Vanadium Corp. of America, Detroit.
Theories of Grey Cast Iron Inoculation, by H. W. Lownie, Jr., formerly Westinghouse Electric Corp., East Pittsburgh, Pa.

STEEL

Friday, May 10, 10:00 A.M.

Presiding—C. H. Lorig, Battelle Memorial Institute, Columbus, Ohio.
Co-Chairman—J. A. Rassenfoss, American Steel Foundries, East Chicago, Ind.
Application of Hardenability Requirements in the Steel Foundry, by G. Vennerholm, Ford Motor Co., Dearborn, Mich.
Methods of Quality Control in the Foundry, by J. W. Juppenlatz, Lebanon Steel Foundry, Lebanon, Pa.
Committee Report—*Production of Steel for Castings,* by C. E. Sims, Battelle Memorial Institute, Columbus, Ohio.
Conventional vs. Washburn Risers for Steel Castings, by W. E. Jones, American Steel Foundries, Bettendorf, Iowa.

STEEL

Friday, May 10, 2:00 P.M.

Presiding—J. W. Juppenlatz, Lebanon Steel Foundry, Lebanon, Pa.
Co-Chairman—J. B. Caine, Sawbrook Steel Castings Co., Cincinnati.
Low Temperature Properties of Certain Steels, by C. E. Sims and F. W. Boulger, Battelle Memorial Institute, Columbus, Ohio.
The Production of Grand Slam Bomb Castings, by Basil Gray, English Steel Corp., Sheffield, England—Institute of British Foundrymen Exchange Paper.
The Susceptibility of Steel to Hot Tear Formation in Casting, by N. B. Gelperin, Moscow Div., State Institute, "Giprotjagmashe," Moscow, U.S.S.R.
The Technological Principles of Casting Design, by Victor M. Shestopal, Stalin Machine Tool Institute, Moscow, U.S.S.R.
Committee Report on Non-Destructive Testing, by C. W. Briggs, Steel Founders' Society of America, Cleveland.

GRAY IRON WELDING

Friday, May 10, 2:00 P.M.

Presiding—E. C. Jeter, Ford Motor Co., Dearborn, Mich.
Co-Chairman—John Crowe, Air Reduction Sales Co., New York.
Metallurgical Aspects of Cast Iron Repair Welding, by J. A. Blastic, Detroit Diesel Div.; J. A. Griffin, Pontiac Motor Div.; and J. M. Diebold, General Motors Truck and Coach Div., General Motors Corp., Detroit.
Physical Aspects of Welding Gray Iron Castings, by W. Pfander, Ford Motor Co., Dearborn, Mich.
Welding Heavy Gray Iron Sections, by L. J. Larsen, Allis-Chalmers Mfg. Co., Milwaukee.

BRASS AND BRONZE

Friday, May 10, 2:00 P.M.

Presiding—G. P. Halliwell, H. Kramer & Co., Chicago.
Co-Chairman—A. H. Hesse, R. Lavin & Sons, Inc., Chicago.
Distribution of Mechanical Properties in Sand Cast Bronzes, by R. H. Brouk, R. G. Hardy, and B. M. Loring, Naval Research Lab., Washington, D. C.
Committee Report on *Manganese Bronze,* by G. K. Dreher, Ampco Metal, Inc., Milwaukee.
Manganese Bronze, by H. E. Arblaster, Stokes & Sons, Melbourne, Australia, Institute of Australian Foundrymen Exchange Paper.

LECTURE COURSE (SESSION 5)

Friday, May 10, 4:00 P.M.

Presiding—W. H. Rother, Buffalo Foundry & Machine Co., Buffalo, N. Y.
Co-Chairman—J. S. Vanick, International Nickel Co., New York.
Subject—*Gray Iron Foundry Control.*
Discussion Leader—W. A. Hambley, Allis-Chalmers Mfg. Co., Milwaukee.

ANNUAL DINNER

Friday, May 10, 7:00 P.M.

Presiding—F. J. Walls, President, A.F.A.

The Annual A.F.A. Dinner of the Association's 50th Anniversary Convention was called to order by President F. J. Walls, presiding. Following entertainment by the Ramblers Quartet, arranged through the Northeastern Ohio Chapter, President Walls introduced Past President R. J. Teetor, Chairman of the Board of Awards.

Mr. Teetor:

"I am very happy that this 50th Anniversary Convention fell to the lot of Pres. Walls, because it is the greatest convention the American Foundrymen's Association has ever had and if any man deserves the distinction of presiding at that convention, it is Fred Walls.

"I do have a duty that is very pleasant this evening, one I have looked forward to and one that I am going to enjoy. Before introducing the gentleman who will present the various awards for 1946, I would like to review briefly the make-up and function of the Board of Awards of the American Foundrymen's Association.

"The Board of Awards consists of the seven most immediate Past Presidents of the Association. It is their duty to consider the achievements of outstanding men who have been active in some important phase of the foundry industry, and to evaluate those achievements as accurately and scientifically as may be possible in order that determination may be made among those who should properly be selected for the honor of receiving those awards which the generosity of their donors has placed at the disposal of the Board of Awards for this purpose.

"The award of a Gold Medal of A.F.A. carries with it the distinction of Honorary Life Membership in the American Foundrymen's Association. For outstanding service to the American Foundrymen's Association, either within its organization or in connection with the foundry community at large, the award of Honorary Life Membership also is available for presentation to others who are not medal recipients. This year the Board of Awards had made available three Gold Medals which carry with them Honorary Life Membership in the American Foundrymen's Association, and two separate Honorary Life Memberships."

Mr. Teetor then introduced Max Kuniansky, Vice-President-elect and 1941 William H. McFadden Medalist, who presented the 1946 William H. McFadden Medal to H. Bornstein. Mr. Kuniansky:

"Five years ago on this date Mr. Bornstein presented me with the Wm. H. McFadden Medal, and it is indeed a source of great pleasure and satisfaction for me thus to be honored with the privilege of presenting him with the same medal.

"Most of you of the American Foundrymen's Association and the foundry industry know Mr. Bornstein for his splendid achievements over the years, for his excellent judgment, for his many contributions in the technical fields of the gray iron foundry industry.

"To try to elaborate too much on his achievements is hardly possible. His record speaks for itself. He has stood in the American Foundrymen's Association and in the industry for the best traditions that we honor.

"Hyman Bornstein, Director of Testing and Research Laboratories, for Deere & Co., Moline, Ill., and a past President of A.F.A., is awarded the William H. McFadden Gold Medal of the Association,

'For his many and outstanding technological contributions to the foundry industry, particularly in the field of gray cast iron.'

"Recognized internationally as one of the world's leading metallurgists and authorities on cast iron, Mr. Bornstein has long been most active in A.F.A. technical affairs. Mr. Bornstein, or 'Hy,' as he is universally known by his many associates and friends, was thoroughly grounded in A.F.A. work, for since going with Deere & Company immediately after World War I, when he served as Captain in the United States Army Ordnance, he has been a most valued and consistent contributor and worker on A.F.A. committees in all branches, especially in the malleable and cast iron divisions. His work was early recognized, for in 1924 he became a member of the Joint Committee on Molding Sand Research. Later he was a member of the Foundry Refractories Committee, and in 1927 he served on the Cast Iron Research Committee and the Malleable Division Executive Committee, being made Chairman of the Gray Iron Division in 1929. Among some of his outstanding activities was his sponsoring of the Joint Symposia with the A.S.T.M. on Cast Metals, which was really the start of the A.F.A. Cast Metals Handbook, in which work Mr. Bornstein was especially active, being one of the leading editorial writers for the initial issue of the Cast Metals Handbook in 1935.

"In addition to his A.F.A. technical work, he holds membership in and has served as a most valued member on committees of the American Society for Testing Materials, American Society for Metals, American Society of Mechanical Engineers, American Chemical Society, and the Society of Automotive Engineers.

"Because of his outstanding work in A.F.A. committees, he was first elected National Director in 1932 and served for a term of three years, going off the Board in 1935. However, because of his outstanding ability, he was recalled to even higher honors by being elected National Vice-President and President in 1937, later again serving as Director in 1937-41. As a past president he has continued his activities, serving for 7 years on the Board of Awards until a year ago. During his service with the Board of Awards, his colleagues, realizing his great service to the industry, frequently wished to recognize him by awarding one of the A.F.A. Medals to him, but Hy, who is always modest and with the sixth sense of propriety, always refused to have his name mentioned.

"Since serving as President, he has also continued being very actively connected with the technical work of A.F.A., and now serves on the Annual Lecture Course Committee, Gray Iron Division Executive Committee, and on other committees too numerous to mention. His guiding hand was found in all of the outstanding activities and achievements, and his valued counsel is sought by all of his colleagues.

"It is my pleasure and privilege to present to you, Mr. Bornstein, the William H. McFadden Medal of A.F.A. for your outstanding achievements in advancing the gray iron foundry industry."

Mr. Bornstein replied:

"I have had the privilege the past few years to present some of these Medals as a member of the Board of Awards, and it certainly was a source of satisfaction to me five years ago to present the William H. McFadden Medal to Mr. Kuniansky. I hope that he received as much pleasure presenting this medal to me as I enjoyed five years ago. It is said that it is more blessed to give than to receive, but I must admit that on this particular evening I am very happy to be on the receiving end.

"In accepting the William H. McFadden Medal, I want to refer to the men who have worked with me over the years. I refer particularly to men who have been associated with me at Deere & Company, to those who have served with me on American Foundrymen's Association committees, and to the members of the American Foundrymen's Association Headquarters staff, particularly Messrs. Hoyt and Kennedy.

"I assure you that I sincerely appreciate the high honor that has been bestowed upon me and I want to express my thanks to the Board of Awards."

Mr. Teetor then introduced L. N. Shannon, Past President of A.F.A., who presented the John H. Whiting Medal to Mr. Peter Blackwood. Mr. Shannon:

"Our John H. Whiting Gold Medal Award is presented this year to a man in recognition of his development work on centrifugal castings and his influence on the free interchange of research data.

"He was born in Scotland and received his formal education in metallurgy, metallography, foundry practice, chemistry, mining and geology in that country. After coming to America, he entered the field that enabled him to gain experience and knowledge of the electric furnace converter, getting his furnace practice in the United States and in Canada, after which he held the position of Superintendent of Foundries at Buick and Pontiac Divisions of General Motors Corporation. While with General Motors he delivered weekly lectures on foundry practice, heat treatment and general metallurgy at the General Motors Institute of Technology.

"He then became associated with the Ford Motor Company of Canada at Windsor, Ontario, where he still is located. At Ford's Canadian plant he has conducted a great deal of experimental and production work upon the centrifugal casting process. He has given of his experience and knowledge freely for a number of years.

"It was in 1914 that he gave his first paper at the Chicago Convention of the American Foundrymen's Association, and he has been actively interested in the work of the American Foundrymen's Association ever since that time. He presented a paper at the Cleveland Convention in 1916. In 1944, at the Buffalo Convention, a paper he presented was one of the outstanding contributions to the foundry industry. It was one of the best and most comprehensive discussions on centrifugal castings. This paper showed amazing developments in the centrifugal castings technique in many fields. Again in 1944, another paper co-authored by his assistant on the subject of the centrifugal casting process excited widespread comment and was instrumental in directing the attention of many foundrymen to the advantage of this process.

"Because of this stimulation of broader foundry work on centrifugal castings, and because he has given freely of his information through talks before the American Foundrymen's Association Chapters throughout this country, we therefore confer on Peter Blackwood at this, our 50th Annual Convention, the John H. Whiting Gold Medal of the Association."

Mr. Blackwood replied:

"How can I express my appreciation for this honor which the American Foundrymen's Association has so graciously bestowed upon me? As a matter of fact, it is not as an individual by the name of Blackwood that I am thrilled by this presentation. Rather, it is the carrying of this medal over the border to God's own country, where the Canadian foundrymen in general gave me a newness of life, new hope, new powers, new inspiration, new courage . . . not formal, but vital; not artificial, but spiritual.

"I will always remember that day about 40 years ago when I landed on the golden shores of America representing an old Scotch family in the making of pig iron. I remember it more because on this day of days, the 50th Anniversary of A.F.A., the American Foundrymen's Association in its brief course has brought to me the realities and possibilities of my own existence. For yesterday is only a dream and tomorrow is only a vision, but today, when we live in the realm of pure, unadulterated brotherhood, makes every yesterday a day of happiness and every tomorrow a vision of hope.

"So once again I thank thee from the bottom of my heart for this honor which has been bestowed upon me. It is singular that during the hectic days of the late war, when we were helping out Uncle Sam over the border in jet propulsion, it reminds me of the Whiting people, who made this medal possible, because its spirit must have been with us to enable us to turn an electric furnace into a side-blown converter.

"Tonight I feel as though I had been turned outside inmost and want to thank you again from the depths of my heart for this honor."

Mr. Teetor then said:

"Now I have the pleasure not only of introducing the gentleman who will present the next award, but also of introducing the Award itself. And I may well begin by saying to you that the acceptance of an endowment for an official medal to be added to the group of similar awards, which have been made available for presentation by the Boards of Awards since 1924, is a subject that has had the most careful study by your Board of Directors and most careful scrutiny by a special committee appointed for that purpose.

"The endowment of a medal, to be acceptable to the precedents and ideals of the American Foundrymen's Association, must not only confer honor upon the recipient as selected by the Board of Awards, but must by its presentation confer honor upon the donor. In the Peter L. Simpson Memorial Medal, the honor is not alone the memory of one of the pioneers in the upbuilding of the foundry industry in the Middle West, but also one of the early builders of improved equipment designed to lighten the labor of the foundryman and to improve the quality of his product.

"There is a further and more potent reason for my especial pleasure and gratification in having the privilege of introducing

the Peter L. Simpson Memorial Medal, and that is the tradition of faith in the greatness of the industry as well as the superior knowledge of his technique which was exemplified by Peter L. Simpson during his eventful and busy life. Added to that was a faith in the integrity of the foundryman himself which Peter L. Simpson exhibited in his dealing with the industry and with the members of the industry, beyond the call of merely good business principles.

"It is my privilege to introduce to you Herbert S. Simpson, Past President of the American Foundrymen's Association, who has made possible the institution of the Award which bears the name of his father and who will make the initial presentation of the Medal, the Peter L. Simpson Medal, to Howard F. Taylor."

Mr. Simpson:

"It is my pleasure to present this Medal. First of all, I want to say that any man must earn his own way in this world. I think we have to learn that these days. There can be no doubt that the reward of life comes only out of a man's own effort. We are trying to teach that in some of our political life today. I think that accurately describes our medalist, Howard Taylor. I will give you just an idea of what I mean by his making his own way.

"Mr. Taylor is a comparatively young man. He was born in 1913 in the State of Michigan, and lived on a farm until he was 11 years of age. When he was 12, his parents moved to a small town in Michigan, where he was a newsboy, working nights and Saturdays in a tire and battery shop. From the age of 13, he worked for five years for the Michigan State Highway Department for four months each summer, and in miscellaneous stores during the school months.

"He attended high school at Elk Rapids, Michigan, where he worked in a hotel for his board and room, until the time he was 18 years of age. He then enrolled at Michigan State College and for the following two years worked nights and week-ends in the college foundry where he molded, melted metal, poured, shook out, tumbled and cleaned all shop castings. They made castings from different metals with the exception of steel, and later Mr. Taylor got into the steel casting field. Later on, he received his B.S. Degree in Chemical Engineering at Michigan State College and then returned for a year to post-graduate work, during which time he worked as graduate assistant in metallography and gas and fuel analysis.

"In 1937 he became associated with the Naval Research Laboratory in Washington as Assistant to the Welding Engineer. Under the able influence of Messrs. Charles W. Briggs and Roy A. Gezelius, he became affiliated with them in the steel casting section and after they left he became head of that section of the Naval Research Laboratory. Later, he became dominant in development of various metals at that laboratory.

"From that time on, his interest in the foundry field led to both ferrous and non-ferrous metals. However, he was always seeking knowledge, so in 1945 he went to Massachusetts Institute of Technology with a leave of absence from the Naval Research Laboratories, to do graduate work. Incidentally, part of his expenses are defrayed by the Gray Iron Founders' Society and the balance of his expense he makes up by his own consulting work.

"He was married in 1939 and now has two potential little foundrymen, one for each side of the ladle.

"His work has been outstanding, particularly his determination and persistence to acquire knowledge, which exemplifies the medal presented to him.

"It is my great honor and pleasure to present, for the first time, the Peter L. Simpson Gold Medal of A.F.A. to Howard F. Taylor."

Mr. Taylor replied:

"In accepting this Peter L. Simpson Award, I do so in appreciation of the work, cooperation and efforts of my colleagues at the Laboratory and the moral support and encouragement of every man with whom I have come in contact in the foundry industry."

"I think that all I need to say is that this is one of the proudest and happiest moments of my life. Thank you."

Mr. Teetor then presented Honorary Life Membership to Harold J. Roast, saying:

"I have the privilege of presenting the Honorary Life Membership to Harold J. Roast in recognition of his many long and continued contributions to the literature and cooperative enterprise of the foundry industry, both in the United States and Canada.

"It is not especially important to report that Harold Roast is Vice-President of the Canadian Bronze Company, Ltd., with its several subsidiaries, and that he is owner of the Roast Laboratories of Montreal, but rather to stress the fact that he has used the opportunity that these distinctions have afforded him to give as much of his time, and overtime, to the preparation of technical and scientific contributions to the technical press of Canada and to the United States. He has for many years served McGill University of Montreal as a lecturer on subjects of metallurgical interest, and he has served the American Foundrymen's Association as Chairman of the Non-Ferrous Division. In Canada he was a stalwart proponent of American Foundrymen's Association interests and is a Past President of the Eastern Canada and Newfoundland Chapter.

"Born in London, England, Harold Roast, notwithstanding his many years in Canada, has retained his active interest in technical developments in England through his Fellowship in the Chemical Society of London.

"In recognition of outstanding achievement in the field of education and literature and scientific pursuits, I have great pleasure in presenting to you, Harold Roast, on behalf of the American Foundrymen's Association, this Honorary Life Membership."

Mr. Roast replied:

"I feel that I should pay tribute to the spirit of unselfishness and good neighborliness that I have found characteristic of American foundrymen in my many years of association with them, and particularly of the recognition which they so generously give to the small minority of Canadian foundrymen.

"The American Foundrymen's Association provides the Canadian foundrymen, particularly the young men, with many opportunities for work and for help. We will do our part in reciprocity by working all we can for the American Foundrymen's Association.

"One thing that I think has helped so much the prosperity of the Association is the fact that in the American Foundrymen's Association, everyone has the opportunity of work and the opportunity of reward. I think we cannot do better than think of Carlisle in connection with this matter of work when he said, 'For work is the grand cure for all the maladies and miseries of life.' But he is careful to enter this postscript: 'Honest work that you intend getting done.'

"Mr. Chairman of the Board of Awards, I would again thank you and the Board for the consideration given to me and for the honor which has been conferred upon me."

Mr. Teetor then introduced Past President L. C. Wilson, who presented the Honorary Life Membership certificate for William J. Coane in these words:

"It is a privilege and an honor for me to present Honorary Life Membership of the American Foundrymen's Association to William J. Coane at the 50th Anniversary Celebration of the American Foundrymen's Association. He is an original member of the American Foundrymen's Association and has continued his affiliation for the 50 years of its existence. He is one of the few living members who attended the first Convention of the Association in 1896, and was active in the local group whose efforts made that meeting possible.

"Mr. Coane was born in Philadelphia, attended the Friends School and finished his education at Central High School. He began his active labors of his career with the Joseph S. Dixon Crucible Company at Jersey City, New Jersey, serving as Manager of the Company's Philadelphia division for 26 years. In 1912, Mr. Coane became Vice-President of the Ajax Metal Company, a position he held continuously until his recent retirement as Vice-President and Director.

"His contributions to business and community life have been winning for him wide admiration and confidence for years. For 15 years he was a member of the Executive Committee of the Non-Ferrous Ingot Metal Institute of Chicago, and for many years has represented that organization in the capacity of National Councilor in the Chamber of Commerce of the United States.

"We regret that Mr. Coane's illness made it impossible for him to receive the Association's recognition in person. However, he has sent an able substitute to represent him and receive the award on his behalf. It is my pleasure, therefore, to introduce Mr. Coane's associate of many years, Dr. G. H. Clamer, president of the Ajax Metal Co. and a Past President of A.F.A."

Dr. Clamer accepted the award on behalf of Mr. Coane with the following remarks:

"Mr. Coane, or 'Uncle Billy,' as he is affectionately known to many of you, has been a business associate of mine for over 30 years. We have been friends beginning with about the time the American Foundrymen's Association was organized in Philadelphia, 50 years ago.

"Mr. Coane now is 79 years old. He is very ill at this time

and greatly regretted that he could not be here. He was recently forced to retire from active business. It was at a time when he was feeling mighty low that the word came to him that he was to be honored by his friends in the American Foundrymen's Association. I leave it to you to imagine the great joy and thrill it was to him under such circumstances to receive word of that kind.

"He has sent me the following message for you:

" 'Kindly express my sincere appreciation to the Board of Awards, to President Walls, and any others who have been responsible for conferring upon me Honorary Life Membership in the American Foundrymen's Association. It is a great satisfaction to me to know that the Association has carried on so successfully for 50 years. I am proud to have been one of the original organizing groups. I deeply regret my inability to be present at the Golden Anniversary to receive in person the honor conferred upon me, but extend to the members my heartiest wishes for the continued success of the American Foundrymen's Association.' "

Mr. Teetor then turned the meeting back to President Fred Walls who introduced the guest speaker, Brig. Gen. Donald M. Armstrong as follows:

"Our speaker's subject, 'The Foundry In War and Peace,' is of paramount interest to everyone connected with the foundry industry, for certainly no industrial group contributed more to the war effort than did the producers of castings. Our speaker is exceptionally qualified to discuss this subject, not only because of his meritorious career in the United States Army, which extends over a period of 35 years, but because he is today responsible for re-establishing liaison between the War Department and industry.

"A graduate of Columbia University, Commissioned as Second Lieutenant in the Coast Artillery in 1910, he served in the first World War on the Staff of the Chief of Artillery of the American Expeditionary Forces and with the Fourth French Army. Following the war, he was detailed for four years as Assistant Military Attache at the American Embassy in Paris and is a graduate of the French Ordnance School.

"Transferred to the Ordnance Department, he served as Instructor at the Army Industrial College during the period of 1927 to 1931. During the years 1931 to 1935, he served as a member of the Field Artillery Board, and then was on duty in the Office of Chief of Ordnance as chief of Maintenance Division the following four years.

"On July 2, 1939, as Lieutenant Colonel, our speaker was appointed Deputy Chief of the Chicago Ordnance District. Within three months he was promoted to Colonel and on May 20, 1942, received the well deserved and coveted star of a Brigadier General. The following September the General was transferred to Detroit to establish the tank automotive center, and three months later was appointed Commanding General of the Ordnance Replacement Training Center, Aberdeen Proving Grounds, Maryland.

"With the attention of the world focused on organizing for peace to prevent future wars, our speaker comes to us at a mostly timely moment. Outstanding in his military career is an exceptional understanding of the potentiality of industry and a complete knowledge of its problems. Most important of all, he has acquired the admiration of all industrialists who have had the good fortune to have had contact with him.

"Today he is charged with the responsibility for peace-time planning for the requirements of war, and his is the responsibility for establishing that highly important liaison between industry and the Industrial Division of the Army.

"I take pleasure in introducing to you one of the foundry's best friends, the Commandant of the Army Industrial College, Washington, D. C., Brigadier General Donald Armstrong."

Brig. Gen. Armstrong then presented his address entitled "The Foundry in War and Peace."

On conclusion of the talk by Brig. Gen. Armstrong, President Walls expressed his appreciation for the General's encouraging and stimulating words and turned the meeting over to T. N. Johnston of the Northeastern Ohio Chapter, Chairman of the Banquet Committee, for a brief message. The banquet meeting adjourned at 10:18 P.M., followed by an evening of dancing.

President's Annual Address

By Fred J. Walls,* Detroit

THIS, our "Golden Jubilee Convention" and our seventh convention in the great city of Cleveland, is more than a celebration of the fiftieth anniversary of the founding of the American Foundrymen's Association. It marks a half-century of accelerated progress in the castings industry unmatched in all the previous 50 centuries or more of its history. This is the golden jubilee of a renaissance, the anniversary of a transition from rule-of-thumb to an industry built upon solid scientific foundations, upon chemistry, upon physics, upon metallurgy, upon mechanical engineering, and upon techniques which modern research is developing daily.

It would not be possible, nor would it appear necessary, to attempt to recount each of the great advances the foundry and its allied industries have made over the past 50 years. They are certainly vivid in the memory of the many here who have served 50 years in the foundry industry; and the youngest among us can appreciate how great has been this transition that brought founding from a mysterious art, fortified by little more than melters' and molders' secrets, to a scientific plane on which it is equal to the most exacting requirements of production for war or for peace.

As members of A.F.A. we are naturally, and justly, proud of the progress in the industry we serve. And we are proud, too, of the fact that the half-century of our industry's greatest advances parallels in time the first 50 years of our Association. This is not to claim, however, that A.F.A. deserves the credit.

All credit for the great progress of the industry belongs to the men of the foundry industry and to the men allied with it, eager for its constant improvement, and to the ever-growing number of tireless workers within and without the industry who have been unceasing in their researches to its advantage. But proud, indeed, we may be of this: These men are A.F.A.

1896 Resolution

In that month of June of 1896 when 345 foundrymen gathered in Philadelphia and organized the A.F.A., they resolved that "it is desirable that the interests of the foundry business be furthered by a closer association among all the foundrymen of America in order that any improvements or valuable information concerning the business, or any portion of it, should become as widely available as possible."

A year later at the Association's second annual convention, held in Detroit, its first president, the late Francis Schuman, said, "We are here to formulate ways and means toward the advancement of our industry, both as a mechanic art and a source of benefit to those engaged in its practice."

Mr. Schuman further stated, "The continued growth in exceptional size, quality and form of foundry products to meet the special purposes intended by the engineer, compel the application of a knowledge of the science far beyond any that our predecessors ever thought of; and not only does this apply to the leader, whether superintendent or foreman, but equally to the handicraftsman in all branches of our business." Mr. Schuman's statement 49 years ago is perhaps more imperative today than it was then.

A more formal expression of the objectives of the A.F.A., as outlined in its articles of incorporation, is, "to promote the arts and sciences applicable to metal casting manufacture and to improve the methods of production and the quality of castings, to the end that the increasing utility of all classes of castings may result advantageously to all persons engaged in the foundry and related industries and to all users of foundry products."

In its 50 years of service to the castings industry, A.F.A. has never lost sight of the objects for which it was organized. It has grown tremendously and parallels the progress made in the great industry it serves.

Its most important work continues to be the maintenance of media for the free interchange of technical information, the encouragement of research and unrestricted discussions of new developments, the cultivation of skills and the dissemination of knowledge among those practicing the scientific art of founding. In short, its purpose is the promotion of the science of founding.

The Transactions, chronological records of foundry progress and practice over 50 years, whose pages total over 31,000, is evidence of how well your Association has served as a clearing house of information. And I dare say, without boasting, that there is not a single major development in foundry practice they do not record.

Out of the effort of A.F.A. to disseminate information of practical lasting value to the industry, and for those who will enter it, has grown a foundry technical library of some 5,000 pages. The most recent addition is a 500-page publication of the Cupola Research Project—The Cupola Operations Handbook.

This book is the work of over 125 outstanding gray iron metallurgists and practical foundrymen. It is the most complete and modern reference ever published on this most used type of melting furnace. Supplementing this library are numerous other publications, symposia, committee reports, and special studies with which you are familiar.

American Foundryman

I should like to say a few words about our monthly

*International Nickel Co., Detroit.

Note: Presented at the Annual Business Meeting at the Fiftieth Annual Convention of A.F.A., in Cleveland, May 9, 1946.

technical magazine, AMERICAN FOUNDRYMAN; and in this reference I am certain we are justified in unlimited boasting. I am proud of AMERICAN FOUNDRYMAN in its new dress, proud of its wide acceptance internationally as well as nationally, proud of the reader interest it commands, proud of the staff who made it what it is and of their future plans; and I cannot help but think that everyone of you are proud of it, too.

The working committees of A.F.A. are the foundation heads of the vast stream of technical and practical information which continues to be channeled with increasing volume to our membership, to schools, to other societies and to all who are interested in the production and use of castings. On these committees over 500 individuals are giving willingly and freely of their time, their skill and their experiences in appreciation of the fact that there is no real progress unless all progress.

We salute these men and sincerely thank them for their invaluable contributions to the great progress of the castings industry and to the assurance they give that even greater strides will be made in the years to come through their collective efforts.

The willingness of men to serve on A.F.A. committees is notable and significant. As early as 1910 Arthur T. Waterfall, then president, commented at the Association's fifteenth annual convention, "Those of you present, who can look back to the earlier years remember the difficulties encountered in getting men with courage to tell of their experiences and their successes in producing a better class of work at less cost, will note today the eagerness of foundrymen to give this information to their chosen industry. No longer do they surround the foundry industry with the air of mystery that formerly enveloped it, realizing now that in the interchange of ideas we are mutually helpful."

Truly, we are celebrating here today the Golden Jubilee of a renaissance, for this philosophy is truer today than it was then.

From the first days of its organization, the Association has cooperated in the problem of training men from the elementary stages of apprenticeship through to compiling and disseminating information constantly sought by students and teachers alike. By providing apprentice awards, it has recognized and given impulse to the importance of training young men in the art and science of founding.

The Cost Committee, one of our oldest committees, has constantly emphasized the importance of sound cost determination and given impetus to the excellent cost systems developed by various foundry trade associations.

8300 Members

Membership growth of A.F.A. has been nearly a straight line since the formation in 1934 of the first of our 33 Chapters. Today, our membership is near 8,300 (to be more exact, 8,285), which is a record high. That the interest in A.F.A. extends throughout the world is attested by the number of foreign members and guests attending this convention.

The leadership which your Association assumed in establishing an exchange of technical information with foundry associations in other countries is renewed and expanded at this convention. Attending this convention are representatives of 19 foreign countries and the presence of these visitors from abroad, many of them A.F.A. members, emphasizes the international aspect of foundry cooperation that forms the basis for your Association's progress.

On Monday, this convention received a paper from the Foundrymen's Association of Czechoslovakia. Today and tomorrow, exchanges from France and England will be presented. The first of a new series of exchange papers, that of the Institute of Australian Foundrymen, will also be presented.

We are happy to cooperate with our friends "down under" and extend to their membership our sincere good wishes. Among my hopes is one for the continuation of our International Foundry Congress. Later today an Organizational Committee is meeting to consider future plans.

Exhibit Space

Everyone here has viewed, and will see again, the interesting and instructive exhibits of foundry equipment, materials and services which fill the six halls covering eight acres of this Auditorium. To the 280 exhibitors at this Golden Jubilee Foundry Show, we express our appreciation and we sincerely hope that their efforts have been worth while.

To the members of the Northeastern Ohio Chapter, our host chapter, we express our sincere thanks for all that you have done in helping us make this event the success it has proved to be.

The technical program, comprising nearly 150 papers and addresses, panels and round table discussions, assumes gigantic proportions when compared with the technical program of our first meeting at which less than half a dozen were presented.

For 50 years it has been the privilege of the presidents of your Association (43 of them) to summarize for the membership the activities of the fiscal year. So many things have happened since last July that I am sure I would be abusing this traditional privilege were I to attempt to mention all of them.

I assure you, however, that your Executive Committee and your Board of Directors have carried out faithfully, ably, and without hesitation, all the responsibilities you have entrusted to them. And each of their actions was guided by one question alone: What will be for the best interests of the foundry industry and the American Foundrymen's Association?

No Executive Committee or Board of Directors ever hopes, of course, to satisfy completely the unstudied desires of more than 8,000 individuals. We have appreciated most of the criticism received because it has been constructive and will be helpful in future decisions of new boards.

Unpredictable circumstances made it necessary to call several special meetings of the Executive Committee in addition to the regular meetings held every two months. As president, I should like to acknowledge my sincere appreciation for their time, their patience and their sound advice.

During the fiscal year, we saw the cessation of hostilities, the end of World War II, and with it the turmoil of reconversion. The uncertainties imposed upon the castings industry by these events have been reflected in the work of the Association and have added tremendously to the burdens of our staff.

This, coupled with the growth in membership and our constant desire to increase our service, necessitated several additions to the headquarters' staff in Chicago. To these new men and women, and to Ed Hoyt and Bob Kennedy, and to our Secretary, Bill Maloney, praise and credit for the success of this convention are accorded. We are proud of the job they have accomplished under very difficult conditions.

Technical Director Hindle

Early this year, Norm Hindle was taken seriously ill and, while he is coming along nicely, his doctor would not permit him to attend this convention. I know that Norm misses being here as much as all of us miss having him here to participate in the fruits of his untiring efforts.

On March 15th of this year, Miss Jane Reininga, the Association's assistant treasurer, retired after active service of 30 years as a staff member. She joined the staff at the time Ed Hoyt was first elected secretary and served successively as stenographer, bookkeeper, office manager, and was appointed assistant treasurer in 1941. We miss her in the many positions in which she had so faithfully and tirelessly served, and we wish her every measure of happiness and the best of health in her retirement.

To meet the challenges from within and without the foundry industry, your Directors authorized the Secretary to proceed with enlarged educational, research, publication and publicity programs, most of which are well under way. The details of these plans will be further outlined at the forthcoming Annual Chapter Chairmen's Conference to be held on July 24-25, prior to the Annual Board Meeting.

A.F.A. Financially Sound

I am more than pleased, and exceedingly proud, to be able to report that, through your cooperative effort, the Association is stronger financially than ever before. Its financial statements reflect not only greater activity and good management in your interest, but also a greater appreciation among foundrymen and the men of allied fields of the dollars-and-cents value of cooperative effort along technical and educational lines, plus a strengthened confidence that by working together we progress and by keeping together we achieve the success we are seeking.

The 1945-46 budget of the American Foundrymen's Association, with a provision for surplus, totals above one-third of a million dollars. Reflecting steady growth, greater service, wider participation, it is the largest budget in the Association's history. More than that, it is conservative. Giving greater service, we are living well within our income; and we are financially able, with our service value-tested, to put forward a greater cooperative effort.

In the 1945-46 budget, income from membership dues for the fiscal year ended June 30th was estimated at $135,000.00. In the first 10 months of the fiscal year dues have actually totaled $148,444.85. Anticipated advertising revenue from AMERICAN FOUNDRYMAN for the fiscal year was placed at $75,000.00. As of May 1st sales of advertising totaled $82,440.76.

Receipts from sales of exhibit space and from exhibit permits is also greater than anticipated. In short, income from all sources in the first 10 months of this fiscal year has been about 16 per cent greater than the total anticipated for the full year.

I have pointed out that we are living well within our budget. As a matter of fact, estimated total cost of operation for 1945-46 has proven, by the experience of 9 months of actuality, to be as high as estimated income has proven to be conservative. As of March 31st, we were within the budget for the full year by better than $158,000.00.

You will be interested in learning that the 1945-46 budget underestimated the item of chapter refunds. For the first 9 months of the 1945-46 fiscal year refunds to chapters have amounted to $25,333.76, an average of more than $2,800 a month, against an estimated average of $2,250 a month. This reflects increased membership, in addition to the increase in chapter refund rates as provided in the new dues structure which went into effect last July.

I regret, exceedingly, that I was unable to visit all the chapters during my term as president. My intentions, I assure you, were excellent, but travel conditions and Association headquarters' demands prevented. You will pardon me if I say that these same excuses apply also to Secretary Bill Maloney.

However, we managed to visit most of the chapters, and those visits were valuable to us in that they afforded an opportunity to discuss with the chapter board members our mutual problems and to take back to the National Board many excellent suggestions of benefit to all the membership.

Regional Conferences Helpful

It was my good fortune, also, to be able to attend four Regional Conferences. They impressed me greatly as educational media and as effective means for widening an appreciation of the foundry industry and its best practices.

I have given you a very sketchy report, but I have purposely hurried in order that I might impose upon your indulgence to give you some dreams as to the future of the American Foundrymen's Association and the great industry it serves. As we pass this fiftieth birthday marker, we can only imagine the magnitude of the one hundredth marker.

We must borrow some time from our present problems and utilize it for the placing of more permanent beacons if we are to guide our ship through the unknown currents we are running into and thus avoid drifting in the wrong direction. Under our charter as a technical society we cannot enter into the construction of, nor the placing of, the most important lighthouses that mark the channel to future progress and prosperity in the castings industry.

If we assume that the now dim lights in these beacons will brighten in due course, and we must have faith that they will, then we have only to locate and build one important lighthouse, science and its application.

It is wonderful for an Association to be an integral part of an industry so old that we cannot trace its beginning and yet so young in its scientific developments that we cannot comprehend its future. The mysteries of its birth may be hidden deep in the earth's rocks, but its future progress lies in the individual and cooperative efforts of those associated with it.

The future of the foundry industry and that of this Association is what we make it, and that is dependent upon our concept of what it should be.

I can visualize the A.F.A. membership tripled and with at least 50 chapters. It will not take a half-century to accomplish this, for it can be done in a decade, and I suggest that we pilot our ship toward that beacon. The organization dates of five chapters fall within the year 1945 (Oregon, Saginaw Valley, Northwestern Pennsylvania, Mexico City and Central Illinois).

Obviously, there is room for chapter growth. Ohio, for example, has 522 foundries (as listed by Penton) and five chapters, or a chapter for approximately every 100 foundries in the state, if we disregard some overlapping. If this is an ideal situation, and it seems to be in keeping with the benefits derived from chapter activities, then we should have 50 chapters in the United States alone.

If we add to these the requests we have had recently from other countries, our goal should be even higher. To me, this is a definite indication of the desire for free exchange of ideas, as mentioned earlier.

What about the future foundry, its equipment, its personnel, and its processes? The great developments, stimulated by the requirements of war, already are projected into peace-time production. It seems only logical that, as challenges from developments in other industries appear, they will stimulate developments in the science of casting.

Although they might now seem fantastic, gasless cores and molds, waterless binders, fireless melting units, castings that require no cleaning and with dimensional tolerances close to machining tolerances, all appear possible.

Mechanical equipment in industry to eliminate drudgery will parallel similar developments in the future home. Working conditions in all industries will parallel changing living standards. We all like pleasant dreams and dread horrid ones. Before I expose my dreams to possible ridicule by future generations, I must return to the present.

Let us have confidence in our ability to overcome all challenges and content our minds with the fact that throughout the history of the world the casting process has always survived and progressed with every advance in science, culture, or even revolutions. We must put on a bigger front and have the courage of our own convictions.

In closing, I wish I were capable of expressing to you how grateful I am for the privilege you have accorded me, that of serving you as president of the American Foundrymen's Association during its Golden Anniversary Year. I consider the Presidency of the A.F.A. the highest honor a foundryman can hope for and the highest honor he can receive. I am certain that every one of the 43 men who have preceded me in this high office were as deeply grateful as I am for the opportunity to serve a great Association.

Report of the Secretary

July 26, 1946

By custom, the Annual Report of the Secretary of the Association details the development and current status of all Association activities. Looking back in perspective, the procedure appears as one involving considerable adjustment, beset by many unusual and perplexing problems. For their solutions we wish to express our sincere appreciation and gratitude for the constant and untiring assistance rendered the National Office Staff by President Walls and all members of the Executive Committee: Vice-President Wood, past President Teetor and Directors Wilson, Wallis and Smith. Their assistance and the counsel of Ed Hoyt and Bob Kennedy, has been of inestimable value in our ability to complete the year successfully and to add further to the strength and prestige of the Association.

Even more of a revelation has been the continued and wholehearted support rendered by the membership at large, which brings into sharp focus the good foundation built by those who preceded the present Secretary. It has been his privilege to inherit the accomplishments of the past, as it is his going responsibility to add further to them.

During the year 1945-46 the following activities and developments deserve special mention:

Membership

Some 44 years ago the then Secretary of American Foundrymen's Association presented an Annual Report and pointed with justifiable pride to a total membership of 314. He mentioned specifically members in England, Scotland, Wales, Germany and Holland and stated that other inquiries "may result in extending the usefulness of our Association and go to show that the world is taking note of American foundry practice, and is presumably wondering how we can afford to be so liberal with our information."

As of June 30, 1946, the membership of A.F.A. totaled 8,539, compared with 7,815 on June 30, 1945. This continued phenomenal growth occurred in spite of a substantial increase in certain member dues, and against all predictions that reconversion following the end of the war would result in material membership losses. As a matter of fact, it was estimated that the immediate result of these two factors might very well be a loss of up to 12.5 per cent of our membership; actually, we achieved a net gain of 724 members, or 9.26 per cent.

Of the total of 8,539 members on June 30, 1946, 7,767 or 90.8 per cent were affiliated with Chapters; 772 held no Chapter affiliation, of which 392 reside in countries outside North America. The chart of membership growth for the past 50 years (Fig. 1), a chart prepared for the 1946 Chapter Chairman Conference, illustrates the phenomenal growth in membership since the first Chapter was established in 1934.

For purposes of achieving continuous growth in membership, we should keep in mind a definite goal on memberships during 1946-47. No membership "campaign" as such is contemplated; rather, the concentration will be on (1) converting Company to Sustaining memberships, (2) converting Personal to Company memberships, and (3) increasing the number of Personal (Affiliate) memberships held by Company members.

Chapters

During the past year two new Chapters were installed, the Mexico City Chapter, and the Central Illinois Chapter with headquarters at Peoria, Ill. This brings to a total of 33 the number of A.F.A. Chapters established since 1934, although one Chapter is now inactive—the University of Minnesota Student Chapter. It is hoped that this Student Chapter can be reorganized as greater interest is shown in foundry work at Minnesota.

There are good possibilities for creation of new Chapters during the coming year, and the National Office, with the approval of the Directors, will concentrate on these possibilities. A Steering Committee has been formed at Seattle, Wash., and prospects for a Chapter this fall are good. Steps are being taken toward establishing A.F.A. groups at other foundry centers, and we still are hopeful for a Chapter in the New England states. A more recent possibility concerns Central Michigan, where interest has already been evidenced.

There is evidence to show that size is not the main criterion for the activity and accomplishments of an A.F.A. Chapter. For example, the Northwestern Pennsylvania Chapter founded in 1944-45 and with a membership of 134 has been most aggressive during the past year. In June they staged a public exhibition to emphasize the importance of the foundry industry and over 5,000 people attended, showing great interest in the displays and events.

During the past year President Walls, in many cases accompanied by the Secretary, visited a large number of Chapters and other visits were made by additional Officers, Directors, and members of the National Office Staff. We recommend for the consideration of the Directors that every A.F.A. Director endeavor to visit at least one Chapter each year, since the burden of official visits by only one or two National Officers has become burdensome and time consuming.

Chapter Chairman Conference. On July 24-25 we staged the Third Annual Chapter Chairman Conference at the Stevens Hotel, this time including more opportunity for discussion and a visit to the Museum of Science & Industry where the Foundry Exhibit, sponsored jointly by A.F.A. and the Chicago Chapter, is outstanding. The 1946 Conference was attended by some 37 Chapter Officers and Directors representing 31 of the 32 active Chapters, Mexico alone not being represented. Five National Officers and Directors and three guests also were present and the dinner meeting July 24 was addressed by Dr. H. T. Heald, President of Illinois Institute of Technology, Chicago.

The Conference is now established as a two-day meeting and for the second consecutive year has been scheduled concurrently with the Annual Board Meetings. We feel that this concurrence may force the Board Meetings to cut short much discussion valuable to Staff thinking and planning, since it would not be practicable to ask that a number of Directors should spend four consecutive days (two for Chapter Conference and two for Board Meetings) away from business at any one time.

The growth of the Association and its finances, however, involve more and more problems and it is believed that two full days should be allotted to Board, Executive Committee and Finance Committee Meetings annually, as a general principle. We recommend, therefore, that consideration be given advancing the date of the 1947 Chapter Chairman Conference so that it may not coincide with the Annual Board Meetings.

Technical Department

During the past year the direction of our technical activities underwent a major change due to the sudden illness of Norman F. Hindle in February. We were fortunate to obtain the services of S. C. Massari, and the advice and background of President Walls on technical matters proved exceptionally valuable as an offset to relative inexperience of the working Staff. From February through the Cleveland Convention, all efforts were concentrated on that event, but since then plans have been made and approved by the Executive Committee for (1) committee reorganization, (2) a 1947 Technical Program, and (3) a definite plan for conducting A.F.A.-sponsored research. It is hoped that further action by the Board will enable us to carry these plans forward promptly, under the leadership of the Technical Director.

During the past year some reorganization of thinking occurred with regard to the Technical Development Program, mainly with regard to its financing. Contributions for the Program were solicited and received only until December 31, 1945, at which time all solicitations ceased with total contributions

as of that date totaling some $92,000. This figure includes the original donations to the former "Foundation." Whereas separate budgets were maintained previously for T.D.P. and the General Activities, the budget for 1945-46 combined all A.F.A. Income and Expense and we are pleased to report that this financial arrangement has proved much more workable.

It should be pointed out that during the past year the T.D.P. fund of securities investments was increased to $100,000.00 and that this fund is kept separate and intact from other reserves.

The work of the Cupola Research Project resulted during the past year in publication of the long awaited CUPOLA HANDBOOK. As the Directors may know, Cupola Research funds are kept separate from all other A.F.A. funds, the Association merely being the administering body. The expense of conducting the Project is charged against Cupola Research and the sale of publications and services performed on behalf of the Project, is returned to the Cupola Research fund for further investigations and work. This fund today stands at $20,000.00 in invested securities and $11,403.85 in net working capital, a total of $31,483.05.

Publications

Special Books. During the past year, in addition to the CUPOLA HANDBOOK, a BIBLIOGRAPHY ON PERMANENT MOLDING was issued and both these publications have been well received. Staff changes and concentration on the Convention made it impossible to complete earlier several additional books which are scheduled for publication during 1946-47.

Forthcoming publications include (1) HISTORY OF THE FOUNDRY INDUSTRY by B. L. Simpson; (2) A SYMPOSIUM ON ENGINEERING PROPERTIES OF GRAY IRON, intended primarily for design engineers; (3) a book on CAUSES OF DEFECTS IN GRAY IRON CASTINGS, and (4) MALLEABLE SYMPOSIUM ON SAND CONTROL.

It should be stated that our policy on the issuance of special publications in the future will be to undertake what we strictly believe can be completed in a reasonable period of time, and to withhold action on new projects until current projects have been accomplished. We now are beginning an examination of TRANSACTIONS volumes to determine the possibilities for a number of symposia on various subjects, and this work is commencing with the topic of Gating and Risering. Conversations also have been held on the advisability of compiling a Dictionary of Foundry Terms, although this work will not be advanced until time permits its being carried through to completion in a reasonable period of time.

TRANSACTIONS. Concentration of all effort on the 1946 Convention has resulted in a delay in publishing the 1945 Bound Volume of TRANSACTIONS. However, the entire book is now in the hands of the printer for publication as early as possible.

AMERICAN FOUNDRYMAN. In publishing the magazine in its present enlarged form we now report that from all indications the first major hurdles have been carried. These hurdles included the obtaining of sufficient good papers, acquiring sufficient staff, obtaining advertising support, printers' strikes, establishment of definite editorial and advertising policies and even more important, building a recognition and acceptance for the monthly magazine in lieu of the former QUARTERLY.

On the subject of advertising, gross revenue for the fiscal year 1945-46 totaled $94,629.25 against a budget of $75,000; net revenue, after commissions, totaled $83,247.31. This indicates that our advertising efforts proved somewhat more successful than anticipated by the budget, although somewhat below the most optimistic gross estimates made a year ago. We realize that some of the advertising received during the past year was placed on a "support for A.F.A." basis. We now are facing our first full year under an established policy whereby advertising is solicited strictly on the basis of "returns for the advertiser."

We are determined to improve the magazine editorially in the belief that additional advertising will come to us as a result. It is our intention to obtain and carry more articles of a practical nature, to give more care to the type of article published and to submit major articles to proper committees for review before acceptance for publication. Several members of the Staff are capable of direct assignment to obtain practical material, one such article appearing in the July issue on *"Precision Castings."*

Convention and Exhibit

Through unavoidable circumstances, A.F.A. undertook its greatest Convention and Exhibit with a relatively untrained and inexperienced Staff, only one-half of whom had ever before worked on a major industrial Convention and Exhibit. To the complete credit of the Staff, the 50th Anniversary Foundry Congress and Show proved to be the largest and most successful event of its kind ever staged by A.F.A., with 273 exhibiting firms occupying 99,433 square feet of space. The previous largest exhibit of A.F.A. was in 1926 when 267 firms utilized 81,000 square feet of space. As might be expected, our Exhibit revenue was in proportion to the size of the event.

On attendance, while no precise means of counting proved possible, our Registered Attendance list includes over 10,000 carefully selected names and it is estimated that total attendance was well over 14,000. Previous estimates of 18,000 are not verifiable but a large number of people attended who were unable to register at all on the first day.

Our Convention and Exhibit expenses, of course, were increased but followed successfully the principle of "spending money to make money." We believe that the foundry industry was proud of the "Golden Jubilee" event which A.F.A. staged, that it was amazed to see what the industry itself could do, and that the result of the event can only be most beneficial in increasing "pride of industry" among foundrymen everywhere.

Especially gratifying was the presence of seven trade associations at the Convention, each with an exhibit booth and each holding meetings during the week. These associations were American Society For Metals, Foundry Equipment Manufacturers' Assn., Gray Iron Founders' Society, Magnesium Association, Malleable Founders' Society, National Founders' Society, and Non-Ferrous Founders' Society. We believe this evidence of cooperation will prove mutually beneficial.

Cooperation with Armed Forces

Possibly as an aftermath of industry cooperation on the war effort, A.F.A. has been contacted by several of the Armed Services during the past two years and every advantage is being taken of the opportunities afforded. First contact was by the Ordnance Department during 1944-45, as a result of which an Ordnance Committee of the Board was formed. Since then further interest has been shown by the Army Industrial College and by the Navy Department for similar and closer cooperation. We therefore recommend that consideration be given the formation of an overall Committee of A.F.A. to consult with any branch of the Armed Services, a single Committee being able to coordinate all such activities to good advantage.

Staff Organization

The staff problems of the past year, with which most members of the Board are well acquainted, have been largely a natural outcome of situations created by a major organizational change and the consequent difficulties of readjustment. They are a natural outcome, too, of what, during the past ten years has been an ever-increasing need for a larger staff and one capable of handling the growing demands of an increasing membership and an industry becoming ever more conscious of its strength and possibilities.

Following the presentation and discussion of a staff organizational plan with the Executive Committee last November, the staff was increased to 29 members (including Ed Hoyt and Norm Hindle) but by February, 1946, this plan was completely out of date. It again must be revised and submitted to the Executive Committee. The problem is to strike a balance between a Staff large enough to handle unusual burdens (e.g., Conventions and Exhibits) without excessive pressure, and at the same time one small enough to handle normal activities at minimum expense.

The Secretary was instructed to prepare and submit for Board approval a Retirement Plan for the National Office staff. The matter has been the subject of preliminary discussions and will be advanced further during the year ahead.

Educational Activities

During the past year a detailed National Educational Program was worked out by Norm Hindle as Director of T.D.P.,

with H. F. Scobie, and approved by the Board. It can now be reported that the plan has been well received and has stimulated a great deal of thinking by the Chapters along educational lines. Essentially a long-range Program, this educational activity is being carried out continuously, now under the direction of S. C. Massari, and consistent results may be expected.

It is probable that one result of a National Educational Program may be the formation of an Educational Division of A.F.A., on a parity with the other six present A.F.A. Divisions. Under such a Division would fall all Committees dealing with Apprentice Training, Foreman Training, Youth Encouragement and cooperation with Engineering Schools and Colleges.

Safety and Hygiene Work

Evidence of increased interest on the part of foundrymen in Safety and Hygiene work having been shown during the past year, a temporary committee to investigate the subject was appointed at the January Board Meetings under the Chairmanship of Director Wilson. As a result of discussions by this committee, a plan is contemplated for a strong A.F.A.-Sponsored Safety and Hygiene Program. We feel that such a Program should eventually be very favorably received by foundry management and that there is real need for A.F.A. to carry on its Safety and Hygiene work following the excellent results achieved by the A.F.A. Committees formerly under the Chairmanship of James R. Allan.

Research Projects

During the past year A.F.A. has carried on three Research Projects of fundamental nature. One is the investigation of the effects of elevated temperatures on properties of foundry sands, conducted at Cornell University by an A.F.A. Fellow. Second is investigation of heat transfer effects being carried on at Columbia University by Dr. Paschkis. A third, now commencing, involves additional work on heat transfer authorized at Battelle Memorial Institute.

Respectfully submitted,

WM. W. MALONEY, *Secretary.*

Report on Association Finances

For Fiscal Year Ended June 30, 1946

THE FISCAL YEAR 1945-46, the Association's 50th Anniversary year, proved to be perhaps the most successful financially in the history of the Association, as may be seen from the attached exhibits: (a) Comparative Condensed Balance Sheet for three consecutive years; (b) Statement of Income and Expense for the major activities of A.F.A., with all overhead and expense items distributed among the Association's several major activities.

Three major revenue sources accounted for the bulk of the Association's income during the past year, as is shown by the latter statement. The considerable increase in Sustaining and Company Member dues, effective July 1, 1945, resulted in some loss in these two member classes, but a very considerable increase in total dues revenue, as was anticipated.

Acceptance of advertising in monthly issues of AMERICAN FOUNDRYMAN, effective May 1945 for the first time, developed a substantial new source of income and, in spite of predictions, the magazine was operated at a profit during its first year in expanded form. A third large source of revenue was the Golden Anniversary Convention and Exhibit staged in Cleveland, and featuring the largest Foundry Show thus far staged by A.F.A.

The accounts of the Association were audited on June 30, 1946, as required by the By-Laws, by the firm of Lybrand, Ross Bros. & Montgomery of Chicago, and the accompanying financial exhibits for that date are taken from their audit. In addition to liquid funds for current operations, the Association maintains the following special funds:

(1) Awards Funds, administered by the Board of Awards, only the interest on the principal being utilized for purposes of periodic awards for outstanding achievements by men of the industry.
(2) General Reserve Fund, added to periodically and held for future contingencies,
(3) Technical Development Fund, at present pegged at approximately $100,000 and maintained in order to assure continuity of the A.F.A. technical program in spite of contingencies.
(4) Cupola Research Fund, utilized only for Cupola Research activities and administered by the Association for that purpose.

All financial operations of the Association are governed by annual budgets set up by a Finance Committee and approved by the Board of Directors. Since A.F.A. is a non-profit incorporated technical society, surplus funds are allocated to the several maintained funds or utilized for operations of the following year. Investments are controlled by the Finance Committee.

During the past year, two steps were taken to improve the Association's accounting practices. The firm of Lybrand, Ross Bros. & Montgomery was retained to study our accounting methods, and most of the recommendations therefrom have been put into effect. In addition, as recommended by the Treasurer a year ago, the practice of maintaining separate budgets and accounts for the Technical Development Program and the Association's General Activities was abandoned, greatly simplifying our accounting.

Major items of Income during the year are shown in the accompanying exhibit. Major items of Expense included: All Staff salaries and compensations, $84,014.54, or less than 20 per cent of total income; AMERICAN FOUNDRYMAN, $87,792.89; Research activities, $10,600.00; Chapter expense, $31,996.41; Special Publications, $12,033.58; 1946 Convention expense, $56,786.06.

The accompanying Condensed Balance Sheet for the past three years illustrates the manner in which the Association has conservatively added to its several reserve and special funds while returning the major portion of its increased income into useful activities as required to serve a steadily expanding membership. Thus Cash On Hand was decreased in order to augment invested funds by some $120,000 and Inventories (mainly of publications) by some $21,000.

The General Reserve Fund, as distinguished from other special funds maintained, was increased solely by accrued interest on the principal, illustrating the Association's policy as a technical society of concentrating on its "working" funds.

The conclusion can be drawn that the Association's finances were in excellent, sound condition at the close of the fiscal year 1945-46, due to the steadily increasing interest in Association activities displayed by the Castings Industry as a whole.

Respectfully submitted,

WM. W. MALONEY, *Treasurer.*

Comparative Condensed Balance Sheet

As of June 30, 1944, 1945 and 1946

ASSETS	*June 30, 1944*	*June 30, 1945*	*June 30, 1946*	*LIABILITIES*	*June 30, 1944*	*June 30, 1945*	*June 30, 1946*
Cash	$140,827.93	$ 86,920.86	$131,341.19	Accounts Payable	$ —	$ 374.40	$ 18,197.03
Securities	61,993.90	131,993.90	181,949.60	Reserves	48,767.78	49,939.99	51,069.06
Accounts Receivable	3,011.14	2,404.59	9,379.56	Accrued Items	359.07	—	2,013.17
Inventories	12,375.24	21,364.52	33,182.95	Deferred Income	9,406.42	8,893.70	12,991.53
Prepaid Expenses	7,710.97	206.73	1,780.21	Unexpended Special Funds	99,514.77	141,705.37	164,699.80
Furniture and Fixtures	1,024.50	2,017.93	3,765.59	Surplus	68,895.64	43,995.07	112,428.51
TOTAL	$226,943.68	$244,908.53	$361,399.10	TOTAL	$226,943.68	$244,908.53	$361,399.10

Condensed Statement of Income and Expense

(All Expenses Distributed to Major Activities)

Fiscal Year July 1, 1945—June 30, 1946

INCOME

Membership Dues	$148,968.86
Publications	17,187.28
AMERICAN FOUNDRYMAN	94,278.35
Convention and Exhibit	154,554.58
Other Income	2,237.25
TOTAL	$417,226.32

EXPENSE

Major Activities	*Total Expense Amount*	*Per Cent*
General Administration	$ 79,006.73	23.61
Chapter Activities	47,529.86	14.21
Technical Activities and Publications	69,086.00	20.65
AMERICAN FOUNDRYMAN	66,527.66	19.89
Convention and Exhibit	72,444.02	21.65
TOTAL	$334,594.27	100.00

Comparative Membership Report

As of June 30, 1942-1946, Inclusive

	June 30, 1942	*June 30, 1943*	*June 30, 1944*	*June 30, 1945*	*June 30, 1946*
Sustaining Members	104	113	169	239	183
Company Members	1,182	1,341	1,471	1,571	1,462
Personal Members	3,467	3,958	4,720	5,722	6,426
Student and Apprentice Members	82	79	42	23	20
Honorary Life Members	53	54	55	55	58
Foreign Members	159	120	163	205	390
Total	5,047	5,665	6,620	7,815	8,539
New Members, 12 Months	1,128	1,172	1,510	1,803	1,743
Resignations	115	150	215	235	370
Delinquents Dropped	123	384	316	356	631
Removed by Death	11	20	24	17	18
Net Gain for Year	879	618	955	1,195	724
Members in Chapters	4,283	5,003	5,902	7,110	7,767

Chapter Distribution

June 30, 1946

Chapters	*Sustaining*	*Company*	*Personal*	*Affiliate*	*Associate*	*Apprentice and Student*	*Honorary and Life Membership*	*Chapter Membership*
Birmingham	3	41	39	167	4	0	2	256
Canton	0	27	12	73	0	0	0	112
Central Illinois	1	8	13	116	0	3	0	141
Central Indiana	4	28	31	137	4	1	0	205
Central New York	1	31	37	72	5	0	1	147
Central Ohio	1	34	23	126	5	0	2	191
Chesapeake	4	29	37	90	43	2	3	208
Chicago	26	109	111	488	22	9	10	775
Cincinnati	6	53	47	129	3	1	1	240
Detroit	13	64	65	239	17	3	5	406
E. Canada and Newfoundland	6	54	61	159	11	0	0	291
Metropolitan	12	60	78	197	15	0	2	364
Mexico City	0	13	25	12	0	0	0	50
Michiana	3	34	32	117	2	0	0	188
Northeastern Ohio	20	95	81	282	10	0	8	496
Northern California	6	29	68	147	12	0	0	262
No. Illinois-So. Wisconsin	3	18	7	45	1	0	1	75
Northwestern Pennsylvania	0	23	20	89	2	0	0	134
Ontario	2	96	92	96	6	0	1	293
Oregon	2	13	13	46	3	0	0	77
Philadelphia	16	61	58	170	11	0	5	321
Quad City	5	39	48	120	6	1	3	222
Rochester	1	14	16	73	2	0	0	106
Saginaw Valley	6	12	9	192	3	3	1	226
St. Louis	3	49	48	129	4	1	0	234
Southern California	4	76	116	144	12	0	1	353
Texas	5	30	28	52	4	0	1	120
Toledo	0	22	23	55	1	0	1	102
Twin City	3	41	34	99	12	2	0	191
Western Michigan	5	31	15	150	2	1	2	206
Western New York	3	37	44	136	3	0	1	224
Wisconsin	16	87	69	368	8	2	1	551
Total in Chapters	180	1,358	1,400	4,515	233	29	52	7,767
Total Foreign								392
Total Outside of Chapters								380
GRAND TOTAL								8,539

Third Annual Chapter Chairman Conference

July 24-25, 1946—Stevens Hotel, Chicago

PROGRAM

Wednesday, July 24

10:00 AM MORNING PROGRAM.
Presiding Chairman, Max Kuniansky. *INTRODUCTORY REMARKS* on purpose of Chapter Chairman Conference.

Welcome to Chapter Officers....President-Elect S. V. Wood

'Round-the-Table Introductions .....Chairman Max Kuniansky

Know Your A.F.A.....President Fred Walls
A brief outline of A.F.A.—its origin, aims and development over a half century, and its place in the Foundry Industry today.

A National Program for A.F.A. Chapters.....Secretary Wm. W. Maloney

Building Better Chapter Programs
OPEN DISCUSSION by Chapter Chairmen, to assist in arranging programs *to best serve the membership.*

12:00 NOON MORNING PROGRAM ADJOURNED

12:15 PM LUNCHEON

1:05 PM ASSEMBLY ON MICHIGAN AVE.

1:15 PM BUSES LEAVE HOTEL FOR MUSEUM

2:00 PM TOUR OF MUSEUM OF SCIENCE AND INDUSTRY

4:00 PM AFTERNOON PROGRAM
Presiding Chairman, Max Kuniansky.
Educational Activities of A.F.A. Youth Encouragement
In-Plant Training and Vocational Schools
Educational Courses
Cooperation With Engineering Schools

5:20 PM AFTERNOON PROGRAM ADJOURNED

7:30 PM CHAPTER CHAIRMEN DINNER
Presiding—President Fred Walls.
Toastmaster—President-Elect S. V. Wood.
Guest Speaker—Dr. Henry T. Heald, President, Illinois Institute of Technology, Chicago.
Subject—"Engineering Education and the Foundry Industry."

Thursday, July 25

10:00 AM MORNING PROGRAM
Presiding Chairman, Max Kuniansky.

Publishing Activities of A.F.A.....Secretary Wm. W. Maloney
Intended to acquaint Chapter Officers with the publishing policies of A.F.A. and to bring them up to date on current and forthcoming publications. Special publications. The annual TRANSACTIONS. The AMERICAN FOUNDRYMAN. The A.F.A. Library.

Technical Activities of A.F.A.....Technical Director S. C. Massari
An outline of the technical committee organization of A.F.A. The Technical Development Program. The 1947 Convention program. A new publication policy of A.F.A. Research projects, and a new research program. How Chapter Chairmen can play a larger part.

National and Chapter Financing
First—Some pertinent information on financing the National organization. Major sources of funds. Budgets and the Finance Committee. Distribution of the A.F.A. Income Dollar. Reserve funds. Financial statements.
Second—General *DISCUSSION* on Chapter financing. The Chapter refund from National Office. Investment of Chapter funds. Sources of Chapter income. Chapter financial reports to National Office.

Highlights of Chapter Operation
Chapter by-laws. Chapter publications. Chapter board meetings and frequency. Nomination and election procedures—members only. Territorial considerations—National policy on formation of new Chapters. Activities not within the province of A.F.A. Chapters. Photographs of Chapter officers and Directors.

12:45 PM LUNCHEON

2:00 PM AFTERNOON PROGRAM
Chairman Max Kuniansky, presiding.

Membership Committee Activities.....Vice-President S. V. Wood
OPEN DISCUSSION by Chapter Chairmen on membership considerations, to assist in increasing present membership.

4:30 PM FINAL ADJOURNMENT

ATTENDANCE

July 24-25, 1946—Stevens Hotel, Chicago

National President, F. J. Walls, Metallurgist, International Nickel Co., Detroit.

National Vice-President, S. V. Wood, President and General Manager, Minneapolis Electric Steel Castings Co., Minneapolis.

National Director and Chairman of Chapter Chairman Conference, Max Kuniansky, Vice-President and General Manager, Lynchburg Foundry Co., Lynchburg, Va.

National Director, E. W. Horlebein, President, Gibson & Kirk Co., Baltimore, Mr.

Guest Speaker, Henry T. Heald, President, Illinois Institute of Technology, Chicago.

Guest, George E. Rauen, Secretary, Olympic Foundry Co., Seattle, Wash.

Guest, Frank Steinebach, Editor *The Foundry,* Cleveland.

Chapter Officers

BIRMINGHAM—*Chairman,* T. H. Benners, Jr., Managing Partner, T. H. Benners & Co., Birmingham.
**Vice-Chairman,* W. E. Jones, Chief Engineer, Stockham Pipe Fittings Co., Birmingham.

CANTON—*Chairman,* I. M. Emery, Works Manager, Massillon Steel Castings Co., Massillon, Ohio.

CENTRAL ILLINOIS—*Chairman,* Z. Madacey, Foundry Superintendent, Caterpillar Tractor Co., Peoria, Ill.
**Sec.-Treas.,* Carl W. Wade, Training Supervisor, Caterpillar Tractor Co., Peoria, Ill.

CENTRAL INDIANA—*Chairman,* J. P. Lentz, Foundry Metallurgist, International Harvester Co., Indianapolis.

CENTRAL NEW YORK—*Chairman,* E. E. Hook, Service Engineer, Dayton Oil Co., Syracuse, N. Y.

CENTRAL OHIO—*Chairman,* N. J. Dunbeck, Vice-President, Eastern Clay Products, Inc., Eifort, Ohio.
**Vice-Chairman,* R. H. Frank, Chief Metallurgist, Bonney-Floyd Co., Columbus, Ohio.

CHESAPEAKE—*Chairman,* David Tamor, Plant Metallurgist, Malleable Foundry Div., American Chain & Cable Co., York, Pa.

CHICAGO—*Chairman,* L. H. Hahn, Metallurgist, Sivyer Steel Casting Co., Chicago.

CINCINNATI—*Chairman,* Joseph S. Schumacher, Chief Engineer, Hill & Griffith Co., Cincinnati.

*Attending on Chapter's request.

DETROIT—*Chairman,* A. H. Allen, Detroit Editor, Penton Publishing Co., Detroit.

CANADA AND NEWFOUNDLAND—*Vice-Chairman,* A. E. Cartwright, Metallurgist, Robert Mitchell Co., Ltd., Montreal. **Secretary,* Robert E. Cameron, Sec.-Treas., Webster & Sons, Ltd., Montreal.

METROPOLITAN—*Chairman,* Harold L. Ullrich, Manager, Sacks-Barlow Foundries, Inc., Newark, N. J.

MICHIANA—*Chairman,* John McAntee, Foundry Manager, Covel Mfg. Co., Benton Harbor, Mich.

NORTHEASTERN OHIO—President, Henry J. Trenkamp, President, The Ohio Foundry Co., Cleveland.

NORTHERN CALIFORNIA—*Chairman,* Richard Vosbrink, Owner, Berkeley Pattern Works, Berkeley, Calif.

N. ILLINOIS & S. WISCONSIN—*Chairman,* John Doerfner, Jr., Works Manager, Gunite Foundries Corp., Rockford, Ill. **Vice-Chairman,* John Clausen, Metallurgist, Greenlee Bros. & Co., Rockford, Ill. **Treasurer,* Lester Fill, Foundry Superintendent, George D. Roper Corp., Rockford, Ill.

NORTHWESTERN PENNSYLVANIA—*Chairman,* Earl M. Strick, Finishing Superintendent, Erie Malleable Iron Co., Erie, Pa.

ONTARIO—*Chairman*—J. A. Wotherspoon, General Manager, Imperial Iron Corp., St. Catharines, Ontario.

OREGON—F. A. Stephenson, Partner, Dependable Pattern Works, Portland, Oregon.

PHILADELPHIA—*Chairman,* B. A. Miller, Chief Metallurgist, Cramp Brass & Iron Foundries, Div., Baldwin Locomotive Works, Eddystone, Pa.

QUAD CITY—*Chairman,* Claude S. Humphrey, Owner, C. S. Humphrey Co., Moline, Ill.

ROCHESTER—*President,* Walter G. Brayer, Foundry Superintendent, Bausch & Lomb Optical Co., Rochester, N. Y.

SAGINAW VALLEY—*Chairman,* John F. Smith, Night Superintendent, Chevrolet Gray Iron Foundry, Saginaw, Mich.

ST. LOUIS DISTRICT—*Chairman,* Roland T. Leisk, Asst. Works Manager, American Steel Foundries, E. St. Louis, Ill.

SOUTHERN CALIFORNIA—*President,* Wm. D. Emmett, Foreman, Los Angeles Steel Casting Co., Los Angeles.

TEXAS—*Vice Chairman,* L. H. August, Foundry Engineer, Hughes Tool Co., Houston, Texas.

TOLEDO—*Chairman,* Brock L. Pickett, Chief Inspector, Unitcast Corp., Toledo, Ohio.

TWIN CITY—*Chairman,* H. M. Patton, Foundry Superintendent, American Hoist & Derrick Co., St. Paul, Minn.

WESTERN MICHIGAN—*Chairman,* Rudolph Flora, Metallurgist, Clover Foundry Co., Muskegon, Mich.

WESTERN NEW YORK—*Chairman,* Henry C. Winte, Metallurgist, Worthington Pump & Machinery Corp., Buffalo, N. Y.

WISCONSIN—*President,* David C. Zuege, Technical Director, Sivyer Steel Casting Co., Milwaukee.

Minutes of the Executive Committee Meeting

Palmer House, Chicago—July 19, 1945

Presiding: President F. J. Walls
Present: President F. J. Walls
Vice-President S. V. Wood
Directors: R. J. Teetor
W. B. Wallis
L. C. Wilson
Secretary R. E. Kennedy
Treasurer C. E. Hoyt
Business Manager W. W. Maloney
Director T.D.P. N. F. Hindle
Assistant Treasurer Jennie Reininga

1. *Staff Considerations.* On recommendations of the Finance Committee, R. J. Teetor, Chairman, the compensation of elective staff officers was approved for the year 1945-46.

On motion by Director Wallis, seconded by Director Wilson, the Secretary was directed to secure specific Finance Committee approval for the employment of Staff assistants at salaries in excess of $3600.00 per year.

2. *Chapter Contacts.* The need for a Chapter Contact representative was stressed by members of the Executive Committee and the desirability of employing a man with executive ability, sales personality, some knowledge of A.F.A. and foundry industry, and ability to approach top foundry management.

On motion by Vice-President Wood, seconded by Director Teetor, the Secretary was instructed to seek such a man for possible employment as a member of the Staff.

3. *T.D.P. Research Activities.* N. F. Hindle, as Director of the Technical Development Program, reported that the T.D.P. contract with Cornell University on Foundry Sand Research had expired July 1, 1945, and that renewal was necessary. He also reported that the T.D.P. contract with Columbia University on work of the Heat Transfer Committee had expired July 1, 1945, and that an interim allowance was necessary for the work to continue. He also indicated that $5500.00 would be required for Sand Research during 1945-46, including $1500.00 for purchase of equipment.

On motion by Director Wallis, seconded by Director Wilson, the Executive Committee approved the additional expenditure of $2000.00 with Cornell University and $2000.00 with Columbia University, for continuation of investigational work conducted at those two schools.

4. *Cupola Research.* The Executive Committee expressed the opinion that approval for expenditures on the *Cupola Handbook* should first be rendered by the Cupola Research Steering Committee, subject to approval by the Executive Committee of the A.F.A. Board.

5. *T.D.P. Solicitation.* It was pointed out that the cost of soliciting funds for the Technical Development Program had thus far been paid out of general accounts. The Executive Committee expressed the opinion that the General Fund should be reimbursed out of T.D.P. earnings so that eventually T.D.P. might bear its own full expense.

6. There being no further discussion, the meeting was adjourned.

Respectfully submitted,
WM. W. MALONEY, *Secretary.*

Minutes 1945-46 Board of Directors

Palmer House, Chicago—July 19, 1945

Presiding: President F. J. Walls
Present: President F. J. Walls
Vice-President S. V. Wood
Directors: D. P. Forbes, R. M. Jacobs, Max Kuniansky, R. J. Teetor, W. B. Wallis, S. D. Russell, R. T. Rycroft, J. Sully, L. C. Wilson, G. K. Dreher, E. W. Horlebein, H. H. Judson, J. H. Smith, F. M. Wittlinger
Secretary R. E. Kennedy
Treasurer C. E. Hoyt
Director T.D.P. N. F. Hindle
Business Manager W. W. Maloney
Assistant Treasurer J. Reininga

1. Retiring President R. J. Teetor called the meeting to order and immediately turned the Chairman's gavel over to incoming President F. J. Walls, officially declared elected at the Annual Business Meeting held the previous day.

2. *Executive Committee Appointed.* A Board Nominating Committee being appointed by the President, consisting of Directors D. P. Forbes, S. D. Russell and E. W. Horlebein, the Committee was instructed to retire and submit nominations for the 1945-46 Executive Committee of the Board and for the election of Staff officers.

For members of the Executive Committee, the Nominating Committee proposed Directors R. J. Teetor, W. B. Wallis, L. C. Wilson and J. H. Smith, in addition to President F. J. Walls as Chairman, and Vice-President S. V. Wood.

Director Max Kuniansky moved formation of the 1945-46 Executive Committee in accordance with the Nominating Committee's report. Director Rycroft seconded the motion, unanimously carried.

3. *Standing and Special Committees.* Director Horlebein moved adoption of a resolution authorizing the President to appoint Standing and Special Committees of the Board. Director Dreher seconded, unanimously carried. President Walls then announced appointments to Standing and Special Committees of the Board as follows:

Executive Committee:
F. J. Walls, Chairman
S. V. Wood
J. H. Smith
R. J. Teetor
W. B. Wallis
L. C. Smith

Finance Committee:
R. J. Teetor, Chairman
F. J. Walls
S. V. Wood

Chapter Contacts Committee:
S. V. Wood, Chairman
J. Sully
R. T. Rycroft

G. K. Dreher
S. D. Russell
F. M. Wittlinger
Harry Reitinger

Management Committee:
R. T. Rycroft, Chairman
Joseph Sully
J. H. Smith
R. M. Jacobs
E. W. Horlebein
Max Kuniansky

Membership Committee:
S. V. Wood, Chairman

AMERICAN FOUNDRYMAN *Committee:*
W. B. Wallis, Chairman
H. H. Judson
Max Kuniansky
Harry Reitinger
D. P. Forbes

Youth Encouragement Committee:
Joseph Sully, Chairman
J. H. Smith
G. K. Dreher
F. J. Dost
S. D. Russell

4. *By-Laws Committee.* Treasurer Hoyt suggested appointment of a committee on By-Laws as a standing committee of the Board. In discussion, the Directors emphasized that such a committee should not be confined entirely to the directorate. It was moved by Director Kuniansky, seconded by Chairman Wilson and carried, that the President be empowered to appoint a standing By-Laws Committee at his discretion.

5. *Recommendations of Retiring President.* Recommendations were presented by retiring President Teetor and discussed in detail, resulting in the following actions on specific points concerning A.F.A. procedures:

(a) It was recommended, and approved by the Board, that Executive Committee meetings be held bi-monthly, in addition to the customary Board meetings in January, July and at convention time.

(b) It was recommended, and approved by the Board, that the Secretary prepare at the earliest possible moment, an Organizational Plan dealing with Staff functions, subject to approval by the Executive Committee.

(c) It was recommended, and left to the discretion of the Secretary, that frequent conferences be held within the National Office dealing with matters of policy, advertising, publications, and committee work.

(d) It was recommended, and approved by the Board, that the Secretary should formulate at the earliest possible date a Retirement Plan for the National Office Staff, such plan to be subject to approval by the Executive Committee and Board. The thought was expressed that such a plan should be financed by the Association.

(e) It was recommended, and approved by the Board, that a salary limit of $3,600.00 per year govern the employment of Staff assistants by the Secretary, and that Finance Committee approval be obtained for the employment of assistants at salaries beyond that figure.

(f) It was recommended that a campaign be instituted for a considerably larger Technical Development Program fund. After considerable discussion on ways and means, it was informally agreed that T.D.P. solicitation should be discontinued at the end of 1945 in favor of concentration on Sustaining Memberships.

(g) It was recommended that greater thoughts be given to ways of inducing the Foundry Trade Associations to play a more active part in the affairs of A.F.A.

(h) It was recommended, and approved by the Board, that greater consideration be given a definite policy on the formation of new Chapters or Chapter groups, and that consideration be given to reducing the minimum support required in certain areas.

(i) It was recommended, and approved by the Board, that greater effort be made toward obtaining the membership of many non-member foundries so as to build up both the prestige of A.F.A. and the value of AMERICAN FOUNDRYMAN as an advertising medium. In this connection, it was recommended that A.F.A. should concentrate on programs already initiated so as to produce results attractive to a larger membership.

(j) It was recommended, and approved by the Board, that all contributions of funds to T.D.P. should be used for purposes intended beyond any reasonable question.

6. *Bank Resolutions.* The usual resolutions were presented covering the requirements of A.F.A. depository banks as to withdrawal of funds. On motion by Vice-President Wood, seconded by Director Wilson, the bank resolutions were approved en toto.

7. *T.D.P. Projects.* Vice-President Wood, as Chairman of the T.D.P. Advisory Committee, pointed out the need for Board clarification on the type of investigational projects which might properly be undertaken by T.D.P. as established. While there was agreement that T.D.P. can undertake any research project of general usefulness to all branches of the industry, a more specific policy was felt desirable on financing projects of more specific usefulness to certain foundry groups. Mr. Hindle stated that the matter would be thoroughly discussed at the next meeting of the T.D.P. Advisory Committee, for preparation of specific recommendations.

8. *Chapter Chairmen Conference.* In view of the success of the past two Chapter Chairmen Conferences, Director Wilson moved approval of a third Annual Conference during the year 1945-46. Director Jacobs seconded, unanimously carried.

9. *Staff Reorganization.* President Walls called attention to action of the old Board, recommending certain Staff reorganizations in accordance with findings arrived at by the Executive Committee in session April 26-27, 1945, as later modified and approved by the old Board. It was pointed out that progressive action on these recommendations was necessary in order for them to be made effective immediately, as drawn up and for the year 1945-46.

Accordingly, the following resolutions were offered:

RESOLUTION

WHEREAS, the sincere expressions of sentiment by the Executive Committee, which expressions are by reference incorporated into these Minutes, are hereby heartily endorsed by the Board of Directors, and

WHEREAS, the Board of Directors nevertheless fully realizes that it could not possibly give to Treasurer C. E. Hoyt and Secretary R. E. Kennedy adequate recognition for what these two revered and loved executives have done for the Association, and

WHEREAS, the Board of Directors is in unanimous agreement that the willingness of Treasurer C. E. Hoyt to continuously respond to needs of the Association as then developed should no longer be permitted to interfere with his expressed intentions of retirement, and

WHEREAS, the Board of Directors is in unanimous agreement that the inspiration of Secretary Robert E. Kennedy, by reason of his long and unselfish service to the Association as Technical Secretary and Secretary, should be preserved for the longest period possible,

NOW THEREFORE BE IT RESOLVED that in keeping with the purposes expressed by the Executive Committee, the retirement of C. E. Hoyt as Treasurer of the Association is hereby recognized as of August 1, 1946, and

RESOLVED that in keeping with the purposes expressed by the Executive Committee, the retirement of Robert E. Kennedy as Secretary of the Association is hereby recognized as of August 1, 1945, and

RESOLVED that further in keeping with the purposes expressed by the Executive Committee, Robert E. Kennedy is hereby accorded the title of Secretary Emeritus of the American Foundrymen's Association for the term of his life, with the understanding that such services as he may hereafter per-

form for the Association shall be at the option and direction of the then Secretary, to be rendered in accordance with Mr. Kennedy's capability and convenience, and

RESOLVED that the compensation of Treasurer C. E. Hoyt after August 1, 1946, and of Secretary Emeritus Robert E. Kennedy, after August 1, 1945, shall be as determined by existing Boards of Directors of the Association, and

BE IT FURTHER RESOLVED that this Board of Directors unanimously and strongly urges that the conclusions as set forth in these resolutions shall be considered as irrevocable, and binding upon all future Boards of Directors of the American Foundrymen's Association.

On motion by Director Wilson, seconded by Director Wagner, the above resolutions were unanimously adopted.

Staff Appointments. Following these actions, the Nominating Committee presented recommendations on Staff appointments for 1945-46, as follows:

Secretary—W. W. Maloney.

Director of Technical Development Program—N. F. Hindle.

Treasurer—C. E. Hoyt.

Assistant Treasurer—Jennie Reininga.

On motion by Director Wallis, seconded by Director Wilson, Staff appointments were approved as recommended, and the Executive Committee authorized to determine proper compensations in accordance with recommendations of the Finance Committee.

10. With the unanimous agreement of the Board members present, Director Rycroft paid tribute to the excellent administration of retired President Teetor during the past year.

11. There being no further business, Chairman Walls adjourned the meeting at 1:20 P.M.

Respectfully submitted,
WM. W. MALONEY, *Secretary.*

Minutes of the Finance Committee Meeting

Palmer House, Chicago—September 6, 1945

Presiding: Past President R. J. Teetor, *Chairman*
Present: President F. J. Walls
Vice-President S. V. Wood
Directors: L. C. Wilson
W. B. Wallis
Secretary Wm. W. Maloney
Secretary Emeritus R. E. Kennedy
Director of T.D.P. N. F. Hindle
Treasurer C. E. Hoyt
Assistant Treasurer J. Reininga

1. *Staff Considerations.* The Secretary proposed the employment of H. F. Scobie, instructor in Foundry Practice at the University of Minnesota, as a member of the Headquarters Staff. It was stated that Mr. Scobie's services should be of great value to the Association, especially along the lines of educational work, and Finance Committee approval was sought for his employment.

On motion by Vice-President Wood, seconded by President Walls, the Finance Committee authorized the Secretary to carry through final arrangements to employ Mr. Scobie.

2. *1945-46 Budget.* In considering an Income and Expense Budget for 1945-46, discussion arose as to whether separate budgets should again be prepared, one for the General activities and one for the Technical Development Program. It was pointed out that A.F.A. has a definite obligation to utilize funds contributed by the industry to the Technical Development Program, solely for purposes of the Program. The unwieldiness of two separate budgets and individual accounts for General and T.D.P. activities was brought out and the suggestion made that a single budget would better serve the purposes of unified and efficient operation.

It was agreed that a single budget should be prepared in such a manner as to make plain the dissemination of contributed T.D.P. funds at any time, with the recommendation that a substantial operating fund for T.D.P. might well be set up and maintained continuously so as not to handicap the program in its essential work for A.F.A. and the industry.

Respectfully submitted,
WM. W. MALONEY, *Secretary.*

Minutes of Executive Committee Meeting

Palmer House, Chicago—September 7, 1945

Presiding: President F. J. Walls
Present: President F. J. Walls
Vice-President S. V. Wood
Directors: R. J. Teetor
W. B. Wallis
L. C. Wilson
Secretary Wm. W. Maloney
Secretary Emeritus R. E. Kennedy
Treasurer C. E. Hoyt
Director of T.D.P. N. F. Hindle
Assistant Treasurer J. Reininga

1. Minutes of meetings of the Executive Committee and Board of Directors, held in Chicago on July 19, were read and approved.

2. *Nominating Committee.* In accordance with the procedure prescribed by the By-Laws, names of candidates submitted for the 1946 Nominating Committee were considered. On motion by Director Teetor, seconded by Vice-President Wood, the following personnel of the 1946 Nominating Committee was appointed:

Past Pres. L. C. Wilson, *Chairman*	H. F. Taylor
Past Pres. R. J. Teetor	N. E. Woldman
J. A. Woody	F. W. Shipley
I. M. Emery	E. E. Ballard
A. M. Fulton	

3. *1945-46 Income and Expense Budget.* An Income and Expense Budget for the fiscal year ending June 30, 1946, combining General and T.D.P. activities, was submitted as recommended by the Finance Committee. On motion by Director Wilson, seconded by Director Wallis, the Budget was approved.

4. *Membership Report.* The Assistant Treasurer reported on dues collections as of August 31, following the increase in Company and Sustaining dues effective July 1st, showing cash receipts of $113,974.00 for 4,969 members. The report also showed the following figures:

Book Membership—June 30, 1945	7,815
New Members to August 31	263
Total	8,078
Resignations	136
Total, August 31, 1945	7,942

5. *Foreign Chapters.* Attention was called to the imminent desire for formation of a chapter in Mexico and the Executive Committee was asked to set a policy with regard to acceptance of a chapter in Mexico and other foreign countries.

Director Wallis moved that the formation of chapters in countries of the Western Hemisphere should be encouraged on the basis of dues effective July 1, 1945, but allowing time for notification of the new dues to reach present members. Motion seconded by Director Wilson and carried.

The Secretary was instructed to give full support to formation of a Mexican chapter on the same basis as would a new Canadian chapter be considered.

6. *Simpson Endowment Fund.* Director Teetor called attention to the offer of Past President H. S. Simpson for establishment of an A.F.A. Endowment Fund in some form as a memorial to the late Peter L. Simpson. Following discussion, Director Teetor offered the following motion:

> THAT A.F.A. accept with gratitude the generous offer of H. S. Simpson to endow an A.F.A. Memorial Fund in memory of the late Peter L. Simpson, and that a special committee of the Board be appointed to confer with the donor at the earliest possible date, to the end that the greatest good be accomplished for the foundry industry in honor of one of its outstanding pioneers.

Motion seconded by Vice-President Wood and unanimously carried.

The Executive Committee then approved appointment of a special committee to confer with H. S. Simpson, consisting of R. J. Teetor, Chairman, Fred J. Walls and S. V. Wood, the Committee to report back to the Executive Committee at its next meeting so that progressive action may be taken.

7. AMERICAN FOUNDRYMAN *Advertising.* A question of extending the acceptance of advertising in AMERICAN FOUNDRYMAN beyond 1945 was considered. It was recalled that, at the July Board Meeting, Past President Teetor had been instructed to communicate to F.E.M.A. and F.S.M.A. the decision of the Board not to rescind its previous action on advertising matters. The Executive Committee had been authorized to decide whether or not advertising should be extended beyond 1945, after consideration of reactions to Director Teetor's letter.

Accordingly, following discussion, Vice-President Wood moved

> THAT A.F.A. continue to accept advertising in AMERICAN FOUNDRYMAN and that the staff be directed to proceed on that basis.

Motion seconded by Director Wallis and carried.

The Secretary was directed to advise F.E.M.A. and F.S.M.A. of the decision of the Executive Committee, acting for the Board of Directors, after careful analysis of the reactions of advertisers and readers prior to and since the 1945 Annual Meeting of the Board of Directors.

8. *By-Laws Committee.* It was pointed out that the A.F.A. By-Laws adopted July 1, 1944, again required revision in several respects due to subsequent Board actions. Recognizing the need for revision, Vice-President Wood moved

> THAT the President appoint a Standing By-Laws Committee to revise the By-Laws in accordance with past actions of the Board, and that such Committee include both Board members and non-Board members.

Motion seconded by Director Teetor and unanimously carried, President Walls to appoint such Committee.

9. *History of Foundry Industry.* The Executive Committee previewed the contents of a "History of the Foundry Industry" prepared by B. L. Simpson on his own initiative, in a desire to fill a long-felt need for an adequate description of developments within the foundry industry since the casting process first was discovered several thousand years ago. In view of the importance of such a work to all branches of the foundry indusry, and its logical connection with the educational work of A.F.A., it was suggested that A.F.A. should seek permission of the author for publication by A.F.A.

On motion by Director Wallis, seconded by Vice-President Wood, proposal to publish accepted, details to be worked out between the Secretary and the author.

10. *Quarterly Transactions.* Attention was called to the fact that publication of the Quarterly TRANSACTIONS had been suspended for 1945 when the Board authorized, on January 17, 1945, adoption of an emergency program of activity occasioned by cancellation of the Detroit Convention. Since then, all TRANSACTIONS material formerly published in the Quarterly has been carried in AMERICAN FOUNDRYMAN. In view of the success and economy of this procedure, Director Wilson moved that suspension of the Quarterly TRANSACTIONS become permanent, and that TRANSACTIONS material continue to be presented in AMERICAN FOUNDRYMAN.

Motion seconded by Director Teetor and carried.

11. *T.D.P. Advisory Committee.* Reappointment of the Advisory Committee of the Technical Development Program was considered and the need pointed out for some definite basis for succession in the choosing of a chairman. Accordingly, the following names were proposed for the 1945-46 Advisory Committee and the Secretary instructed to obtain acceptances:

Chairman S. V. Wood, Minneapolis Electric Steel Casting Co., Minneapolis.

Vice-Chairman George K. Dreher, Ampco Metal, Inc., Milwaukee, Wis.

J. H. Smith, General Motors Corp., Detroit, representing Malleable Iron.

Dr. D. Basch, General Electric Company, Schenectady, N. Y., representing Aluminum and Magnesium.

B. A. Miller, Baldwin Locomotive Works, Cramp Brass & Iron Foundries Div., Philadelphia, representing Brass and Bronze.

W. L. Woody, National Malleable & Steel Castings Co., Cleveland, representing Steel.

Max Kuniansky, Lynchburg Foundry Co., Lynchburg, Va., representing Gray Iron.

O. A. Pfaff, American Foundry Equipment Co., Mishawaka, Ind., representing Foundry Equipment.

C. J. Freund, Dean of Engineering, University of Detroit, Detroit, representing Educational interests.

Dean Hollister, Cornell University, Ithaca, N. Y., representing Educational interests.

12. *A.F.A. Membership Booklet.* The requests of several Chapters for a more elaborate piece of literature on A.F.A. and its activities were considered. On motion by Director Teetor, seconded by Director Wallis, the Executive Committee approved preparation of such a booklet at the earliest date.

13. *Foundry Personnel Problems.* Numerous requests having been received by the Staff from persons seeking new plant connections and available production capacity, due to reconversion problems, the Executive Committee urged that information on plant capacity and plants seeking supervisory personnel, be obtained for Staff information and record.

14. *Trade Association Cooperation.* Correspondence between the Staff and the Steel Founders' Society of a cooperative nature was read. On motion by Director Wilson, seconded by Director Teetor, the Executive Committee urged that a resolution of the Board be prepared and forwarded to the Steel Founders' Society in appreciation of their cooperation with A.F.A. and expressing the Association's intention to work for continued close relations.

15. *T.D.P. Funds.* The Treasurer pointed out an excessive amount of T.D.P. cash in banks at the end of the 1944-45 fiscal year, in view of the fact that the Finance Committee had authorized a combined budget covering both General and T.D.P. activities. Treasurer Hoyt suggested that $30,000.00 should be invested in bonds, $5,174.06 retained in a T.D.P. savings account, and that any unexpended appropriations held at the Northern Trust Co. in account of T.D.P. be transferred to the General accounts at the Harris Trust & Savings Bank.

On motion by Director Teetor, seconded by Director Wilson, proposed investment and transfer of funds approved.

16. There being no further business, meeting adjourned.

Respectfully submitted,

WM. W. MALONEY, *Secretary.*

Minutes of Executive Committee Meeting

Palmer House, Chicago—November 14, 1945

Presiding: President F. J. Walls
Present: President F. J. Walls
Vice-President S. V. Wood
Directors: R. J. Teetor
L. C. Wilson
W. B. Wallis
James H. Smith
Secretary Wm. W. Maloney
Secretary Emeritus R. E. Kennedy
Treasurer C. E. Hoyt
Director of T.D.P. N. F. Hindle
Educational Assistant H. F. Scobie
Assistant Treasurer J. Reininga

1. Minutes of the following meetings were read and approved:
Finance Committee Meeting—September 6, 1945
Executive Committee Meeting—September 7, 1945

2. *Membership Report.* The Assistant Treasurer presented figures showing total dues collections of $134,205.60 to October 31, against the Budget estimate of $135,000.00 for the entire fiscal year. Dues collections for the same period in 1944 totaled approximately $92,000.

The report also showed Resignations and Deaths of members totaling 287, and 525 Delinquents dropped, since July 1, 1945, a total of 812, as against 559 New Members added; a net loss of 253, giving a Book Membership of 7,562 on October 31, 1945.

The point was emphasized that in view of the effect of the new dues and reconversion conditions, both of which had previously been estimated, the National Office should concentrate on Membership work during the balance of the fiscal year. The report indicated a net loss in Members far below the estimates made prior to the January 1945 Board Meeting, at which time the dues increase was voted.

3. *New Chapters and Possibilities.* It was announced that following the petition from a group of Mexican foundrymen, the Board of Directors had approved by letter ballot formation of the 32nd chapter of A.F.A. in Mexico City. It was urged that Directors planning a West Coast or Mexico trip might well include a visit to the new Chapter in their itinerary and, in such event, should communicate with the A.F.A. Secretary in Chicago.

Announcement also was made that a group of foundrymen in Peoria were prepared to submit a petition for admission to Chapter status as the Central Illinois Chapter. Additional Chapter possibilities discussed included Pittsburgh, Pa.; New England, Kaukauna, Wis., and Vancouver, B. C. Members of the Executive Committee further encouraged consideration be given the formation of Chapter sections or groups.

Director Smith suggested the possibility of retaining a professional speaker acquainted with industrial subjects, to address Chapters on general problems within the foundry industry such as Education, Training, Manpower, etc.

4. *50th Anniversary Convention and Exhibit.* Reports were presented on progress of the Convention program and the 1946 Exhibit, indicating great interest on the part of both foundrymen and equipment and supply firms.

A lengthy discussion took place on the International aspects of the 1946 Convention, indicating that it would be highly undesirable, in view of present conditions of reconversion, to consider any A.F.A.-sponsored official plant tours by visitors from abroad. It was pointed out that considerable burden is involved in arranging, handling and conducting plant tours in our foundries under present conditions. Accordingly, the Executive Committee expressed the unanimous opinion that A.F.A. should not sponsor official plant tours of foreign foundrymen at the time of the 1946 Convention but that all arrangements for plant tours should be on an individual basis. The Secretary and Staff were instructed to act accordingly.

5. *Peter L. Simpson Memorial Award.* Announcement was made that A.F.A. had received and signed agreement papers with Past President H. S. Simpson, consummating the acceptance of Mr. Simpson's offer to endow a Peter L. Simpson Memorial Award of A.F.A. Under this agreement, Mr. Simpson deposited with A.F.A. an Endowment Fund of $5,000.00 plus an amount to cover the cost of striking off a medal for presentation in 1946. The award is intended to be given to individuals "for noteworthy achievements during the previous year" and may be awarded annually. In order to authorize the work of designing and preparing dies for the medal involved, Director Smith moved that A.F.A. stand the cost of preparation and dies. Seconded by Director Wallis and unanimously carried.

6. *Standing By-Laws Committee.* President Walls announced the appointment of a Standing By-Laws Committee as follows:
Chairman, H. Bornstein, Testing & Research Division, Deere & Co., Moline, Ill.
Horace A. Deane, American Brake Shoe Co., 230 Park Ave., New York.
W. V. Johnson, Oliver Corp., South Bend, Ind.
W. B. Wallis, Pittsburgh Lectromelt Furnace Corp., Pittsburgh, Pa.
R. F. Lincoln, Russell F. Lincoln & Co., Lakewood, Ohio
E. W. Horlebein, Gibson & Kirk Co., Baltimore, Md.
C. E. Hoyt, National Office, Chicago (Consultant)
Wm. W. Maloney, A.F.A. Secretary, exofficio.

The Committee personnel having the full approval of the members of the Executive Committee, the Secretary was instructed to secure the necessary acceptances and to arrange for a meeting in the near future.

7. *1945 Bound Volume of* TRANSACTIONS. Following discussion, it was agreed that one copy of the 1945 Bound Volume of TRANSACTIONS would be offered gratis to each Company and Sustaining member of record, but that volumes would be offered all other members only at a discounted price. It was recognized that in some cases it might be necessary to provide more than one gratis copy to certain Company and Sustaining members who have long been granted this privilege.

8. *Technical Activities Correlation Committee.* It being pointed out that the Technical Activities Correlation Committee had not been reappointed for 1945-46, the question was raised as to whether this Committee and the Advisory Committee of T.D.P. actually overlap in their respective functions. President Walls declared that he would study the matter and consult further with members of the Executive Committee as to the relative status of these two groups.

9. *T.D.P. Solicitation.* The Executive Committee reiterated that the solicitation of funds for T.D.P. would cease after the end of the calendar year 1945, and that thereafter the concentration would be on Sustaining memberships. The Secretary stated that one last solicitation was planned for 1945.

10. *National Metals Congress.* Following discussion, Director Wallis moved authorization for an A.F.A. exhibit booth at the National Metals Congress in February, 1946, at which time members of the National Office staff would be available for discussion of Exhibit space as well as advertising in AMERICAN FOUNDRYMAN. Seconded by Director Smith and carried.

11. *Educational Program.* T.P.D. Director N. F. Hindle presented for consideration an outline for a long-range National Educational Program calculated to be of lasting benefit to the entire Foundry Industry. Following discussions, the Executive Committee expressed enthusiastic approval of the program and offered a vote of confidence to T.D.P. Director Hindle and H. F. Scobie for their work in preparing the first comprehensive and detailed outline of A Foundry Educational Program. The Staff was directed to put the plan into operation.

12. *1946 Apprentice Contest.* Following discussion of the 1946 National Apprentice Contests of A.F.A., Director Wallis moved that 1st Prize winners of each group (Gray Iron, Steel, Brass & Bronze, Patternmaking) be brought to the 50th Anniversary Convention at the expense of A.F.A., as a part of the Educational Program, in the event their respective companies might be unable individually to take care of the expense involved. Seconded by Director Wilson and carried.

Management Sessions. It was reported that a number of Chapters had displayed interest in holding Management-Labor sessions and that considerable interest has been aroused wherever these have been held. It was stated that the Staff would continue to book additional meetings of this type so as to acquaint "top" management of foundries with the work of their local Chapters.

14. AMERICAN FOUNDRYMAN. Secretary Maloney reported on attendance at the 1945 annual meeting of F.E.M.A., at which time he discussed in open meeting the subject of advertising in AMERICAN FOUNDRYMAN and explained the reasons why the program had been adopted and continued by the A.F.A. Board. He reported that some questions had been asked from the floor and expressed the thought that it would be necessary to constantly build up the prestige of the AMERICAN FOUNDRYMAN through improved editorial content. He stated that the budget of advertising previously set for 1945 had been surpassed but that considerable work would have to be done in order to equal or better the average issue during 1946.

15. *Staff Considerations.* In Executive session, the Executive Committee considered an Organizational Plan presented by the Secretary and approved the employment of Frank E. Wartgow, formerly of American Steel Foundries and most recently with Hasbrouck Haynes Engineers of Chicago, as a member of the headquarters staff to concentrate first on staging the 1946 Exhibits and thereafter on Chapter contacts. All members of the Finance Committee being present, approval was given proposed compensation.

Other details of the Staff Organizational Plan were discussed at length and the Secretary authorized to proceed along lines agreed upon.

16. There being no further business, the meeting was declared adjourned.

Respectfully submitted,

WM. W. MALONEY, *Secretary.*

Minutes of Executive Committee Meeting

Palmer House, Chicago—January 15, 1946

Presiding: President F. J. Walls
Present: President F. J. Walls
Vice-President S. V. Wood
Directors: R. J. Teetor
L. C. Wilson
W. B. Wallis
Secretary Wm. W. Maloney
Secretary Emeritus R. E. Kennedy
Director T.D.P. N. F. Hindle
Treasurer C. E. Hoyt
Assistant Treasurer J. Reininga

1. The Minutes of the Executive Committee Meeting held in Chicago November 14, 1945, were read and approved.

2. *Membership Report.* The Assistant Treasurer presented figures showing total dues collections of $141,775.93 to December 31 (budget for entire fiscal year $135,000), against $98,192.88 on December 31, 1944.

The report showed Resignations and Deaths of Members totaling 311 and 592 Delinquents dropped since July 1, 1945, a total of 903; as against 829 New Members added.

It was brought out that Book Membership on June 30, 1945, totaled 7,815, against 7,741 on December 31, 1945, a net loss of 74. In view of the estimates of possible Membership losses made by the Special Sub-Committee on Dues Structure in December, 1944, and the end of the intensive war production period, it was pointed out that the actual loss during the previous six months had been less than 1% of the total Membership. The Special Sub-Committee had estimated that the loss due to increasing the dues might run as high as 12.5%. The following tabulation shows losses and gains in Membership during the six months period June 30-December 31, 1945:

Class	*As of June 30, 1945*	*As of Dec. 31, 1945*
Sustaining	239	194
Company	1,571	1,436
Personal	1,062	1,237
Personal (Affiliate)	4,419	4,272
Personal (Associate)	241	258
Apprentice and Student	23	19
Foreign	205	270
Honorary	55	55
Total	7,815	7,741

Director Wilson moved that the method of pro-rating of Dues on the fiscal year basis be abandoned and that Memberships be accepted and billed on a yearly basis on the first of the month following date of receipt of membership. Motion seconded by Director Wallis and carried.

3. *Treasurer's Report.* The report of the Treasurer was submitted, comparing the fiscal year budget with Income and Expense for the first six months ending December 31, 1945: Actual income, $232,945.54, or approximately 68% of the total income budget of $341,800.00; actual expense, $105,150.17, or approximately 30% of the budgeted expense of $341,800.00.

The Committee declared A.F.A. finances for the first six months of the current fiscal year to be in excellent shape and that, barring circumstances unforeseen, should enable the Association to stay well within the budgeted figures.

4. *Technical Development Program.* Director of the Program pointed out that, in addition to the "Cupola Handbook," several other publications are now in the making. These include an outline of the "Causes of Gray Iron Castings Defects," a bibliography on Permanent Molds, a bibliography on Centrifugal Castings and an "Inspection Handbook." The Director also reported that a "Symposia on Malleable Sand Practices," prepared for the 1945 Annual Convention, would be published in pamphlet form.

Director Hindle requested that a sum of $2,000.00 be allowed the Battelle Memorial Institute for work on the Heat Transfer project. It was moved by Director Wallis, seconded by Director Wilson and carried, that this sum be turned over to Battelle at the earliest possible date.

Brief discussion was held as to the advisability of continuing solicitation for further T.D.P. contributions. Without vote, the Committee agreed unanimously that no further solicitation of T.D.P. funds should be contemplated for at least another year, when results from the program become more evident.

5. AMERICAN FOUNDRYMAN. The Secretary announced complete redesigning of AMERICAN FOUNDRYMAN in order to improve its attractiveness for both readers and advertising men, and steps taken to add to the editorial staff. He reported no adverse comments on the magazine from advertising agencies and again stressed the point that building the AMERICAN FOUNDRYMAN as an advertising medium would take considerable time but that, in the meanwhile, every effort would be made to improve and broaden the magazine's editorial content.

6. *Safety and Hygiene Committee.* On invitation, James R. Allan of International Harvester Co. and C. F. Larsson of American Air Filter Co., appeared before the Committee and called attention to a serious problem confronting the industry in its Safety and Hygiene efforts. Mr. Allan stated that he would no longer be able to continue as Chairman of the Safety and Hygiene Committee and recommended that A.F.A. add to its Staff a full time employee, fully experienced in industrial, mechanical and legal aspects of Dust Control opera-

tion. He urged that immediate steps be taken to insure continuation of Safety and Hygiene operation between the industry and Governmental agencies in the various states concerned with such problems, to the end that such Safety and Hygiene Codes might be prepared on a factual and engineering basis in such a manner as not to seriously affect the ability of foundries to comply.

Following discussion, it was agreed that a Special Committee be appointed to make recommendations to the Board of Directors at the earliest possible date on the proposals and recommendations submitted relative to Safety and Hygiene work. Accordingly, President Walls appointed: L. C. Wilson, Chairman; R. T. Rycroft; F. J. Dost, and R. J. Teetor. The Committee was instructed to discuss the matter with interested parties so as to learn the scope of the problem and in order to prepare a proper basis for possible future action by the Board.

7. *Chicago Museum of Science and Industry Exhibit.* The Secretary presented a letter from the Chairman of the Chicago Chapter concerning the need for funds to continue the operating Foundry Exhibit at the Chicago Museum of Science and Industry. This foundry exhibit, one of the most popular exhibits at the Museum, has long been supported solely by the Museum and the Chicago Chapter, following the original conception and installation accomplished by the National Office. The Chicago Chapter stated that a total sum of $2,300.00 would be needed to continue the Foundry Exhibit on a five-day schedule of operation throughout the year and suggested that A.F.A. contribute to the needed operating funds.

Following discussion, it was moved by Director Wallis and seconded by Director Wilson, and carried, that the Executive Committee recommend to the Board an expenditure of $2,000.00 on the Museum Foundry Exhibit for the period February 1, 1946, to February 1, 1947, if the Chicago Chapter would contribute the necessary additional $300.00 a year for maintenance.

It was also agreed by the Committee that more publicity should be given this Exhibit in AMERICAN FOUNDRYMAN and that complete information be placed in the hands of all Chapter Chairmen as an example of a worthwhile Chapter activity.

8. *Copyright of Convention Papers.* It was pointed out that A.F.A. has not been protected by copyright in the publication of papers offered the Association for presentation at the Annual Convention. A motion was made by Director Teetor, seconded by Director Wallis and carried that A.F.A. should maintain prior right of publication over all Convention papers, and that all such papers should be copyrighted by the Association. It was understood that this action applied only to publications and not to the remarks of individual speakers.

Following discussion, the Executive Committee approved pre-printing, prior to the 1946 Convention, a number of papers published in AMERICAN FOUNDRYMAN during 1945, such papers having been submitted for the cancelled 1945 Convention. Action taken on motion by Vice-President Wood, seconded by Director Teetor.

9. There being no further business, the meeting was adjourned.

Respectfully submitted,

WM. W. MALONEY, *Secretary.*

Minutes of Midyear Meeting of Board of Directors

Palmer House, Chicago—January 16, 1946

Presiding: President F. J. Walls

Present: President F. J. Walls
Vice-President S. V. Wood
Directors: F. J. Dost, G. K. Dreher, D. P. Forbes, R. M. Jacobs, H. R. Judson, Max Kuniansky, H. Reitinger, S. D. Russell, R. T. Rycroft, J. Sully, R. J. Teetor, W. B. Wallis, L. C. Wilson, F. M. Wittlinger
Secretary Wm. W. Maloney

1. Minutes of the Board Meeting held July 19, 1945, were dispensed with in view of the fact that they had previously been sent all Officers and Directors for approval.

Minutes of the following Executive Committee Meetings were read and approved:

(a) Executive Committee Meeting—September 6, 1945.
(b) Executive Committee Meeting—November 14, 1945.
(c) Actions taken by the Executive Committee at meeting held January 15, 1946.

2. TRANSACTIONS. In response to question, it was stated that the 1945 Bound Volume of TRANSACTIONS would again be published, but henceforth in the same size as the AMERICAN FOUNDRYMAN magazine.

3. *Report of Finance Committee.* Chairman Teetor presented the report of the Finance Committee in the form of (1) Combined Income and Expense report, and (2) Statement of Receipts and Disbursements, both covering the period July 1-December 31, 1945. Particular attention was called to the statement of Income, showing 68% of the year's Budget Income received, only 30% expended, during the first six months of the current fiscal year.

It was noted that the year's Budget of Dues Income had already been exceeded by some $6,000; 58% of budgeted Advertising income already received; that Income item of $26,000 for Space Rentals was not representative of the fact that over 75% of total Exhibit Space available had been applied for by January 1. In connection with Expense items, it was commented that Salaries totaled, for six months, less than 50% of Budget; expense for Office Rent less than contemplated, due to inability to obtain desired larger quarters; that the large proportion of Dues Refunds paid out during the first six months was natural because of large percentage of memberships falling due on July 1 each year.

The Treasurer called attention to the fact that a statement of Net Worth of the Association on December 31, 1945, would show, in general figures (1) Cash On Hand, $143,000, and (2) Securities on Investment, $182,000; indicating a Total Net Worth on December 31, 1945, exclusive of Inventories, of approximately $325,000. The Treasurer estimated that A.F.A. Income for the last six months of the current fiscal year would probably take care of all expenses incurred and that this would leave the Association's Financial Condition on June 30, 1946, in far better shape than ever before experienced.

4. *Report of Membership Committee.* Chairman Wood reported a Membership position and Dues Income as of December 31, 1945, far exceeding estimates of a year ago as to the losses that might be incurred as a result of increased dues and the end of World War II. He pointed out dues collections, in six months in excess of the entire year's Budget. He called attention to the fact that estimates made a year ago took into consideration a loss of as much as 12.5% in membership as a result of doubling the Company and Sustaining dues, whereas the actual net loss in the first six months was less than 1%. On June 30, 1945, Book Membership stood at 7815; on December 31, 1945, 7741; net loss, 74.

In response to question, Chairman Wood stated that this net loss in Membership during the first six months had been more than made up by new memberships received after January 1, 1946.

5. *Chapter Contacts Committee.* Chairman Wood reported a large number of Chapters visited by Officers and Directors and members of the Headquarters Staff during the current fiscal year. He also reported steps taken to add to the Staff a full-fledged Chapter Contacts man whose duty it will be, following the Convention, to maintain continuous and frequent contact with all Chapters.

He urged National Directors play a more active part in visiting Chapters within their own territories and when on trips to other cities at the time Chapter meetings are scheduled. President Walls urged Directors planning a Western trip to consider similar visits to Chapters on the Pacific Coast, in Mexico, and in Texas.

Director Reitinger commented on favorable reactions to the Annual Chapter Chairman Conferences, as a result of frequent visits to many Chapters, especially in the Eastern area.

6. *Management Committee.* Director Rycroft reported three Management sessions held in the current fiscal year—at Northern Illinois-Southern Wisconsin, Northwestern Pennsylvania and Toledo Chapters. He stated that efforts were being made to develop other sessions and urged the attendance of as many Directors as possible wherever such meetings are held. President Walls pointed out that Management meetings have resulted in increased Chapter membership, with most favorable reaction on the part of the top Management of foundries attending.

7. *Youth Encouragement Committee.* Director Sully reported briefly on progress by his Committee, especially in the Canadian Chapter territories. Other Directors commented on the great need for encouraging further activities on the part of the local Chapters toward interesting youths to enter the foundry industry. Several Directors suggested various means of accomplishing the aims of the Committee, and the National Office Staff was urged to do everything possible to assist schools in acquiring surplus foundry equipment for instruction in foundry practice.

8. AMERICAN FOUNDRYMAN *Policy Committee.* Chairman W. B. Wallis called on the Secretary for a report on AMERICAN FOUNDRYMAN advertising, showing gross Advertising Revenue from eight (8) issues in 1945, $61,295; Agency commissions paid, $7,035; Net, $54,260. It was stated that average gross for eight months was $7,661; average net, $6,782. The Secretary stated that the editorial content of the magazine was being completely redesigned with the February, 1946, issue. He stated that the reaction of advertising agencies has been most favorable and that prospects for 1946 were good, including prospects for a Golden Anniversary number to be published in April, just prior to the 50th Anniversary Convention.

In response to question, the Secretary stated that effort was being made to maintain the volume of editorial matter in any one issue at 50%, whereas the average magazine operates on a basis of 30-33%.

9. *T.D.P. Advisory Committee.* Chairman Wood reported that the T.D.P. Advisory Committee had been unable to hold a meeting during the current fiscal year. He reported that the *Cupola Handbook* had been scheduled for publication by the time of the 1946 Convention, other papers in line for nearby publication. In addition, he reported briefly on research being conducted under the Technical Development Program at Cornell University, Columbia University and Battelle Institute.

10. *1946 Nominating Committee.* Past President Wilson, as Chairman of the 1946 Nominating Committee, presented the following report:

"The 1946 Nominating Committee, as appointed by President F. J. Walls, met in Chicago January 14, 1946, to consider and nominate candidates for the following offices in accordance with the A.F.A. By-Laws:

President, to serve 1 year.
Vice-President, to serve 1 year.
5 Directors, to serve 3 years each.

"After due deliberation, the Nominating Committee presents, for approval by the Board of Directors, the following nominations:

For President, S. V. Wood, President, Minneapolis Electric Steel Casting Co., Minneapolis.

For Vice-President, Max Kuniansky, Vice-President, Lynchburg Foundry Co., Lynchburg, Va.

For Directors:

H. A. Deane, Works Manager, American Brake Shoe Co., New York (Gray Iron—Metropolitan Chapter)

Stowell C. Wasson, Manager, National Malleable & Steel Castings Co., Chicago (Malleable Iron—Chicago Chapter)

H. G. Lamker, Superintendent of Foundries, Wright Aeronautical Corp., Paterson, N. J. (Aluminum & Magnesium—Metropolitan Chapter)

J. E. Kolb, Pattern Shop Superintendent, Caterpillar Tractor Co., Peoria, Ill. (Patternmaking—Central Illinois Chapter)

B. L. Simpson, President, National Engineering Co., Chicago (Equipment & Supplies—Chicago Chapter)

"The Nominating Committee, in submitting this report, desires to point out the relative impossibility for any group of nine men to know personally all available and deserving candidates. Therefore, the Committee recommends, for future nominating procedure, that the individual Chapter Boards be asked to submit names of candidates known to them, together with complete information on their qualifications, for consideration by the Committee."

11. *50th Anniversary Foundry Congress.* The Secretary reported on progress in connection with the 50th Anniversary Convention and Exhibit. He stated that all Members and Exhibitors had been mailed the official hotel application form, along with a Tentative Schedule of Convention Sessions. He reported some 200 Exhibit applications received to date, calling for over 75% of total space available.

The Secretary reported considerable convention interest from England, but stated that no A.F.A.-sponsored tour would be conducted although the Association will continue to cooperate in assisting foreign visitors to visit foundry plants willing to receive them.

12. *Executive Committee Recommendations.* The following recommendations from the Executive Committee meeting of January 15, were considered:

(1) Recommendation that the method of pro-rating Dues on the fiscal year basis be abandoned progressively and that memberships hereafter be accepted and billed on a yearly basis, on the 1st of the month following date of receipt of membership. On motion duly seconded, approved.

(2) Recommendation that National A.F.A. funds, to the extent of $2,000.00, be expended to maintain a 5-day foundry operating exhibit at the Chicago Museum of Science and Industry. On motion duly seconded, approved.

(3) Recommendation that the Association copyright all papers presented at the Annual Conventions and maintain the stipulation with authors that the Association is entitled to prior publication rights. On motion duly seconded, approved.

13. There being no further business, the meeting was adjourned.

Respectfully submitted,
WM. W. MALONEY, *Secretary.*

Minutes of Board of Awards

Hotel LaSalle, Chicago—March 11, 1946

Presiding: L. C. Wilson, Chairman pro tem.
Present: Past Presidents L. C. Wilson, D. P. Forbes, L. M. Shannon, H. S. Simpson.
Secretary Emeritus R. E. Kennedy, Acting Secretary.

1. In the absence of Chairman R. J. Teetor, Director Lee Wilson served as Chairman pro tem.

2. As the first order of business, a motion was made and seconded that the Board of Awards join with the Executive Committee at the latter's meeting on March 12 for the purpose of dispatching a message of deepest sympathy to Chairman R. J. Teetor and Mrs. Teetor on the death of their son Lt. James Teetor, killed March 1st near Albuquerque, N. M.

3. Acting Secretary Kennedy reported on the following standing of the Board of Awards funds as of February 28, 1946, as follows:

McFadden Fund	$ 763.69
Penton Fund	615.22
Seaman Fund	195.91
Whiting Fund	709.26
Obermayer Fund	463.84
Simpson Fund	125.00
	$2,872.92

4. The A.F.A. Secretary's summary of disposition of medals awarded by the Board of Awards during the past several years as contained in a letter dated December 11, 1945, was read.

5. Decision was reached by the Board of Awards members present that the number of medals to be awarded would hinge upon consideration of available candidates. A number of names of candidates then were submitted, and after due consideration the necessary motions were made, seconded and carried that the Board of Awards recommends to the Board of Directors

THAT the Wm. H. McFadden Gold Medal of A.F.A. be awarded to Hyman Bornstein, Director of Laboratories, Deere & Company, Moline, Ill., "for his many and outstanding contributions to the foundry industry, particularly in the field of gray cast iron."

THAT the John H. Whiting Gold Medal of A.F.A. be awarded to Peter Blackwood, Foundry Superintendent, Ford Motor Co. of Canada Ltd., Toronto, Ontario, "for his development work in and influence on free interchange of information pertaining to centrifugal castings."

THAT the Peter L. Simpson Memorial Medal of A.F.A. be awarded to Howard F. Taylor, formerly of the Naval Research Laboratory and now Research Associate, Massachusetts Institute or Technology, Cambridge, Mass., "for his unfailing interest in and contributions to foundry research, and for his influence on the broadening of naval research in the field of cast metals."

Chairman Wilson stated that a certificate of Honorary Life Membership would accompany the above medal awards.

A motion was made, seconded and carried that no further medals be awarded for the year 1946.

5. After due consideration of the awarding of Honorary Life Memberships in the Association, motions were made, seconded and carried unanimously, that the Board of Awards recommends to the Board of Directors

THAT Honorary Life Membership be conferred upon Harold J. Roast, Vice-President, Canadian Bronze Co., Ltd., Montreal, Quebec, "in recognition of his many and long continued contributions to the literature and cooperative enterprise of the foundry industry, both in the United States and Canada."

THAT Honorary Life Membership be conferred upon William J. Coane, Vice-President, Ajax Metal Co., Philadelphia, "because of his long and continued service to the industry, and in recognition of the fact that, as one of the few living men who attended the first convention of A.F.A. at Philadelphia in 1896, he has remained continuously active in the industry."

THAT Honorary Life Membership be conferred upon President Fred J. Walls, Metallurgist, International Nickel Co., Detroit, on completion of his term as President of the Association, and in recognition of his service to the Association and the foundry industry.

6. In considering some suitable recognition for annual Foundation Lecturers, motion was made, seconded and carried that an appropriately worded and engrossed certificate be prepared for presentation to the 1946 Foundation Lecturer.

7. Consideration was given some form of recognition to a company for outstanding works and developments made freely available to the industry, as a result of which the Board of Awards requested authorization by the Board of Directors to develop an appropriate certificate that might be awarded to companies found worthy of such a citation.

8. Having completed its duties for the year, the Board of Awards desires to go on record with further recommendations having to do with the procedure under which the Board is expected to function. In analyzing potential candidates for medal awards, the Board finds itself with meager information as to qualifications of individuals who have performed outstanding service to the industry, and it feels the need for much more complete information with which to judge soundly the relative merits of candidates proposed.

The Board feels, as do the members of the Nominating Committee which reported to the Board of Directors at its January 1946 meeting, that the Board of Awards cannot accomplish the purpose for which it was formed without more adequate coverage of the accomplishments of men engaged in the foundry industry. Basically, a Board of seven members cannot intimately know and, therefore, evaluate the service rendered by scores of worthy individuals, because of lack of records available outlining such qualifications. It does not feel that the few names annually submitted for analysis at the Board of Awards meetings sufficiently cover outstanding work constantly being performed.

After careful consideration of the problem, the Board recommends to the Executive Committee a reorganization of the procedures guiding the Board of Awards, whose express function is to judge the merits of cases presented. While the Board does not presume to recommend inflexible rules of procedure, it does present to the Executive Committee for consideration the following thoughts and expressed suggestions:

(1) Because the Chairman of the Board of Awards is the last retiring president, lacking essential background of the work of the Board, the Board feels that this procedure should be altered so that the oldest ranking retired president should be designated to act as Chairman of the Board of Awards.

(2) In order to present to the Board of Awards adequate information on outstanding accomplishments of men in the industry, the Board recommends that a Staff appointment be made for the specific purpose of gathering information through published articles and contact from members of chapters and the industry in general, on the developments and improvements constantly occurring. Such a compilation of technical developments by individuals responsible for them should be a work of continued nature, to be kept up-to-date in every way.

(3) The Chairman of the Board of Awards annually should designate from among its members, a secretary to collaborate with the National Office Staff in selecting the most outstanding candidates for presentation to and analyses of the Board of Awards.

Respectfully submitted,

L. C. WILSON, *Chairman pro tem*
D. P. FORBES
L. M. SHANNON
H. S. SIMPSON

Minutes of Executive Committee Meeting

A.F.A. Headquarters, Chicago—March 12, 1946

Presiding: President F. J. Walls
Present: President F. J. Walls
Vice-President S. V. Wood
Directors: L. C. Wilson
W. B. Wallis
J. H. Smith
Secretary Wm. W. Maloney

1. Minutes of the January 15, 1946, Executive Committee meeting were read and approved. Reading of the minutes of the January 16, 1946, Board of Directors meeting were dispensed with, these minutes having been sent to all A.F.A. Directors previously.

2. *1946 Convention and Exhibit.* A report on the 50th Anniversary Convention and Exhibit was presented by the Secretary in the form of an outline program of technical sessions and list of current exhibitors. The report showed income on space rentals as of February 28, $47,362.00, against the budget figure of $84,000.00, with the major portion of collections still to be made.

The Secretary stated that invitations had been issued to the foundry trade associations representing steel, malleable, gray iron, non-ferrous, magnesium, foundry equipment and the National Founders' Association to be represented at the foundry show with an exhibit, space to be provided by A.F.A. gratis. He stated that favorable replies had been received from the gray iron, malleable, non-ferrous and foundry equipment associations.

With regard to foreign attendance, the Secretary stated that between fifty and sixty foundrymen from abroad had indicated their intention of attending. In line with previous Board policy, no official tour is being sponsored by A.F.A. but all convention facilities will be made available and some provision will be made for their housing.

On the subject of housing at the Convention, the Secretary stated that in all probability many visitors would have to be housed in private homes because of the great expected attendance. It was stated that members and exhibitors were being given priority in the matter of accommodations and that every effort would be made to provide rooms for National Officers and Directors, Chapter Officers and Directors and persons scheduled on the program.

3. *Chapter Activities.* Several chapter possibilities were discussed, particular attention being given efforts to create a new Seattle, Wash., chapter. The Secretary stated that Fred Sefing of International Nickel Co., on a west coast trip at the time, would address the Seattle group on March 19, at which time he would endeavor to interest them still further in forming an A.F.A. chapter.

The possibility of a request for a chapter in Sweden was mentioned by President Walls. After discussion, it was agreed informally that it would not be best to establish chapters in European countries but that A.F.A. should co-operate with foreign foundry groups in every possible way, as in the past, and might make some special arrangements for the handling of A.F.A. literature abroad.

4. *Technical Publications.* It was stated in connection with the forthcoming A.F.A. publications that the new *Cupola Handbook* would be made available at the convention in May with sufficient copies on hand for display and taking orders.

The Secretary reported that the forthcoming *History of the Foundry Industry* by B. L. Simpson would not be available by convention time in view of the fact that the printer had been unable to obtain paper stock sufficient for its printing. It was stated that the history would be published at the earliest possible date.

A letter from R. L. Lee as Chairman of the A.F.A. Cost Committee was presented containing the recommendation "that A.F.A., in cooperation with the Gray Iron Founders' Society and Non-Ferrous Founders' Society, publish a booklet of the cost systems of the gray iron and non-ferrous groups, including in the publication the Cost Committee Classification of Foundry Cost Factors, also Recommendations to Buyers of Castings, and Recommended Material for a Cost System for Small Foundries" and that such a booklet be made available to the foundry industry. Following discussion, the Executive Committee expressed its opinion that such a booklet should not be issued by A.F.A. in view of the fact that (1) the material has already been widely distributed by the various trade societies, and (2) that A.F.A. should not enter into the matter of specific costs of foundry jobs. The opinion was expressed that A.F.A. should continue to encourage the use of cost systems prepared for specific branches of the industry by the various trade associations.

The Secretary presented a letter dealing with possible sponsorship by A.F.A. of a translation of books by Piwowarsky on gray cast iron but pointed out that Battelle Memorial Institute was also interested in this translation work. After discussion, the Executive Committee expressed the opinion that Battelle was in a better position to sponsor this translation than was A.F.A., and instructed the Secretary to notify Battelle accordingly.

The Secretary called attention to the fact that A.F.A. has received a number of requests from libraries in devastated portions of Europe asking the Association to provide them with books on foundry practice gratis. The Executive Committee expressed the opinion that A.F.A. should cooperate with such organizations in every possible way and suggested that such requests might well be discussed with delegates from abroad at the time of the Cleveland convention.

5. *Chapter Publications.* A projected publication of the Northern California Chapter on sand mulling was discussed at some length. The Chapter's original plan was to publish 10,000 copies of the booklet and to ask the sand mixer manufacturers to subsidize the cost of the booklet through advertising at a cost of some $2,500. The Secretary stated that when this plan was first divulged, A.F.A. had written the Chairman of the project and had phoned Director S. D. Russell in San Francisco pointing out the inadvisability of such a program. He reported that the plan had been abandoned and that the Chapter would issue a much smaller number of copies for distribution on the west coast at nominal cost per copy.

Members of the Executive Committee expressed the opinion that there should be some closer tie between the National Office and Chapters with respect to future Chapter publications, and the Secretary was instructed to place the matter on the docket of the 1946 Chapter Chairman Conference.

6. AMERICAN FOUNDRYMAN. A report on the AMERICAN FOUNDRYMAN, both editorial and advertising, was presented and discussed. The report showed that the AMERICAN FOUNDRYMAN had exceeded its 1945 budget for advertising revenue and that a Pre-Convention issue would be published in April, 1946. It was stated that a growing acceptance on the part of advertisers had become more evident in recent months and that this has led to a complete redesigning of the magazine, effective with the February, 1946, issue.

The Secretary stated that a questionnaire on the new design and other features of the magazine had been sent all members in the United States and Canada and that the reports were most favorable with regard to reader interest and appreciation of the AMERICAN FOUNDRYMAN editorial content.

7. *Chapter Chairman Conference.* The 1946 Chapter Chairman Conference was discussed and the Secretary instructed to issue a call at the proper time for this meeting, to be held during the same week of the annual Board meeting in July, on the preceding two full days.

8. *Safety and Hygiene.* Director Wilson, as Chairman of the special committee on Safety and Hygiene, reported on his committee's work in sounding out the desirability of a permanent Industrial Hygiene Department of A.F.A. He stated that a number of opinions had been obtained on the subject and on the question of state codes of hygiene but reported that further conversations would be held with the American Standards Association and that his committee would report back to the Executive Committee with recommendations at the earliest possible date.

9. *Resolutions.* The deaths of Past President S. Wells Utley of Detroit Steel Casting Co. and Past Director S. C. Vessy of W. W. Sly Mfg. Co. were reported and the Secretary instructed to draw up resolutions of sympathy accordingly.

10. *Technical Development Program.* It was stated that the amount of accumulated cash in the Technical Development Program account at the Northern Trust Co. made advisable the reinvestment of some $15,000.00 in U. S. Government Bonds, in view of the fact that the Technical Development Program now is included in a combined budget for the entire work of A.F.A. On motion by Director Wilson, seconded by Director Wallis, it was voted unanimously to reinvest $15,000.00 of T.D.P. funds in appropriate U. S. Government Bonds.

11. *Resignation of J. Reininga.* The Secretary reported to the Executive Committee that Jane Reininga, Assistant Treasurer of the Association and a Staff member for 30 years, had expressed her intention of retiring on March 15, 1946, due largely to health considerations brought about by the increasing burdens involved in servicing a steadily expanding Membership with all the resultant details. Members of the Executive Committee were unanimous in their expressions of appreciation for her long and faithful period of service in connection with Membership work, finances, Convention activities and many other phases of the Association development. With deepest regret the Committee voted to accept Jane Reininga's retirement intentions, and at the same time, voted unanimously to extend to her separation compensation as some measure of their appreciation for extended service. President Walls, in making this presentation, expressed on behalf of the Association the hope that Jane would enjoy many years of leisure and happiness and invited her to be a guest of the Association at the 50th Anniversary Convention in Cleveland.

12. *Staff Additions.* The Secretary reported that N. F. Hindle, Director of T.D.P., had suffered a nervous collapse in February and had been confined to his bed for at least two months at his home at 5 W. Warren St., Calumet City, Ill. He reported that in view of Norm Hindle's illness, immediate steps had been taken with the approval of President Walls to hire S. C. Massari, formerly Chief Metallurgist, Association of Chilled Car Wheel Manufacturers, and lieutenant colonel with the Chicago Ordnance District, on a temporary basis to head up and carry through the technical program for the Cleveland convention. He also reported that John P. Mullen, formerly Publicity Director for the Gray Iron Founders' Society, had also been retained on a temporary basis to carry through publicity and other matters for the convention. It was agreed that further discussion relative to future work with A.F.A. would be held at a later date.

13. *Membership.* A Membership Report was submitted by the Secretary showing book membership on February 28, 1946, of 7,943, compared with 7,815 on June 30, 1945, and 7,194 on January 1, 1945. It was stated that 1,100 New Members had been added to the rolls between last June and February 28, with a loss of 339 due to resignation and deaths and 633 for non-payment of dues, a total of 972; a net gain of 128, in spite of reconversion, dues increases and the like.

The report showed a total dues collection for the current fiscal year to February 28, 1946, of $145,953, against collections of $100,948 for the same period one year ago.

14. *Finances.* A financial report was submitted to the Executive Committee in the form of a "Net Worth Statement" showing the following increases in various accounts during the eight-month period June 30, 1945, to February 28, 1946; (1) Checking Accounts Increase, $69,568.57; (2) Savings Account, $1,276.40, and (3) Investment Securities, $40,000.00.

The net worth statement showed the following itemization:

Checking Accounts	*As of June 30, 1945*	*As of Feb. 28, 1946*	*Increase or Decrease*
General Funds	$17,439.54	$113,546.50	$96,106.96
Cupola Research	12,401.97	12,016.69	385.28*
Technical Development Program	45,003.92	18,850.81	26,153.11*
Total Checking	$74,845.45	$144,414.00	$69,568.57
Savings Accounts			
Award Funds Interest	$ 2,809.74	$ 2,914.45	$ 104.71
T.D.P. Interest	183.34	509.81	326.47
General Reserve Interest	9,295.69	10,140.91	845.22
Total Savings	$12,288.77	$ 13,565.17	$ 1,276.40
Investments			
Securities:			
Award Funds	$ 21,349.60	$ 26,349.60	$ 5,000.00
Cupola Research	20,000.00	20,000.00	—
General Reserve	40,644.30	40,644.30	—
T.D.P. Funds	50,000.00	85,000.00	35,000.00
Total Investments	$131,993.90	$171,993.90	$40,000.00
Increase, 8-month period			$110,844.97

*Decrease.

15. There being no further business, the meeting was adjourned.

Respectfully submitted,
WM. W. MALONEY, *Secretary.*

Minutes of Executive Committee Meeting

Hotel Statler, Cleveland—May 7, 1946

Presiding: President F. J. Walls
Present: President F. J. Walls
Vice-President S. V. Wood
Directors: R. J. Teetor
L. C. Wilson
W. B. Wallis
J. H. Smith
Secretary Wm. W. Maloney

1. Minutes of the March 12, 1946, meeting of the Executive Committee were read and approved.

2. *Financial Reports.* A report on finances was submitted by the Secretary in two parts. The report on Income and Expenses showed as of March 31, 1946, an excess of $15,057.17 of income over the budget of $341,800.00 for the fiscal year; and total expense $158,574.31 less than the expense budget.

A statement of assets as of March 31, 1946, compared with June 30, 1945, showed a total increase in assets of $166,896.63 for the 9-month period. Included was an increase of $125,470.23 in checking accounts, an increase of $1,426.40 in savings accounts, and an increase of $40,000.00 in securities investments. Total assets on March 31, 1946, $386,024.73.

3. *Membership.* A Membership report was submitted by the Secretary showing book membership of 8,073 on March 31, 1946, as against 7,338 on March 31, 1945. Dues collections for the 9-month period ending March 31, 1946, totaled $147,781.05, against a budget estimate of $135,000.00 for the entire fiscal year.

The report on membership was brought up to date as of April 30, 1946, to show a total membership of 8,282, a net gain of 467 since June 30, 1945, after overcoming losses due to increased dues, reconversion, etc.

4. *Convention Activities.* A brief report on activities connected with the 50th Anniversary Convention was presented by the Secretary, calling attention to events of particular interest to members of the Executive Committee.

Safety and Hygiene. The Executive Committee discussed report prepared by Director L. C. Wilson, Chairman of a special Committee appointed to investigate the need for and feasability of a permanent Safety and Hygiene Program of A.F.A. Correspondence with the American Standards Association was presented indicating their desire for A.F.A. to sponsor codes of safety and hygiene for consideration by A.S.A. as American Standards. Following discussion a motion was made by Director Wallis, seconded by Vice-President Wood, unanimously carried

THAT the Executive Committee recommends to the Board of Directors that L. C. Wilson be appointed as Temporary Chairman of a Safety and Hygiene Special Committee to cooperate with the American Standards Association, for the purpose of sponsoring a Foundry Code for consideration by A.S.A. as an American Standard.

It was agreed that further discussions should be held at the earliest possible date in order to resolve what the A.F.A. position would be with regard to dust counts, a permanent Chairman, and similar problems.

6. *Sand Research.* The Secretary presented for consideration a request by the A.F.A. Fellow at Cornell, D. C. Williams, for an increase of $50.00 per month in the amount of the Fellowship, effective July 1, 1946. A brief background on the founding of the Fellowship was presented and the Executive Committee expressed the thought that Sand Research at Cornell should continue to be carefully supervised so as to obtain maximum good from this project. It was also felt that a question of an increase in the Fellowship should be passed upon by the Foundry Sand Research Committee. Therefore, a motion was made by Director Wilson, seconded by Director Smith and unanimously carried

THAT the Executive Committee recommends to the Board of Directors continuation of the A.F.A. Fellowship at Cornell University and that the amount of the Fellowship be increased by $50.00 per month for the fiscal year 1946-47, subject to approval by the Foundry Sand Research Committee.

7. *Technical Committee Organization.* The Secretary stated that S. C. Massari as Acting Technical Director was preparing a plan to accomplish some reorganization of our A.F.A. Technical Committees with a view to eliminating inactive Committees and delineating for each Committee its specific duties and responsibilities. The Executive Committee urged that such a plan be carried forward and presented for approval at an early date so as to become effective with the formation of Committees for the year 1946-47.

8. There being no further business, the meeting was adjourned.

Respectfully submitted,
WM. W. MALONEY, *Secretary.*

Minutes of Board of Directors Meeting

Hotel Statler, Cleveland—May 7, 1946

Presiding: President F. J. Walls
Present: President F. J. Walls
Vice-President S. V. Wood
Directors: R. M. Jacobs, Joseph Sully, Max Kuniansky, R. J. Teetor, W. B. Wallis, F. J. Dost, S. D. Russell, R. T. Rycroft, L. C. Wilson, G. K. Dreher, E. W. Horlebein, H. H. Judson, J. H. Smith, F. M. Wittlinger
Secretary Wm. W. Maloney
Treasurer C. E. Hoyt
Secretary Emeritus R. E. Kennedy
Also Present: Past Director B. D. Fuller, Member Advisory Board.

1. Minutes of the following meetings were read and approved:
(a) Executive Committee Meeting—January 15, 1946.
(b) Board of Directors Meeting—January 16, 1946.
(c) Executive Committee Meeting—March 2, 1946.

In connection with the Executive Committee Meeting held the morning of May 7, 1946, the Secretary presented for discussion at the proper time on the Board Agenda the recommendations of the Executive Committee.

2. *Reports of the Secretary.* The Secretary presented the following reports for the information of the Board of Directors.
(a) Income and Expense Budget Report for March 31, 1946.
(b) Statement of Assets as of March 31, 1946.
(c) Membership Report for May 2, 1946.
(c) Executive Committee Meeting—March 12, 1946.

3. *Convention Report.* A brief report on the 50th Anniversary Convention activities was presented by the Secretary, calling attention to items of particular interest to the Directors, including the Annual Business Meeting May 9, Alumni Dinner May 9, and the Annual A.F.A. Dinner May 10.

4. *Recommendations of Board of Awards.* The report of the Board of Awards on the awarding of A.F.A. medals and Honorary Life Memberships at the 1946 Convention was presented. In addition, the Board of Awards made certain recommendations for approval by the Board of Directors. It was moved by Director Wallis, seconded by Director Smith, and carried, that the recommendations of the Board of Awards be accepted and the Staff was instructed to comply.

5. *Safety and Hygiene.* Recommendation of the May 7, 1946, meeting of the Executive Committee was presented, as follows:

THAT L. C. Wilson be appointed as Temporary Chairman of a Safety and Hygiene Special Committee to cooperate with the American Standards Association, for the purpose of sponsoring a Foundry Code for consideration by A.S.A. as an American Standard.

It was stated that a real need for a Safety and Hygiene program in the interests of the entire foundry industry had been found, and that A.F.A. seemed to be the only group which could write and sponsor foundry codes which might receive full consideration by all interests involved. Director Wilson as Chairman of a Special Investigating Committee on the subject called attention to the formulation of state codes which have been considered inimical to foundry interests. He pointed out that A.F.A. had sponsored jointly with N.F.A. a foundry code which was approved by A.S.A. as an American Standard in 1932, but that this code is badly in need of revision. He also pointed out that all A.F.A. codes issued since that date have been sponsored and published solely by A.F.A. but that there seemed good reason for future codes being presented to the A.S.A. for issuance as American Standards.

On motion by Director Horlebein, seconded by Director Dreher, the recommendation of the Executive Committee for appointment of Director L. C. Wilson as Temporary Chairman of a Safety and Hygiene Special Committee, was approved.

6. *Sand Research.* Recommendation of the May 7, 1946, meeting of the Executive Committee, with reference to A.F.A. Sand Research at Cornell University, was read:

THAT continuation of the A.F.A. Fellowship at Cornell University and that the amount of the Fellowship be increased by $50.00 per month for the fiscal year 1946-47, subject to approval by the Foundry Sand Research Committee.

President Walls stated that no action by the Board should be taken until receipt of approval by the Foundry Sand Research Committee for the July Board Meeting.

7. *Chapter Publications.* Director Russell called attention to the issuance of a publication entitled "The Man at the Muller" by the Northern California Chapter and declared that the original plans of the Chapter for national distribution of the

publication had been scaled down to more or less local distribution at the suggestion of the National Office. The Board expressed the opinion that publications of this type by Chapters, bearing the name of the A.F.A., should properly be reviewed before publication but admitted that no national policy exists with regard to Chapter publications of any type. The Secretary pointed out that there is no by-laws provision with regard to publications issued by Chapters over the name of the Association.

No action was taken with regard to future publications that may be issued by the Chapters.

8. *1946 Chapter Chairman Conference.* Following discussion the Board approved the dates of July 24 and 25 for the 1946 Chapter Chairman Conference, with July 26 set aside for the Annual Meetings of the Board of Directors. It was agreed that the 1945-46 Board would hold its final meeting the morning of July 26, with the first meeting of the 1946-47 Board scheduled for that afternoon.

9. There being no further business, the meeting was adjourned.

Respectfully submitted,

WM. W. MALONEY, *Secretary.*

Minutes of Executive Committee of Directors Meeting

A.F.A. Headquarters, Chicago—June 28, 1946

Presiding: President F. J. Walls
Present: President F. J. Walls
Vice-President S. V. Wood
Directors: R. J. Teetor
L. C. Wilson
W. B. Wallis
Secretary Wm. W. Maloney

1. Minutes of the Executive Committee Meeting and Board of Directors Meetings, held in Cleveland May 7, 1946, were presented and approved.

2. *Financial Report.* A report of Income and Expense was submitted by the Secretary showing the budget for the fiscal year 1945-46 as compared with actual Income and Expense for the first 11 months, and estimated Income-Expense for June, 1946. This report showed the following condensed figures:

INCOME—		
Budget	$341,800.00	
Actual	426,533.00	
Excess		$84,733.00
EXPENSE—		
Budget	$341,800.00	
Actual	330,400.00	
Remainder	11,400.00	
SURPLUS		$96,133.00

3. *Membership Report.* A membership report was submitted by the Secretary for June, 1946, showing book membership on June 30, 1946, of 8539, compared with 7815 on June 30, 1945, a net gain of 724 members in spite of losses due to increased dues, reconversion, etc.

The report also showed total dues collections for the fiscal year ending June 30, 1946, totaled $148,819.71 as against $103,281.22 on June 30, 1945, a gain of $45,538.49. It was pointed out that the membership again is at an all-time high.

4. *Annual Lecture Committee.* The Secretary called attention to recommendations made to the Executive Committee at a meeting held in Chicago June 12, 1946, these recommendations and actions being as follows:

(a) The Committee recommended that the name "Foundation Lecture" be changed and instead be named after some prominent foundryman.

The Executive Committee looked favorably on this suggestion and requested that the Annual Lecture Committee submit two or three names for consideration at the Annual Board Meeting, July 26, 1946.

(b) The Lecture Committee requested that consideration be given setting up an emolument for the Annual Foundation Lecturer. The Executive Committee expressed the desire for more specific recommendations by the Lecture Committee, and tabled the subject until the Annual Board Meeting, July 26.

(c) The Lecture Committee recommended that more consideration be given the time of holding the Foundation Lecture at A.F.A. Conventions, especially during an Exhibit year.

The Executive Committee instructed the Secretary to give careful consideration to the matter in connection with future Convention programs.

(d) The Lecture Committee recommended that additional members be appointed to the Committee so as to represent the various foundry branches of A.F.A. This matter was referred to the President for appointments, since the Lecture Committee is an Appointive Committee of the Board.

5. AMERICAN FOUNDRYMAN. A report on advertising revenue of AMERICAN FOUNDRYMAN was presented by the Secretary showing a gross revenue of $112,930.50 for the 14-month period May, 1945-June, 1946, inclusive, representing a net (after agency commissions) of $99,539.55.

For the fiscal year July, 1945-June, 1946, gross advertising revenue totaled $94,629.25, against the budget of $75,000.00 gross revenue. Net revenue, after agency commissions, for the fiscal year totaled $83,247.31.

Attention was called to the growing acceptance of AMERICAN FOUNDRYMAN by the membership and by advertisers, and the Executive Committee expressed approval of continuation of the editorial policy, but with the suggestion that more attention be given the publication of practical foundry articles.

6. *Committee Organization, 1946 Technical Program, A.F.A. Publication Policy, A.F.A. Sponsored Research.* All of these subjects were considered together following presentation of complete plans and recommendations by Acting Technical Director S. C. Massari. The entire presentation was unanimously approved, and it was the consensus of opinion that the entire plan should be kept intact for the information of any new Officer or Director of the Association. Mr. Massari was strongly commended for the presentation.

7. *Meeting of Foundry Groups.* President Walls read a letter received from the President of the National Founders' Association requesting attendance by the A.F.A. President and members of the Executive Committee at a meeting in Cleveland July 10 for the purpose of considering formation of an "Overall Foundry Council." Following discussion, the Executive Committee agreed that A.F.A. should attend this meeting, President Walls to be accompanied by Vice-President Wood and Directors Wallis and Wilson.

8. *Chapter Possibilities.* Various possibilities for the formation of new Chapters were discussed. The Secretary was instructed to continuously cooperate with any sound move toward establishment of new Chapters in foundry centers.

9. *Apprentice Contest Prizes.* The Secretary called attention to the fact that the cash awards of $30.00, $20.00 and $10.00 now offered in the four Divisions of the National Apprentice Contests have not been considered consistent with the honor of winning a national contest. It was pointed out that several Chapters offered prizes for their local contests in amounts equal to those awarded in the national contest and it was sug-

greed that the Executive Committee give consideration to increasing the prize awards to $100.00 for first prize, $50.00 second prize, $25.00 third prize.

The Executive Committee expressed approval of the reasons involved for increasing the awards and instructed the Secretary to investigate the availability of funds for making such increased awards, for presentation of a specific recommendation at the Annual Board Meeting, July 26.

16. *American Standards Association.* The Secretary presented a letter from the American Standards Association requesting A.F.A. to appoint a representative to participate in developing Standards for protective occupational footwear. In view of the fact that other matters of cooperation with A.S.A. were pending the Secretary was instructed to withhold appointment of a representative as requested until the complete matter has been resolved.

11. *Retirement Compensations.* Attention was called to the fact that action by the Board of Directors at its July, 1945, Annual Meeting, with reference to retirement pay for Secretary Emeritus R. E. Kennedy, could be taken only for the fiscal year ending July 1, 1946. Accordingly a motion was made by Vice-President Wood, seconded by Director R. J. Teetor, and unanimously carried

THAT the Board resolutions adopted July 18, 1945, with reference to retirement compensation of Secretary Emeritus R. E. Kennedy be extended the fiscal year ending June 30, 1947.

Recommended to the Board of Directors for approval.

Attention was called to the fact that resolutions by the Board of Directors adopted at the Annual Board Meeting, July 18, 1945, set forth the retirement of Treasurer C. E. Hoyt on August 1, 1946, with retirement compensation thereafter. In view of the necessity of making these actions official for the fiscal year 1946-47 under a changing Board, it was moved by Vice-President Wood, seconded by Director R. J. Teetor, and carried

THAT the Board resolutions adopted July 18, 1945, with reference to retirement compensation for Treasurer C. E. Hoyt, commencing August 1, 1946, be authorized for the fiscal year ending June 30, 1947.

Recommended to the Board of Directors for approval.

12. *Bank Requirements.* In view of the retirement of C. E. Hoyt as Treasurer on August 1, 1946, and the resignation of Assistant Treasurer Jane Reininga on March 15, 1946, the Secretary pointed out the necessity for action authorizing staff signatures necessary for the signing of Association checks. He also pointed out the advisability of having three authorized Staff signatures as in the past. On motion by Director Wallis, seconded by Vice-President Wood, the Executive Committee recommended

THAT the offices of Secretary and Treasurer be combined for the fiscal year ending June 30, 1947, and that Secretary Wm. W. Maloney be elected Secretary-Treasurer for this period.

Recommended to the Board of Directors for concurrence.

Some discussion took place regarding the necessity for having two authorized signatures on Association checks and the consensus of opinion was that the presence of either one or two signatures would in no way alter the degree of responsibility or authentication necessary to the accounting of Association funds. Accordingly on motion by Director Wilson, seconded by Vice-President Wood, the Executive Committee voted

THAT after August 1, 1946, Association checks may be issued bearing any one of the following authorized signatures: President, Vice-President, Secretary-Treasurer, Technical Director, Assistant Treasurer.

Recommended to the Board of Directors for concurrence.

13. *Staff Considerations.* Authorization was given Secretary Maloney to make the necessary arrangements for signatures, and for arranging bond in the proper amount for those Staff members authorized to sign Association checks or to handle Association moneys.

The Secretary informed the Executive Committee that S. C. Massari, who joined the staff in February, 1946, on a temporary basis to July 1, had agreed to remain as a Staff member. A motion then was made by Vice-President Wood, seconded by Director Wallis and carried

THAT S. C. Massari be retained, at compensation recommended by the Finance Committee, as Technical Director for the fiscal year July 1, 1946-June 30, 1947.

Recommended to the Board of Directors for approval.

The Secretary informed the Executive Committee that John P. Mullen who joined the staff in February, 1946, on a temporary basis to July 1, had agreed to remain as a Staff member. A motion then was made by Vice-President Wood, seconded by Director Wallis

THAT John P. Mullen be retained in the capacity of Asst. Treasurer for the fiscal year July 1, 1946-June 30, 1947, at compensation recommended by the Finance Committee.

Recommended to the Board of Directors for approval.

The Secretary called attention to the work of Therese Koeller who has been in charge of AMERICAN FOUNDRYMAN advertising since May, 1945, and recommended that recognition be given her work in this capacity.

President Walls and Vice-President Wood reported the results of conversations held with N. F. Hindle, Director of the Technical Development Program, whose serious illness, incurred in February, 1946, had made it impossible for him to return to staff activities. President Walls stated that Mr. Hindle still was unable to say when his recovery might be sufficient to permit his return to active status but that the Association was primarily interested in Mr. Hindle's complete recovery at the earliest possible date. It was pointed out, however, that the Association would have to go forward in its activities, taking actions deemed necessary to that end. Accordingly, it was moved by Director Wallis, seconded by Director Teetor and carried

THAT N. F. Hindle be satisfactorily compensated, for such period as he may be still convalescent during the fiscal year July 1, 1946, through June 30, 1947, in the hopes that his complete recovery may be accomplished during this period of time.

Recommended to the Board of Directors for approval.

14. There being no further business, the meeting was adjourned.

Respectfully submitted,

WM. W. MALONEY, *Secretary.*

Minutes of Annual Meeting 1945-46 Board of Directors

Hotel Stevens, Chicago—July 26, 1946

Presiding: President F. J. Walls
Present: President F. J. Walls
Vice-President S. V. Wood
Directors: R. M. Jacobs, Joseph Sully, Max Kuniansky, R. J. Teetor, W. B. Wallis, F. J. Dost, S. D. Russell, R. T. Rycroft, L. C. Wilson, E. W. Horlebein, H. H. Judson, J. H. Smith, F. M. Wittlinger
Secretary Wm. W. Maloney
Treasurer C. E. Hoyt
Also Present: Directors Elect B. L. Simpson, H. A. Deane, S. C. Wasson, J. E. Kolb.

1. Minutes of the following meetings were read and approved:
(a) Executive Committee Meeting, Cleveland, May 7, 1946.
(b) Board of Directors Meeting, Cleveland, May 7, 1946.
(c) Executive Committee Meeting, Chicago, June 28, 1946.

2. *Finance Report.* A report of the Finance Committee was offered by Chairman R. J. Teetor who called attention to the auditor's official report of the Association accounts for the fiscal year ended June 30, 1946. Particular attention was directed to a Consolidated Fund Statement, in the nature of a Net Worth Statement, showing that the Net Worth of the Association on June 30, 1946, was $341,188.90.

Attention was called to the auditor's Statement of Income and Expense for the General activities of A.F.A., comparing actual figures with budgeted figures. This statement showed actual Income of $417,226.32, against the budgeted Income of $341,800; actual Expense $323,792.88, against budgeted Expense of $323,400.00. Budgeted "Surplus" of Income over Expense, $18,400.00; actual "Surplus" of Income over Expense, $93,433.44.

It was pointed out that $32,800.00 of Income is shown as Income Deferred for 1946-47 operations, leaving a balance of $60,633.44 deferred to the Fund Principal of the Association. Attention was called to the fact that the various Reserve Funds of the Association showed the following balances as of June 30, 1946: Technical Development Program Fund, $104,064.40; Cupola Research Fund, $31,403.85; General Reserve Fund, $51,069.06; Awards Fund, $29,231.55.

On motion by Director Horlebein, seconded by Director Wilson, the Report of the Finance Committee was accepted.

3. *Membership Report.* Vice-President Wood as Chairman of the National Membership Committee, reported the membership of the Association on June 30, 1946, as 8,530, against 7,815 at the same time the previous year, a net gain of 724 or 9.26%. Chairman Wood called attention to the fact that this net gain had been accomplished in spite of losses due to the end of the War and an increase in the Sustaining and Company membership dues effective July 1, 1945.

In commenting on the growth in the Association membership, the Secretary pointed out that in spite of a loss of 56 Sustaining members, and a loss of 109 Company members, Dues Income increased by $32,400.00 over the previous year. It was pointed out that such a condition was anticipated prior to action being taken by the Board toward increasing Company and Sustaining member dues.

On motion by Director Kuniansky, seconded by Director Wallis, Report of the Membership Committee was accepted.

4. *Report of the Secretary.* The Secretary submitted a written report covering Association activities during 1945-46, which was filed without reading, for the record.

5. *Report of Chapter Contacts Committee.* Vice-President Wood, Chairman of the Chapter Contacts Committee, reported on visits made to various Chapters by National Officers and Directors and stressed the fact that, while such visits have been eminently well received, even closer contacts with A.F.A. Chapters are necessary in the future. He reiterated a suggestion made at a previous Board Meeting that only a full-time Chapter contact man could accomplish this work with complete success and maximum results.

6. *Report of Committee on* AMERICAN FOUNDRYMAN *Policy.* Director Wallis, Chairman, reported on advertising revenues and costs during the year 1945-46. Quoting from the auditor's annual report he pointed out Gross Advertising Income of $94,278.35 as against Production Cost (including printing, postage, agency commissions and promotional work) of $87,792.89, in spite of original prophesies that the magazine could not be operated at a profit during its first year.

He also pointed out that, with TRANSACTIONS material now being printed in AMERICAN FOUNDRYMAN instead of a Quarterly, the resultant saving in production cost must be shown on the credit side of AMERICAN FOUNDRYMAN; since the cost of the Quarterly in its last previous year approximated $14,000.00, he estimated that the spread between Gross Advertising Revenue and Gross Production Cost during the past year was considerably greater than the figures indicated.

The Secretary pointed out that advertising in AMERICAN FOUNDRYMAN is being sold solely on the basis of returns for the advertiser and not as support for A.F.A. activities in general. He stated that effort would be made consistently to improve the editorial content in accordance with expressed desires of the membership. He also stated that the magazine would require two or three years of assistance to develop its possibilities beyond the results shown at the end of the first year of operation with the inclusion of regular monthly advertising.

7. *Report of the Management Committee.* Director Rycroft, as Chairman, reported on Management Meetings held with several local Chapters during the past year and urged that such meetings be continued in the future. It was stated that several Chapters have already evidenced the desire for Management Meetings in their Chapter areas so as to better acquaint top management of organizations with the essential work of the Association.

8. *Recommendations of the Executive Committee.* Action was taken on recommendations of the Executive Committee as follows:

(a) The recommendation of the 1946 Board of Awards was discussed, that consideration be given some form of recognition for companies whose willingness to contribute freely to the literature of the industry has been outstanding. Motion was made by Director Wilson and seconded by Director Wallis and carried, that the recommendation of the Board of Awards be referred to the 1946-47 Board of Directors, with the suggestion that a Board Committee be appointed to study the matter and report at a future meeting.

(b) S. C. Massari, as Acting Technical Director of A.F.A., presented a program for (1) *Technical Committee Organization,* (2) a *1947 Technical Convention Program,* and (3) *A.F.A.-Sponsored Research.* He outlined all three of these matters and their estimated effect toward obtaining greater Committee activity and more frequent reports, improved quality of Convention papers and early preprinting of same, and the improved method of originating, conducting and carrying through A.F.A. Research Programs of a fundamental nature.

In connection with the Research Program it was agreed that the research facilities of A.F.A. should be extended to the Patternmaking Division as well as to the Steel, Gray Iron, Malleable Iron, Brass and Bronze, Aluminum and Magnesium Divisions, and the Foundry Sand Research Project.

On motion by Director Kuniansky, seconded by Director Horlebein, all three proposals were approved in blanket and recommended to 1946-47 Board for adoption.

(c) Mr. Masarri also presented an outline for a definite A.F.A. *Publication Policy,* intended to increase the prestige of presenting papers at A.F.A. Conventions, and to protect the interests of both A.F.A. and the authors of technical papers.

On motion by Director Smith, seconded by Director Kuniansky, the Publication Policy was approved and recommended to 1946-47 Board for adoption.

(d) *Bank Resolutions.* In accordance with the requirements of A.F.A. depository banks for resolutions pertaining to authorized signatures for the deposit and withdrawal of funds, motion was made by Director Wittlinger, seconded by Director Horlebein and carried, authorizing the execution of such resolutions.

Following discussion, the Board continued previous Association practice by approving the suggestion that checks issued by A.F.A. should bear two authorized signatures, and that this provision be incorporated in the necessary bank resolutions, subject to approval of 1946-47 Board.

The Secretary called attention to the desirability of changing the names of two Association bank accounts as follows:

(1) Changing the Harris Trust Company checking account heretofore entitled "Exhibit Sinking Fund, American Foundrymen's Association" to *"General Reserve Fund, American Foundrymen's Association."*

(2) Changing the Northern Trust Company checking account heretofore entitled "Technical Development Program, American Foundrymen's Association" to *"Cupola Research Fund, American Foundrymen's Association,"* in view of the intention to retain uninvested Technical Development Program funds hereafter in a savings account rather than in a checking account. It was pointed out that the latter change would enable at any time a financial separation between all Technical Development Program and all Cupola Research funds so that each fund may be accounted for separately.

On motion by Director Wallis, seconded by Director Judson, authorization was given for making the above changes in Association bank accounts, subject to approval by 1946-47 Board.

(e) In Executive Session, the 1945-46 Board of Directors approved, and recommended to the 1946-47 Board for appropriate action, resolutions made by the Executive Committee in session June 28, 1946, on the following matters:

Proposed compensation for Treasurer C. E. Hoyt as expressed in resolutions of the Board of Directors on June 19, 1945;

Extension of retirement compensation for Secretary Emeritus R. E. Kennedy in accordance with actions by the Board of Directors on July 19, 1945;

Compensation of N. F. Hindle beyond July 1, 1946.

9. *Chapter Possibilities.* The Secretary discussed briefly possibilities for organization of new Chapters in several areas including Seattle and Central Michigan. It was stated that a Steering Committee has already been formed for a Seattle Chapter and that the Committee expects to hold an organization meeting early this fall.

The Secretary reported conversations indicating interest in forming a Central Michigan Chapter and stated that this matter was being pursued in conjunction with surrounding Chapters already in existence. Interest in forming a Western Massachusetts Chapter, developed during the past year, was also reported.

10. *Chapter Chairman Conference.* Director Kuniansky, as Chairman of the 1946 Chapter Chairman Conference, held in Chicago, July 24-25, reported a highly successful meeting and stated that a transcript of the Conference would be sent to all who attended that meeting as well as to all National Officers and Directors. It was stated that every Chapter had been officially represented with the exception of Mexico, the first time there had been such complete attendance.

11. *Foundry Sand Research.* A previous recommendation of the Executive Committee was read pertaining to continuation of the A.F.A. Sand Research Fellowship at Cornell University. The Secretary reported that a complete review of the Fellowship was being prepared and would be presented to the Executive Committee at an early date for approval and extension. He stated that Technical Director Massari would perform the review but that Executive Committee authority would be requested for any interim funds necessary to carrying on the work.

12. *Apprentice Contest Prize Awards.* Acting on recommendations of the Executive Committee, on motion by Director Horlebein, seconded by Director Wilson, the Board approved unanimously that the Apprentice prizes for 1947 be increased from $30.00 to $100.00 for first prize, $20.00 to $50.00 for second prize, and $10.00 to $25.00 for third prize. The Board recommended that these amounts be approved for the coming year through action by the 1946-47 Board.

13. *Cooperation With Trade Associations.* Vice-President Wood reported on a meeting held in Cleveland July 10 between national officers of A.F.A., Gray Iron Founders' Society, Malleable Founders' Society, Steel Founders' Society, Non-Ferrous Founders' Society and National Founders' Association. He reported that the meeting had not accomplished anything definitely save to indicate a unanimous feeling that such an overall Committee was desirable and might be effected.

He stated that the Presidents of the various Associations were scheduled to meet again in Cleveland August 14 for more detailed discussion of ways and means, and that this group would then report back for further and progressive action following approval by their respective societies of any recommendations developed at the August meeting.

Following discussion, the 1945-46 Board recommended to the new Board that A.F.A. participation in an overall foundry group be explored further, with possible commitments subject to approval by the Executive Committee. It was the sense of the meeting that A.F.A. should cooperate in this project to the best of its ability, within its fundamental objects as a Technical Society.

14. *Joint Educational Committee.* President Walls presented a letter from F. G. Sefing as Chairman of the A.F.A. Committee on Cooperation with Engineering Schools and Colleges, reporting a meeting held in Cleveland July 12 by a Joint Educational Committee for the ferrous foundry industry. The report stated the intention of raising some $60,000.00 each for scholarships at five universities and colleges, A.F.A. participation being requested.

The request was referred to the Secretary and Technical Director for recommendations, such recommendations to be presented before the September meeting of the Executive Committee.

15. There being no further business, the meeting was adjourned.

Respectfully submitted,

Wm. W. Maloney, *Secretary.*

TEST BARS FOR 85-5-5-5 ALLOY
Their Design
And Some Factors Affecting Their Properties

By G. H. Clamer[1]
Ajax Metal Co., Philadelphia

FOURTH A.F.A. FOUNDATION LECTURE — 1946

I—Introduction

Test bars are an important subject in foundry operation because hundreds of thousands of them are cast and tested each year in nonferrous foundries at a tremendous cost of effort, time, and money. Furthermore, both the foundry and the casting purchaser rely upon them as a barometer of casting quality. The one point concerning test bars on which there is unanimity of opinion is that they constitute a considerable problem to both the foundrymen and to the casting purchaser.

There are three main aspects of this problem. First, the application of the best known procedures in order that the test bars may at least partially warrant the importance attached to them. Second, granting that the best practices are, or will be, utilized, the limitation of the test bars as a barometer of casting quality should be established. Third, a continued effort should be made to develop and standardize test-bar practices further, in order to make them a more reliable barometer of casting quality.

A corollary of the above statement of the problem would involve the elimination of the poor test-bar practices currently used by some foundries which unnecessarily cast a shadow over many of their foundry operations.

This paper deals with all three phases of the problem, as well as some effects of melting practice. Investigations to date have resulted in a better understanding of test bars, the establishment of the effect of some melting practices on test-bar properties, and the attainment of a clearer conception of proper test-bar practices. However, the millennium whereby a single test-bar value becomes a reliable barometer of casting quality has definitely not been, and may never be, reached. It will be developed later that, in any case, the test-bar can be a barometer of casting quality only insofar as the properties or serviceability of the castings are dependent upon melt quality.

If the test-bar design, gating practice, and pouring temperature, are rigidly standardized, the correlation which exists between melt quality and test-bar properties depends almost entirely upon the effects of the melt quality on the occurrence of a certain type of unsoundness in the test-bar castings. This unsoundness has been variously referred to in the literature as shrinkage, porosity, shrinkage porosity, gas porosity, incipient shrinkage, sponginess, unsoundness, microshrinkage, and microporosity.

Accordingly, no paper of this type would be complete without some description of this defect and its causes. Fortunately for the general understanding of this defect, it is not confined to 85-5-5-5 alloy. On the contrary, it is common in all known casting alloys of the solid-solution type, namely, alloys which solidify and form substantially a single-phase alloy. Several aluminum-base alloys and practically all commercial magnesium-base casting alloys come under this description, and all are subject to various degrees of this defect.

Among light-alloy-foundry metallurgists, this unsoundness is known as microporosity or microshrinkage, but the former term is preferred because it conveys no erroneous idea in respect to the causes of the defect. This defect, which in this paper hereafter will be called microporosity, has been described in many recent papers in the United States[5, 6, 7, 8, 9] and in England[10].

The defect known as microporosity results from a fairly complex mechanism involving both shrinkage and the evolution of gas from the solidifying melts. It is confined largely to the solid-solution type of alloy of which 85-5-5-5 is one, because these alloys solidify over a considerable range, forming substantially a single phase with the production of little or no eutectic liquid, aside from the lead.

As a result, the solidifying casting forms a wide mushy zone where the section is partially solid and through which it becomes increasingly difficult for liquid to feed and compensate for the shrinkage. As a further result, small, more or less angular, connecting voids surrounding the grains are formed, and since no liquid can proceed satisfactorily through this mushy zone, these voids remain as the defect known as microporosity.

Obviously, if there are even small traces of gas evolved during solidification, this gas will fill these microvoids, making them still more difficult to feed. In the absence of gas, these voids would be a near vacuum and would tend to draw metal into them as a result of the external atmospheric pressure, tending to produce very sound castings.

If gas is evolved during solidification, the gas-filled voids are very difficult to feed, and as a result, the greater the gas content of the melt, the more accentuated the microporosity becomes, and the lower the test-bar properties. It is this fact that makes it possible to have some correlation between the melt quality on

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the one hand and the test-bar properties on the other.

II—MELT QUALITY AND ITS MEASUREMENT

The concept of melt quality is an important one to this paper because the contents herein deal almost entirely with factors affecting melt quality, and methods of measuring it. The word "quality" implies the possession of certain desired distinctive properties. As with other melts of casting alloys, the distinctive property which a melt of 85-5-5-5 alloy should have is the production of maximum serviceability in the castings.

The problem is complicated by the fact that the same melt quality may not be optimum for castings of various designs and purposes. Thus, the desired distinctive properties of a melt for a casting in which maximum leak tightness is desired may be somewhat different than those for castings designed for other service.

To be more explicit, the term "melt quality" refers to the quality of the melt in the pot and may be used synonymously with "melt composition"; covering gases, suspended dross, etc., thus being a different matter from composition as determined by ordinary chemical or even spectrographic analyses. The highest melt quality then becomes the "composition" which will produce, all other factors the same, castings of maximum serviceability.

Casting serviceability is difficult to evaluate, because it varies so completely with each design and with each type of service. But the casting characteristics most universally sought are high mechanical properties. The connection is obvious, and for practical reasons, when carrying out the experimental work described here, the mechanical properties of test bars have been used as a measure of melt quality.

However, the mechanical properties of castings, including test bars, depend on many factors in addition to the melt quality. Furthermore, the mechanical properties of the test bars are not necessarily indicative of the mechanical properties of any other castings poured from the same melt, and since this is now generally accepted as a fact, the problem is somewhat clarified.

In addition to to melt quality, the mechanical properties of the test bars, as well as of other castings, are affected by such factors as the type of gating, the degree of feeding provided, the sand used, the pouring temperature, and many other foundry variables that are not so readily controlled. It is axiomatic that all these factors should be standardized and controlled insofar as it is possible to do so when casting test bars. The melt quality then becomes the major factor determining test-bar properties. This is an important matter if the test bars are to have maximum significance as a means of quality control in the foundry.

However, it must be admitted in the beginning that, because some of the enumerated factors are difficult to control completely, the mechanical properties of the test bars are not an unqualified measure of melt quality, but some test-bar designs may be better than others. Accordingly, the relative merits of various designs of test bars will, therefore, be given some attention.

It has been mentioned that melt quality is synonymous with melt "composition." One of the most readily made checks on melt composition is the ordinary chemical analysis; the concentration of the principal alloying elements and many of the common impurities can be readily determined, and no further consideration of this phase of the evaluation of melt quality need be given.

However, any satisfactory method of measuring melt quality should primarily be reflective of the presence of minor, but potent, constituents which cannot now be readily evaluated by ordinary chemical analysis. While undissolved foreign particles in the melt, such as suspended dross, may occasionally be a factor, the most important minor constituents are those which are in solution in the melt, and of these, the ones most difficult to control are the dissolved gases or compounds containing them.

Although chemical analysis lends itself very readily to the determination of various other impurities in the melt, it is not, at the present stage of development, suitable for the quantitative determination of dissolved gases in the melt. It has become necessary, then, for the foundryman, the purchaser of the castings, or other agencies, to try to devise other means of measuring melt quality. Fundamentally, this resolves itself into a means of measuring the amount of dissolved gas or gas formers by measuring their effect upon the mechanical properties of a suitable test casting.

Main Factors Determining Melt Quality

If it is now agreed that melt quality and melt "composition" are synonymous, there are but three factors which determine melt quality.

1. The composition of all the raw materials placed in the charge, including flux additions, if any.

Table 1

AMOUNT OF GAS OR GAS FORMER REQUIRED TO PRECIPITATE BY TYPE 1 GAS EVOLUTION OR REACT IN TYPE 2 GAS EVOLUTION TO PRODUCE 1 PER CENT VOIDS AT 1900° F. IN A SOLIDIFYING COPPER-BASE ALLOY HAVING A DENSITY OF 8.6 AT 1900° F.[c]

Gas Evolution, Type	*Gas or Gas Former*	*Gas or Gas Former Required* — *c.c. per 100 grams*[a]	*Gas or Gas Former Required* — *Per Cent by Weight*
1	H_2	0.0242	0.0000022
1	CO	0.0242	0.0000302
1	CO_2	0.0242	0.0000475
1	SO_2	0.0242	0.0000692
1	N_2	0.0242	0.0000302
2[b]	H_2—Reaction (A)	0.0242	0.0000022
2	S—Reaction (B)	——	0.0000346
2	C—Reaction (C)	——	0.0000130
2	CO—Reaction (D)	0.0242	0.0000302

[a] Measured at 32° F. and existing atmospheric pressure.

[b] Also, 0.0121 c.c. of O_2 or 0.0000173 per cent oxygen by weight is required for Reactions (A), (C), and (D); 0.0242 c.c. or 0.0000346 per cent of O_2 is required for Reaction (B) to form a precipitated gas causing 1 per cent voids.

[c] It is assumed that the gas in the voids is at atmospheric pressure. Actually the pressure might be slightly more, in which case proportionately more gas would be required to produce 1 per cent voids than the quantities listed.

2. The contamination of the charge during melting and while superheating it to the pouring temperature. Contamination while transferring and pouring the melt into the mold cavity, including possible contamination of the melt in the mold cavity itself, may also be factors affecting melt quality after the melt leaves the pot, but these should be minimized by the adoption of sound, standardized test-bar practices in the shop.
3. The treatment to which the melt in the pot is subjected. This treatment is normally designed to remove partially or completely the harmful impurities, particularly the dissolved gases.

Gas Evolution

Since nongaseous impurities lend themselves to detection by ordinary chemical analysis, many of them by spectrographic methods, the primary concern then is with the content of gas or gas-forming constituents in the melt. In general, unsoundness from dissolved gas may result from two types of gas evolution. These are:

Type 1. Gas evolved during solidification as a result of decreased solubility.

Type 2. Gas evolved as the result of a chemical reaction in the melt.

The gases most likely to be evolved by the Type 1 gas evolution are hydrogen, carbon monoxide, carbon dioxide, sulphur dioxide, and nitrogen. It is probable that only the first of these, namely hydrogen, is an important source of unsoundness. The most likely reactions forming Type 2 gas evolution are as follows:

A. $2H + O$ (or an oxide) $\rightleftarrows H_2O$
B. S (or a sulphide) $+ 2O$ (or an oxide) $\rightleftarrows SO_2$
C. $C + O$ (or an oxide) $\rightleftarrows CO$
D. $CO + O$ (or an oxide) $\rightleftarrows CO_2$

All of these reacting constituents may be dissolved in the melt where they are in equilibrium until the melt solidifies. When solidification occurs, any one or more of these reactions may proceed to the right, evolving the gaseous compound. Any one or more of these reactions are possible for the following reasons: Before solidification begins, the reacting constituents are, or may be, in equilibrium concentrations, and hence the reactions proceed neither to the right nor to the left.

However, when solidification begins, the gases or gas formers are concentrated in the remaining liquid, the equilibrium is destroyed, and the reactants being in greater-than-equilibrium concentrations cause the reaction to proceed to the right, evolving a gaseous compound during at least the latter part of the solidification. Of these four reactions, "A" is the most likely source of unsoundness, while "C" and/or "D" are possibilities. The amounts of these gases or gas formers to produce 1 per cent of voids in a solidifying casting are listed in Table 1.

Very unsound castings generally contain no more than 4 per cent voids. It is immediately evident that the quantities of impurities involved are very small, and it readily explains one reason for the great difficulty which would be encountered if ordinary chemical analysis were employed to detect and measure the content of these gases and gas formers.

Sources of Gas Contamination

As previously noted, the concern is, primarily, with the dissolved gases or gas formers in the melt; the sources of these gases or gas formers are (1) the raw materials, and (2) contamination during melting, heating, holding, transferring, and pouring, including the interval the melt is molten in the mold cavity.

Contamination During Melting. Contamination of the melt with gases or gas formers in the melting room is inevitable, and the degree of contamination will depend largely upon the melting practice and conditions which surround the melting operation. This contamination during melting may be from the furnace atmosphere, fluxes, crucible or furnace linings, from the adsorbed and chemically combined gases on the surface of the metal charge, and from various tools and implements inserted into the melt.

The possible contamination during the transfer and pour is, primarily, from the atmosphere and from the lining of the pouring ladle. The contamination, if any, of the metal while it is molten in the mold cavity will, of course, depend upon the mold material, moisture content, mold permeability, and many other similar factors.

The amount of gas in the melt will also depend upon the success of the efforts made to eliminate it through so-called refining or purifying operations. When melting copper-base alloys, good furnace-room practice generally dictates that a melting procedure be used that will reduce gas absorption while melting and holding.

In a sense then, refining or purifying goes along simultaneously with the melting and holding operation. In any event, it will be apparent from the experimental data submitted that very wide variations in melt quality can be obtained from the same lot of metal, depending upon the melting technique and melting conditions applied.

Raw Materials. Fortunately, the raw materials are not serious sources of gas because the dissolved gas in the melt before the ingot is cast will be precipitated during the solidification of the ingot. Only the very minor amounts, which were soluble in the solid state, would remain in solution in the ingot. Since the solid solubility of gases in copper-base alloys is relatively low, the quantity of the gas remaining in solution in the metal is small, even though the gas content of the melt before the ingot was cast may have been fairly high.

There is, of course, the possibility that some of the precipitated gas may be trapped in the ingots as the metal solidifies. Such trapped gases in cavities in the ingots are of little significance in comparison to the adsorbed gases on the surface of the metal. These adsorbed gases on the surface of the raw materials are inevitable because their source is the atmosphere, and the amount of these adsorbed or surface gases will largely depend upon the atmospheric exposure to which the metal has been subjected after the ingots were cast.

The greater proportion of these surface gases is removed during the heating operation prior to melting. It is well known that these surface gases are important sources of gas when melting aluminum and magnesium alloys[5]. However, the role surface gases play when melting copper-base alloys is relatively unknown.

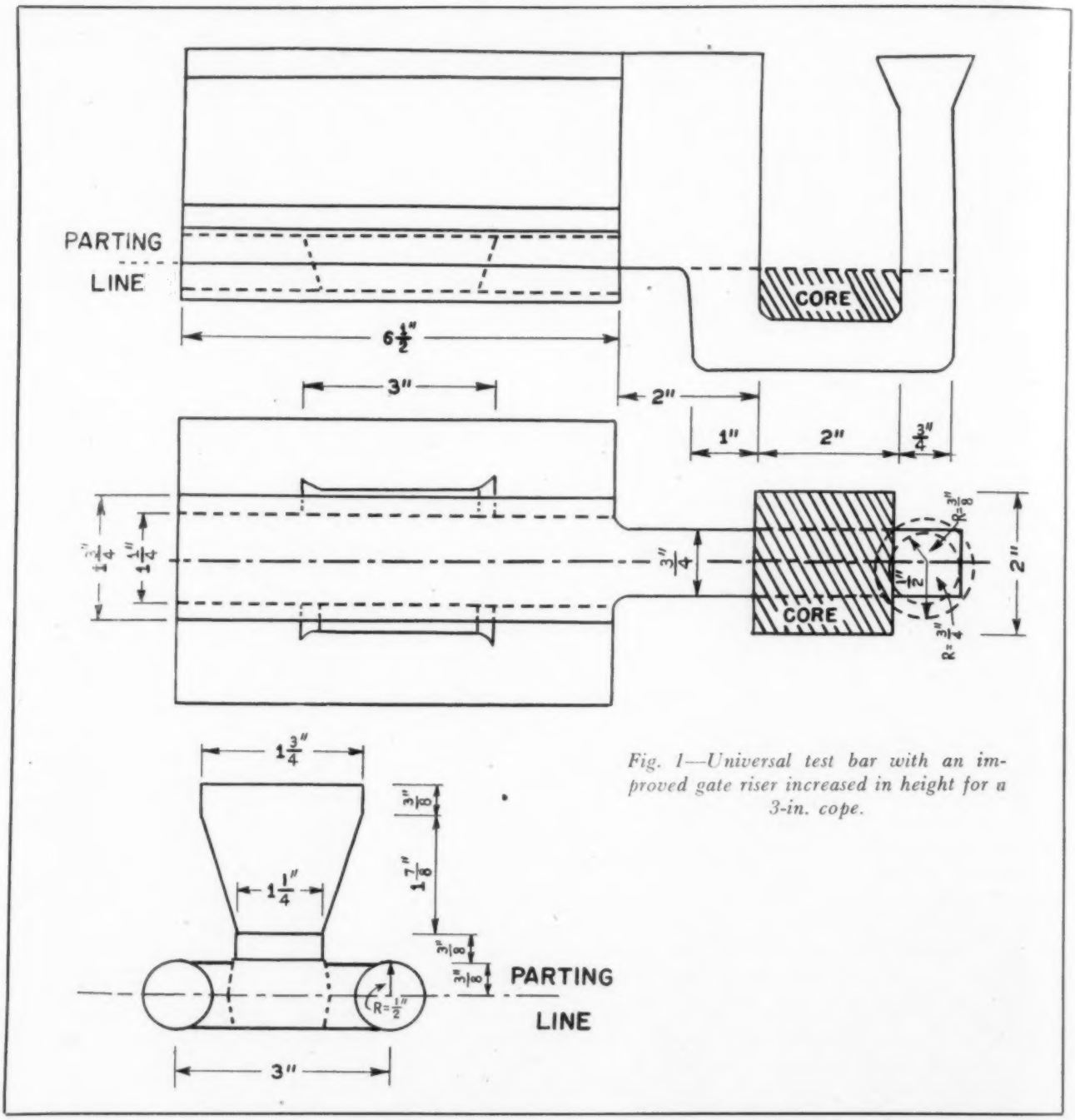


Fig. 1—Universal test bar with an improved gate riser increased in height for a 3-in. cope.

III—Test Bars Employed to Measure the Melt Quality

It may be concluded that melt quality is synonymous with melt composition, and that the constituents in the melt not readily evaluated by chemical analysis are the gases and gas formers. A means of measuring this more subtle aspect of melt quality is required; accordingly, the suitability of various test bars for this purpose has been explored.

The mechanical properties of test bars, like other castings, are dependent upon (1) melt quality, and (2) the foundry practice used to make the casting; namely, the manner of pouring, gating, risering, chilling, the molding technique, and sand used. It is obvious that if the test-bar properties are to reflect melt quality, foundry defects such as those arising from turbulence during the pour, dirty metal and dross formed during the pouring operations, and shrinkage from poor feeding, must be avoided so that the test-bar properties will be dependent upon, and reflective of, those defects which are dependent upon melt quality.

Thus, the test bar should be so designed as to be sensitive to melt defects; that is, the tensile properties should reflect defects arising from poor melt quality such as (1) gas content, (2) inherent tendency to produce coarse grain where this is an important factor, and (3) dross films in the melt. In addition, the test bars specified should be easy to mold and cheap to make. It is desirable also that they be cast in a separate mold, or, if cast in the same mold with another casting, they should be attached to the gate system, but not to the casting, as this latter practice is now generally re-

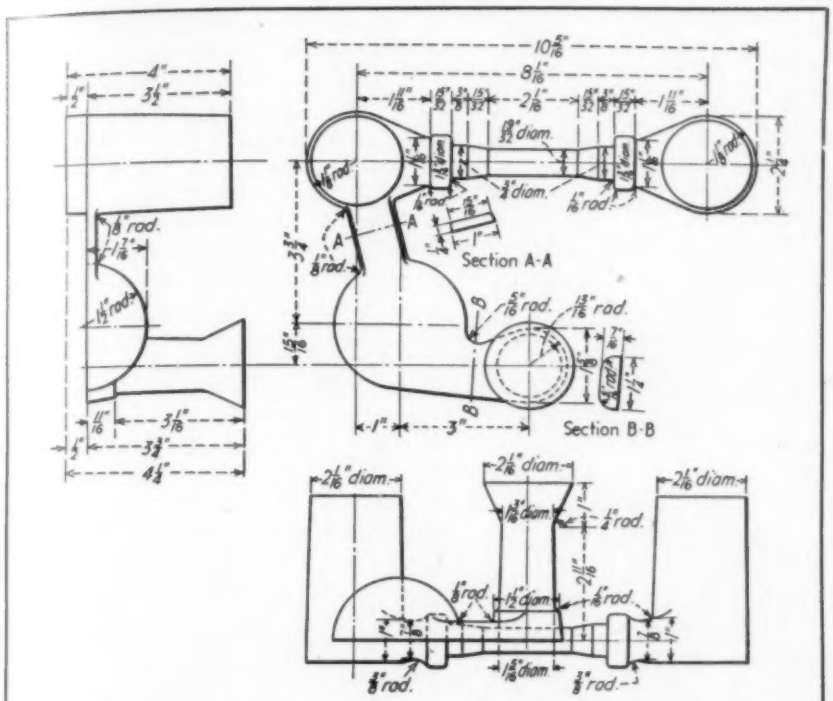


Fig. 2—L.T.B.-1 type test bar was modified without changing the dimensions of the gate to form two cavities per mold. The risers were reduced to 3 in. to permit the use of a 3-in. cope.

garded as being hazardous.

Test-Bar Design

Eleven different test-bar designs have been investigated, and these are illustrated by Figs. 1 to 9, inclusive. Some of these designs have been utilized throughout, while others were discontinued after preliminary investigations.

Figure 1 represents the so-called Universal test bar which, excepting the gate, was suggested by W. A. Baker[1]. It was found that the vertical slot gate originally recommended produced very unsound test bars, and it was soon abandoned in favor of the gate illustrated by Fig. 1.

Figure 2 illustrates the L.T.B.-1 test bar[3]. This is a single-cavity mold later changed to a two-cavity mold without, in any way, altering the dimensions of the casting or gates, but the original diameter of the sprue was retained on the double cavity mold. The double cavity mold is referred to as the L.T.B.-2.

Figure 3 illustrates a Webbert-type test bar[11] attached to either edge of a 1-in. plate with a riser at the center. This test bar has been termed the Horizontal 5/8-in. web-Webbert.

Figure 4 represents a standard keel-block casting with two coupons.

Figure 5 represents a two-bar casting which, like the L.T.B. shown by Fig. 2, has the advantage that bars can be cast to size. Experience demonstrated, however, that the properties of this casting paralleled those obtained in the L.T.B.-1, Fig. 2, though slightly better properties were obtained. Since the properties of these two designs paralleled one another, the two-bar casting shown by Fig. 5 was abandoned. It may be said, however, that any merit which the L.T.B. may have can also be ascribed to this two-bar casting.

Figure 6 represents the standard bar of Fig. 10 of the Federal Specification QQ-M-15a. This bar is referred to in this paper as the Navy 5/16-in. web-Webbert, because the connecting web between the riser and bar, formed in a dry sand core, is 5/16-in. thick, and it extends along the full length of the casting. A modification of this bar, called the Navy 5/8-in. web-Webbert, was also used. This modification, however, is not strictly a Federal specification bar since the connecting web is 5/8 in. thick instead of 5/16 in.

Figure 7 shows a test bar corresponding to Fig. 10a of the Federal Specification QQ-M-15a. This bar is referred to as the Navy 5/8-in. web-Crown, because it is the Crown-type bar with a 5/8-in. web, formed in a dry sand core, connecting the riser with only the grip ends of the test bar.

Figure 8 illustrates the British standard D.T.D. bar recommended for some nonferrous metals in Britain.

Figure 9 illustrates another type of bar[2] which contains four test bars per mold cavity, two of the Webbert type and two of the Crown type. They are referred to in this paper as the Modified 5/8-in. web-Webbert and the Modified Crown bar. Both bar types have a 5/8-in. web, but in the Crown type, the web connects the riser with the grip end of the test bars only.

Only two of these bars were employed throughout the entire inves-

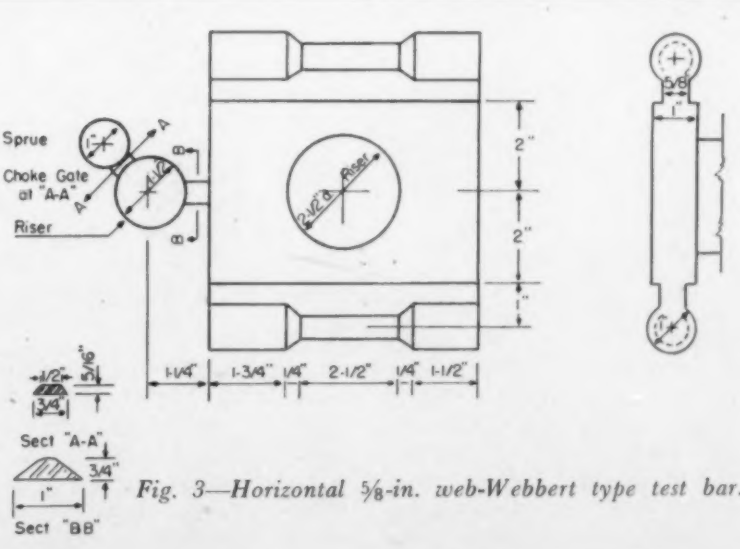


Fig. 3—Horizontal 5/8-in. web-Webbert type test bar.

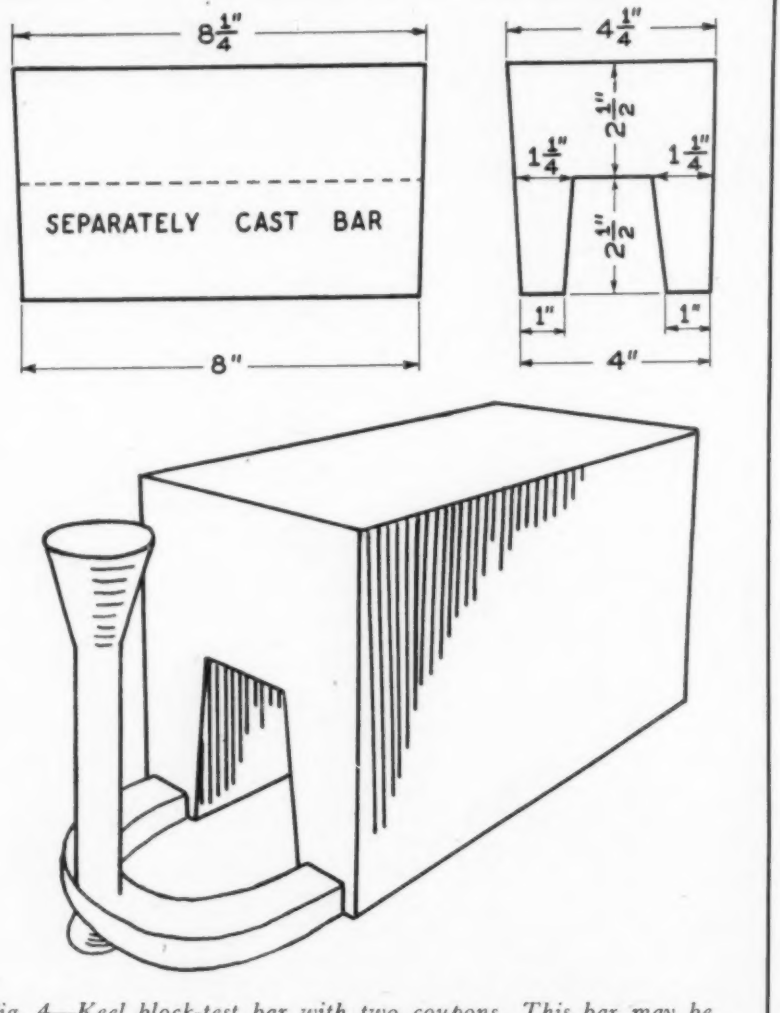


Fig. 4—Keel block-test bar with two coupons. This bar may be gated at the bottom or at the top of the coupon, and is normally used for alloys subject to shrinkage defects.

tigation which has involved melting over 30,000 lb. of metal in 250 to 350-lb. heats having the following typical composition:

Element	*Per Cent*
Copper	84.8
Tin	4.56
Lead	4.56
Zinc	5.16
Nickel	0.57
Antimony	0.14
Iron	0.15

The initial investigations were carried out with five different designs; the L.T.B.-1, two-bar, Horizontal 5/8-in. web-Webbert, Universal, and the keel-block casting. The initial keel-block castings were made with the gate entering the bottom of the coupon. Later this was altered and the gate was introduced at a point just above the coupon. None of the tensile properties obtained in this casting were very satisfactory, and it was concluded that this design was unsuited for 85-5-5-5 alloy.

Although a considerable amount of additional data was obtained, those listed in Table 2 are representative, except that the properties of the Universal bar are unusually high and the melt quality is better than that usually obtained by the melting technique described in the footnote. As previously noted, the properties of the two-bar casting, Fig. 5, were quite similar to the L.T.B., Fig. 2. Consequently, the two-bar casting, though slightly superior to the L.T.B., was abandoned. Likewise, the keel block, Fig. 4, was also abandoned because of the obvious unsuitability of this design for 85-5-5-5 alloy. Because the keel block represented a heavy casting, this design was used for certain investigations, to be described later, on the effects of mold permeability.

Effect of Blind Risers on the Tensile Properties Obtained in the L.T.B. Test Bar. The initial properties obtained on the L.T.B. test bar were somewhat lower than expected, and as indicated in Table 2, the risers were blind. As a consequence, the blind risers were compared with open risers with the following results listed in Table 3.

The adverse effect of the blind riser was rather surprising, but apparently the open riser provides better feeding than the blind riser. Subsequently, all the later data on this test bar were obtained with open risers 3 in. high above the parting line.

Modification of the Design of the Horizontal 5/8-In. Web-Webbert Test Bar and the Effect of Pouring Technique. It is a characteristic of Webbert-type test bars to produce several per cent of shoulder fractures. In an effort to reduce this tendency and to improve the average properties obtained, the Horizontal 5/8-in. web-Webbert, as shown by Fig. 3, was modified. The single center riser was replaced by two, each extending lengthwise across the end between the grip sections of the two bars. The gates were then cut so that one gate entered each of these risers for the purpose of producing hot metal in them. The average tensile properties obtained by this method were identical to those obtained by the standard bar as shown by Fig. 3. Furthermore, the proportion of shoulder fractures was approximately the same.

A second modification was tried whereby the two risers across the plate were used as described, but the connecting web between the riser and test bar extended along the grips only, making a Crown-type bar. This modification produced low and erratic properties. No other modifications were tried, and the original design, shown by Fig. 3, was used throughout the remainder of the investigation.

Some tests were conducted on the effects of pouring technique on the average and consistency of the tensile properties of this design. While no data are included here, the results definitely indicated that the mold should not be tilted, the ladle lip should be held low, and the sprue kept full during the pour;

otherwise, lower or more erratic properties are obtained.

The D.T.D. Test Bar. The D.T.D. test bar, shown by Fig. 8, is commonly used for light alloys in England. Preliminary investigations were made of this design with the results listed in Table 4.

It is obvious that the D.T.D. test bar produces properties markedly inferior to the other two designs. While an exhaustive test was not made of this bar, it was concluded that, in view of the low tensile properties, factors other than melt quality must considerably influence the properties obtained with 85-5-5-5 alloy. Accordingly, no additional investigation was made of this design. It should be observed, however, that the above listed properties on the D.T.D. bar are superior to those reported by English investigators[12].

IV—Correlation of Micro- and Macrostructure with the Fracture and Test-Bar Properties

Test-bar designs are of two general types, the first is the Webbert type on which the connecting web extends the full length of the casting between the bar and the riser. The second type of test bar is cast to shape, and the riser is connected with the test bar only along the grip ends as in the Crown-type bars, or at the ends of the test bars as in the L.T.B. and two-bar castings. These two types of bars have distinguishing characteristics in respect to their structure and fracture. In general, the fracture of 85-5-5-5 alloy has a fine grayish appearance with or without areas which appear to be open and have a brassy or coppery color.

Fracture Appearance of Webbert-Type Bars

In general, the open or coppery area in Webbert-type bars extends from the place of attachment of the web to beyond the center of the test bar. Figure 10 shows a photograph at X13 of the structure of a cross section of a Webbert-type bar having a rather open fracture and low properties indicative of a melt of low quality. That portion of the fracture lying within the line drawn on the photograph was characterized by an open, coppery fracture. It will be noted that this area is also unsound.

All of the larger dark areas on this photograph are voids which, in conformity with usage of the term for other metals having similar defects, are called microporosity, as described before in this paper. On the other hand, this coppery area in the center of the fracture outlined on the photograph merges into the fine gray fracture nearer the exterior of the bar corresponding to the sound portion. This structure is a characteristic of all Webbert-type bars. The area corresponding to the unsound portion, which in turn corresponds to the open coppery fracture, increases with any factor that increases

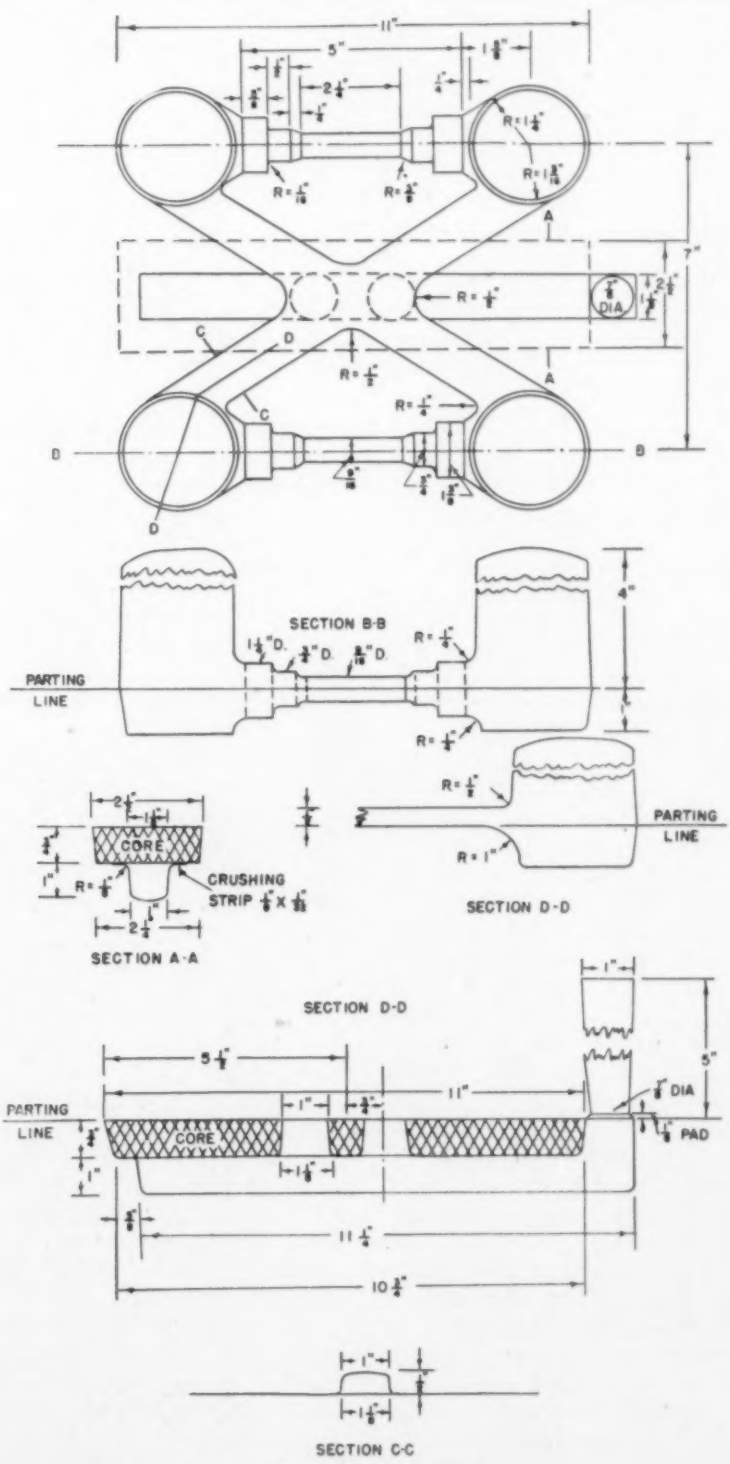


Fig. 5—The two-bar design.

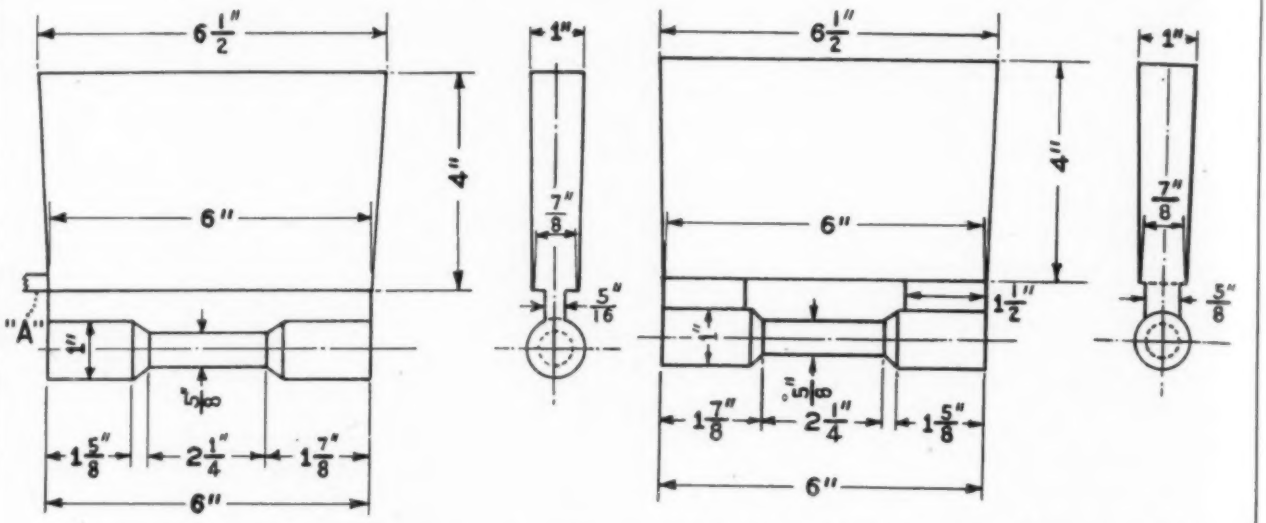


Fig. 6 (Left)—Navy 5/16-in. web-Webbert test bar (Fig. 10 of Federal Specification QQ-M-15a). This design was modified to form the Navy 5/8-in. web-Webbert by increasing the web thickness to 5/8 in. The mold is gated at "A" and two bars per mold are poured from a common runner. Fig. 7 (Right)—Navy 5/8-in. web-Crown test bar (Fig. 10a of Federal Specification QQ-M-15a).

the gas content of the melt or reduces the feeding of the test bar.

As the connecting web thickness is increased, at least within limits, the area of the unsound portion will increase also, all other factors being the same. The fine grayish portion of the fracture occurs through the lead particles, producing the characteristic color of lead. The open coppery areas of the fracture are through the microporosity, revealing the coppery surfaces of the cavities. The appearance of the fracture thus shows the relative amount of microporosity at the fracture.

Fracture Appearance of Cast-to-Shape Bars

The second type of bars, characterized by the Crown-type and the L.T.B. bars which are cast to shape, invariably show a radial columnar structure if they are poured in 85-5-5-5 alloy from a high temperature. These bars are not characterized by the presence of unsoundness in the form of microporosity because of the strongly directional solidification. As a consequence, the fracture is invariably of a fine grayish appearance since it occurs through the lead particles.

However, the lead particles lying between the axes of the alpha dendrites are related to the radial structure, making this radial structure apparent on the fracture. If localized coppery areas are visible on such test bars, they invariably occur in the center, and are caused by localized shrinkage of the centerline type. This is especially common at low pouring temperatures.

Microstructure

Photographs at X13 of unetched sections of both types of test bars are shown by Figs. 11 and 12. Figure 11 shows the unetched structure of an L.T.B. test bar which had a fine gray columnar structure. Since this is a longitudinal section, the radial arrangement of the lead particles is not apparent. The structure shown by this photograph is typical as there is a high order of soundness, and the lead particles are very fine.

Figure 12 is a typical structure of a Webbert-type bar on a longitudinal section through the plane of the web. The larger dark patches on this photograph are microporosity which is quite evident on two-thirds of the section adjacent to the web. The very fine small dark particles are lead which are very much coarser than those shown by Fig. 11. Despite the greater unsoundness of this test bar, the tensile properties are substantially similar to those obtained in the sounder bar shown by Fig. 11. As indicated in the following paragraphs, this is probably because of the more favorable lead size and shape in the bar shown by Fig. 12.

While the correlation of structure and fracture is fairly straightforward and apparently well understood, the correlation of structure and fracture with the tensile properties of the bar is somewhat more difficult. Although the test-bar properties cannot be accurately predicted from the appearance of the structure and the fracture, the two main factors determining test-bar properties, at least of Webbert-type bars apparently are as follows:

1. The amount of microporosity as judged by the size and degree of openness of the coppery area on the fracture.
2. The size, shape, and distribution of the lead particles.

In general, as the gas content of the melt increases, or as the degree of feeding provided decreases, all other factors the same, the more open the coppery area on the fracture of Webbert-type bars, the greater the amount of microporosity formed, and the lower the tensile properties obtained. This is well illustrated by comparison of Figs. 12 and 13. The bar represented by Fig. 13 was poured from a very gassy melt; namely, one of low quality; the fracture was an open coppery one, the amount of microporosity in the section was great, and

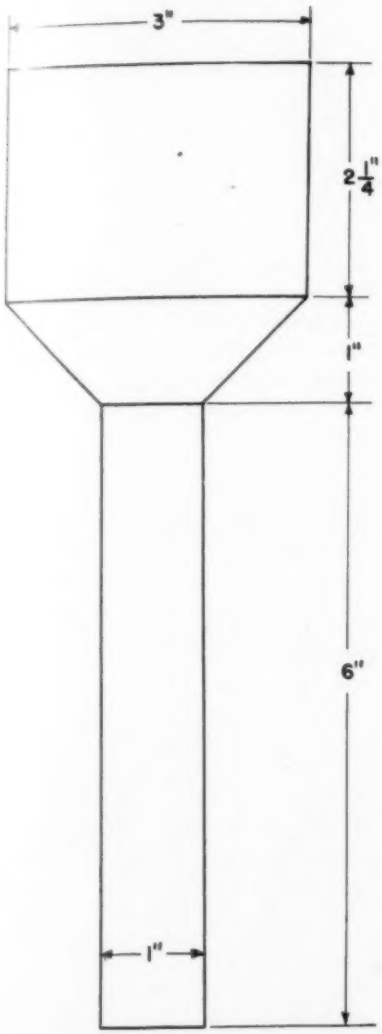


Fig. 8—British D.T.D. type test bar as standardized for light alloys.

the properties low. This is a typical example showing the effect of a melt of low quality on the incidence of microporosity in a Webbert-type bar and its resultant effects on the fracture appearance, and on the properties of the bar.

In general, very fine lead particles or coarser angular lead particles have an adverse effect on tensile properties. Conversely, coarser, rather rounded lead particles are the most desirable and most conducive to the production of good tensile properties. Figures 14a, 15a, 16a, and 17a show microradiographs enlarged thirteen times. All of the castings represented by these four photographs were poured from the same melt from approximately the same temperature into the same design of test bar, but into various types of molds.

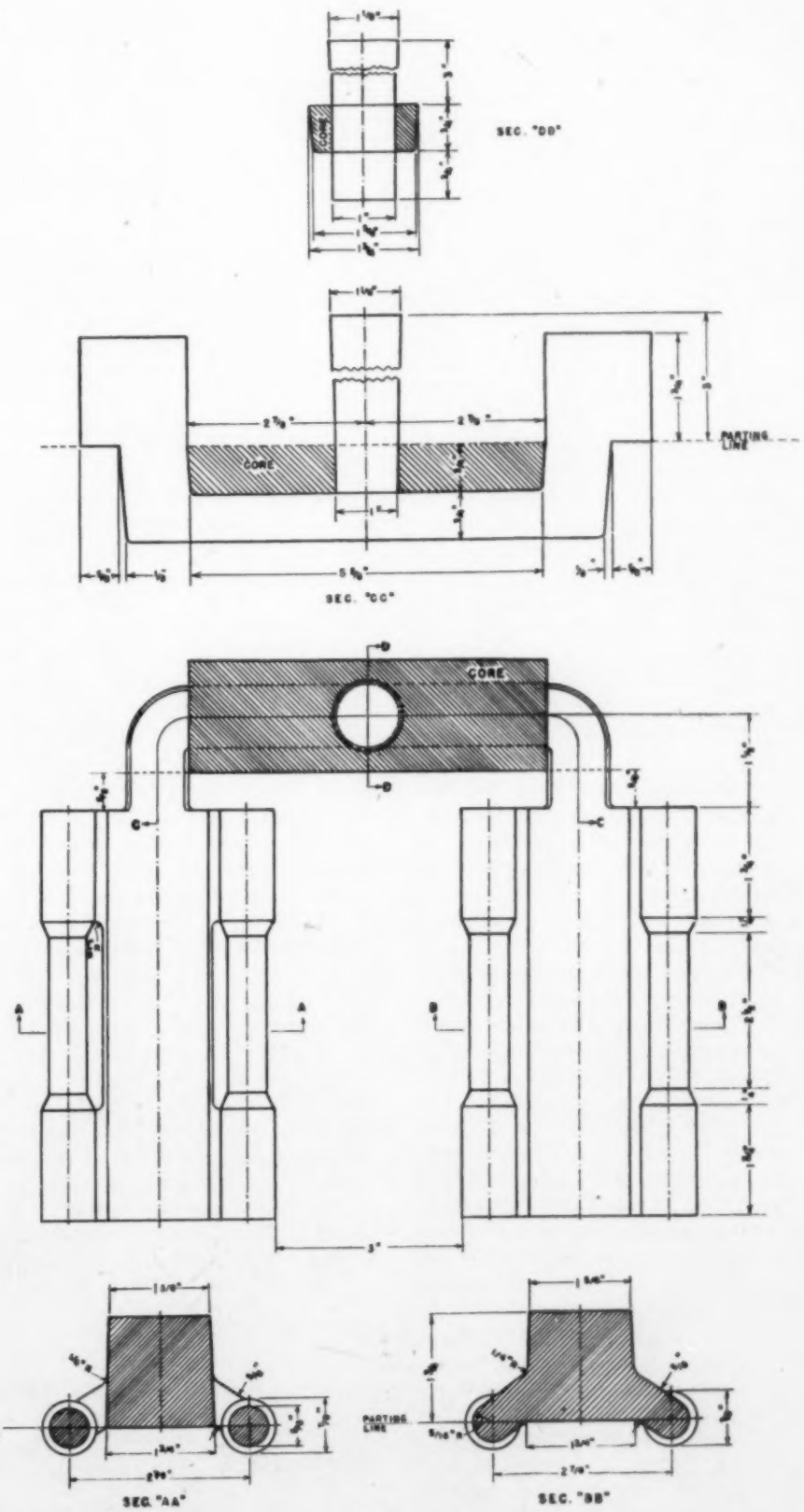


Fig. 9—Modified 5/8-in. web-Webbert and Crown-type test bars.

Table 2

PRELIMINARY DATA ON THE COMPARATIVE PROPERTIES OF FIVE DIFFERENT TEST-BAR DESIGNS[a]

Test-Bar Design	*Average Pouring Temp., °F*	*Tensile Strength, psi.*	*Yield Strength, psi.*	*Elongation, per cent*
Horizontal 5/8-in. web-Webbert	2110	37,800	16,700	37.4
Universal	2100	34,775	15,600	34.4
Two-Bar[b]	2090	37,075	17,100	30.2
L.T.B.[b]	2120	35,675	17,175	29.5
Keel Block	2080	29,650	15,150	21.2

[a] An all ingot metal charge of 85-5-5-5 alloy was melted in the high-frequency, clay-graphite lined, induction furnace, using a glass slag, 0.25 per cent zinc added, melt skimmed and exposed to oxidation for about 5 min., then transferred to a ladle, 2 oz. of phosphor-copper added per 100 lb. of melt, then poured into natural green-sand molds containing about 6 per cent moisture with an A.F.A. permeability of approximately 20.

[b] Made with blind risers.

As noted in the figure captions, there is a considerable difference in tensile properties obtained which can, in general, be explained by the size, shape, and distribution of the lead particles as described above and as illustrated by Figs. 14b, 15b, 16b, and 17b. Further illustrations of the effects of pouring temperature on the soundness and structure of test bars are considered under the heading "Effect of Pouring Temperature on the Microstructure and Soundness."

The grain size or shape does not appear to be a very important factor, except insofar as it affects the size, shape, and distribution of the lead particles. It has been noted upon many occasions that, as the pouring temperature is decreased, the grain size decreases. When equiaxed fine grains are produced, the lead particles become quite angular, producing lower elongation values.

V—Effect of Mold Material

To determine the effects of moisture content and A.F.A. permeability of natural and synthetic sands upon the tensile properties produced, the Horizontal 5/8-in. web-Webbert test bar and the keel-block castings were selected as being representative of a fairly light and a fairly heavy casting, respectively. The A.F.A. permeability of natural-green sand was varied from approximately 15 to 142 by mixing various proportions of a No. 1 Albany sand with an open Tennessee natural sand. Approximately six molds were prepared having an A.F.A. permeability varying from 15 to 142.

One series of molds was prepared with increasing permeability, and a second series with decreasing permeability in order to offset the effect of the slightly decreasing pouring temperature as the molds were successively poured. When investigating the effect of moisture content, the permeability was held approximately constant excepting for the slight decrease with increasing moisture, and the moisture content of the sand was increased or decreased stepwise as described above.

Fifteen melts were prepared in a high-frequency induction furnace using a technique to produce high-quality metal, and to make reasonably certain this was accomplished, a 5/8-in. web-Webbert test bar was poured from a temperature of 2150° F. or above. Table 5 shows the heats prepared, the range in permeability, the moisture content employed, and the average results obtained at the two extremes of the variant listed in the third and fourth columns.

It is evident from Table 5 that the tensile properties obtained in the Horizontal 5/8-in. web-Webbert bar decrease slightly with increasing moisture in the natural sand. On the average, the tensile properties of this bar are practically independent of the A.F.A. permeability of either the natural or synthetic sands. On the other hand, the tensile properties of the keel-block casting increase slightly with increasing A.F.A. permeability of the natural sand.

While the data on the effects of A.F.A. permeability of the synthetic sand or of the moisture in natural sand are inadequate, apparently these factors do not greatly affect the tensile properties of this casting. In general, it may be concluded that only large increases in moisture content or A.F.A. permeability produce changes in the tensile properties, and even then, in these particular castings the effects are small. This does not mean, however, that commercial castings will show the same insensitivity because of the very considerable casting differences which may exist.

Thus, high moisture content is not recommended because it might possibly produce difficulties with blows and other foundry defects. The significance of these data lies in the conclusion that the 85-5-5-5 alloy apparently does not absorb gas to a noticeable extent during the pour or in the mold cavity. Otherwise, it would be expected that there would be considerable effects of the variations in moisture content or A.F.A. permeability within the range investigated.

It will be noted that the tensile properties obtained in the Horizontal 5/8-in. web-Webbert casting were markedly lower when poured in the synthetic sand than when the same metal was poured into the natural green-sand mold. On the other hand, the tensile properties of the keel block were the same in the synthetic sand mold as in the natural sand mold. Apparently then, the effect of mold material will depend upon the design of the castings.

Data in Table 6 show the effect of mold material upon the tensile

Table 3

RELATIVE MERITS OF OPEN AND BLIND RISERS ON THE L.T.B.-1 TEST BAR[a]

	Tensile Strength, psi.			*Per Cent Elongation in 2 In.*		
Riser Type	*Minimum*	*Average[b]*	*Maximum*	*Minimum*	*Average[b]*	*Maximum*
Open Riser	34,500	37,450	39,000	22.2	34.1	46.2
Blind Riser	32,500	34,975	36,500	18.7	24.4	30.1

[a] A charge made of 50 per cent remelt and 50 per cent new ingot of 85-5-5-5 alloy was melted in a silica-lined, high-frequency induction furnace without the use of any cover. The melt was transferred to a ladle where 1 per cent zinc and 4 oz. of phosphor-copper per 100 lb. of metal were added during the pour. The metal was then poured into natural green-sand molds having an A.F.A. permeability of about 20 at 6 per cent moisture. Two heats were prepared, and molds having alternately open and blind risers were poured from a temperature range of 2250 to 2100° F.

[b] Average of seven bars poured from two different melts.

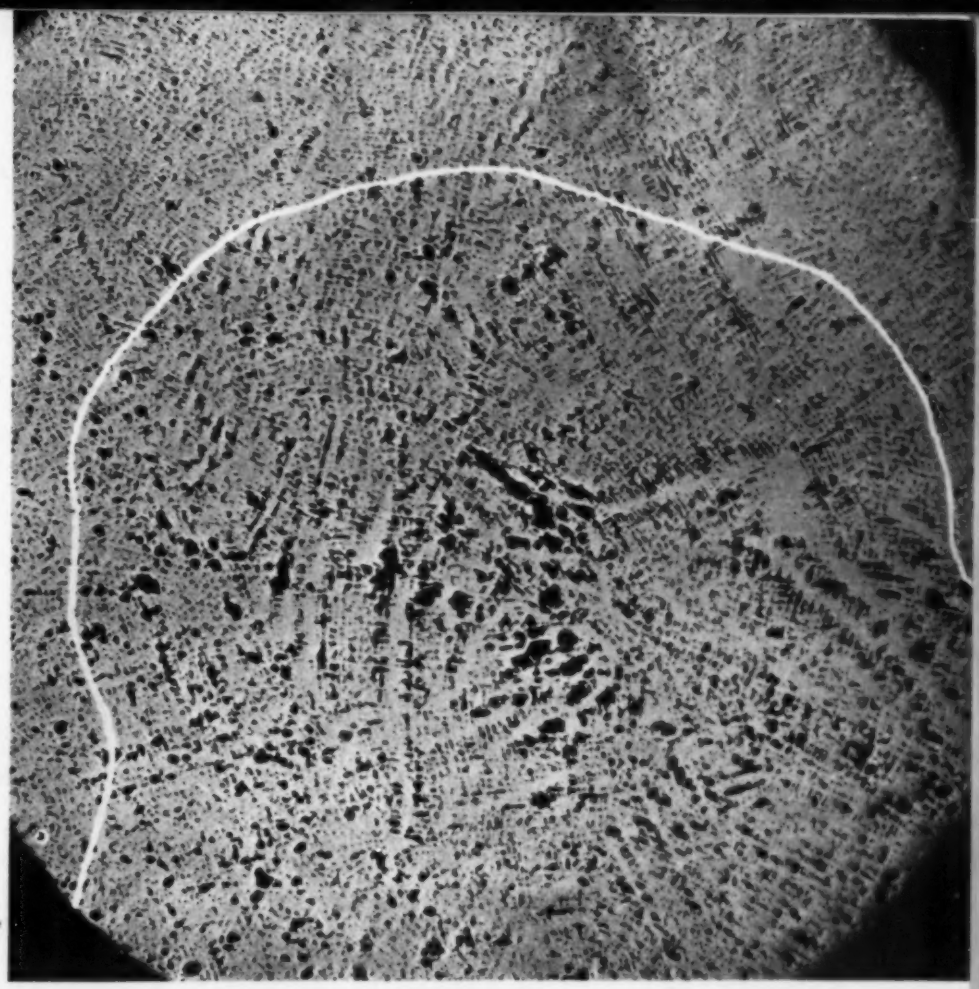


Fig. 10—X13. Unetched. Polished cross section at fracture of a Horizontal 5/8-in. web-Webbert test bar of 85-5-5-5 alloy poured at 2250° F. from poor quality metal. Tensile strength was 34,400 psi., and elongation was 29.3 per cent. The fracture was fine gray at the exterior with a large open coppery area in the center enclosed by the line and extending to the 5/8-in. web. This is a typical cross section of a Webbert bar poured from a poor quality melt. The large quantity of microporosity corresponding to the open coppery area is responsible for the relatively low tensile properties. The size of the voids and the total area of the unsound portion decreases with decreasing gas contents of the melt. Examination of the fracture shows the extent of this unsoundness, and for a given test-bar design of the Webbert type, good correlation between fracture and test-bar properties can be noted.

properties obtained in the 5/8-in. web-Webbert test bar. The natural Tennessee sand had a permeability of 38 at approximately 6 per cent moisture. The synthetic sand was prepared from a washed silica sand which had an A.F.A. permeability of 108 when it was bonded with 2 per cent western bentonite, 6.50 per cent Revivo (fire clay type binder), and 3.6 per cent moisture. As indicated by the data in Table 6, various bonding materials and baking temperatures were employed.

It is obvious from the data in Table 6, that the synthetic sand baked at 1700° F., produced the highest tensile properties, while natural sand produced slightly lower properties. The synthetic green-sand molds and those baked at 450° F., regardless of the bonding material, all produced very low tensile properties in comparison with those obtained in either natural green sand or in the synthetic sand baked at 1700° F. These results were repeatedly checked, but the cause of the effects was not definitely established.

Considerable variation in microstructure occurs in the castings poured into the various molds. Figures 14a, 15a, 16a, and 17a are microradiographs of sections of test bars made in the molds prepared respectively from synthetic sand baked at 1700° F., natural sand, synthetic sand baked at 450° F., and synthetic sand not baked. As indicated by these photographs and accompanying captions, the differences in lead-particle size and shape account for the differences in properties obtained. The causes of these differences in structure, however, have not been established, but it has been tentatively concluded that the differences are caused in part by variations in cooling rate.

It will be noted that the greatest degree of soundness was obtained in the bar prepared in synthetic sand baked at 1700° F., and represented by Fig. 14a. This, however, is not typical for this mold material. The results of investigations of other similar sets of castings indicate that there is, in general, no real difference in the relative soundness of the test bars produced in the different mold materials. It should be emphasized, however, that the effect of mold material, as illustrated by the data in Table 6, applies to the Horizontal 5/8-in. web-Webbert test bar only, and may not hold for castings of other designs.

Table 4

COMPARISON OF THE HORIZONTAL 5/8-IN. WEB-WEBBERT, MODIFIED 5/8-IN. WEB-WEBBERT, AND THE BRITISH D.T.D. TEST BARS[a]

Test-Bar Design	*Tensile Strength, psi.*			*Per Cent Elongation in 2 In.*		
	Minimum	*Average*[b]	*Maximum*	*Minimum*	*Average*[b]	*Maximum*
Horizontal 5/8-in. w-W	37,800	38,090	38,600	40.8	47.3	51.4
Modified 5/8-in.-w-W	36,100	37,760	39,550	31.0	37.6	43.2
D.T.D.	34,400	34,900	35,500	23.6	27.2	29.3

[a] A 300-lb. charge consisting of 50 per cent gates and risers, and 50 per cent new ingots of 85-5-5-5 alloy was melted in a high-frequency induction furnace, lined with silica. No cover was used, the melt was transferred to a pouring ladle where 0.25 per cent zinc and 2 oz. of phosphor-copper per 100 lb. of melt were added during the transfer. The melt was then poured into natural green-sand molds having an A.F.A. permeability of approximately 20 at 6 per cent moisture. The pouring temperature range was 2070 to 2170° F.

[b] Average of four bars, two bars of each design, poured from two different melts.

VI—EVALUATION OF MELT QUALITY

It has been indicated previously that melt quality is synonymous with melt composition, with special refer-



Fig. 11 (left)—X13. Unetched. Polished longitudinal section near fracture of an L.T.B.-1 test bar of 85-5-5-5 alloy poured at 2250° F. from a melt of high quality. Density in the shoulder: 8.84. This bar had columnar radial grains extending from the exterior to the interior, and the fracture was a fine gray with this columnar radial structure apparent. Tensile properties are 36,600 psi., and 45.9 per cent elongation. Since this is a longitudinal section, the radial structure is not evident on this photograph. The high degree of soundness and very fine lead particles are typical of test bars of this design. The large dark areas are tears produced when the test bar was fractured.

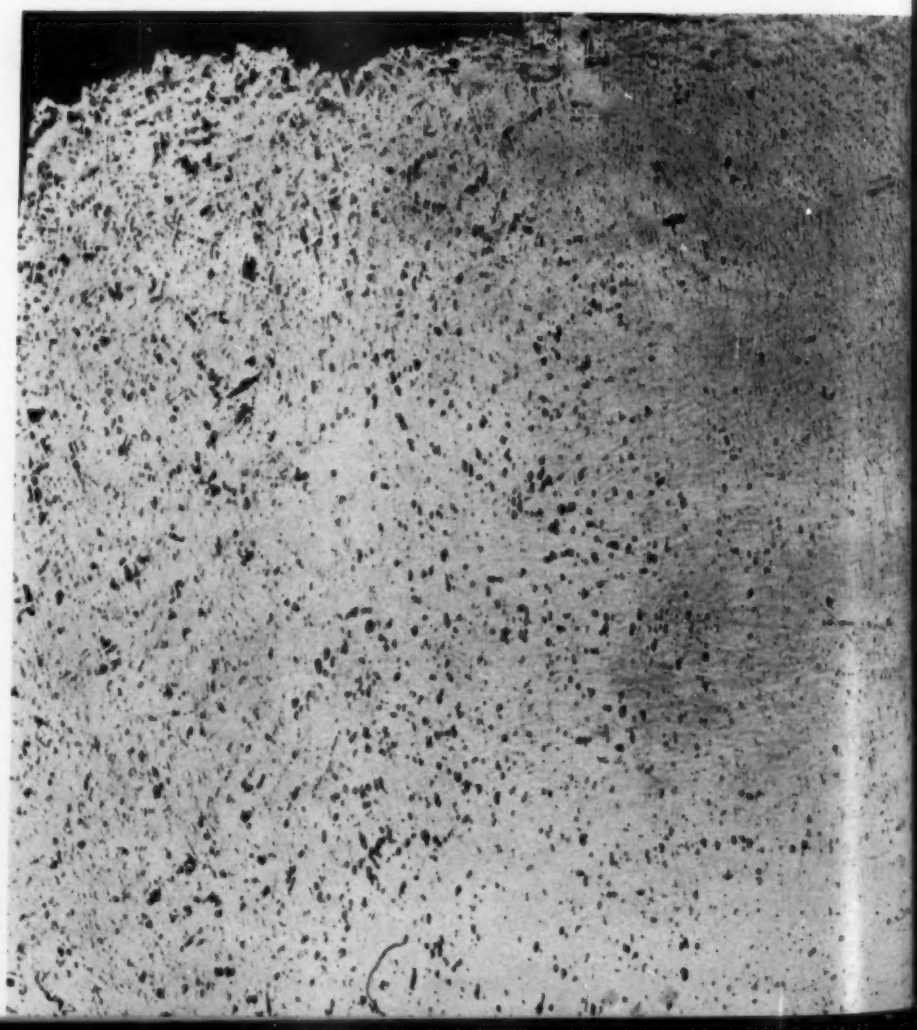


Fig. 12 (right)—X13. Unetched. Polished section of a Horizontal ⅝-in. web-Webbert test bar of 85-5-5-5 alloy poured at 2240° F. from a melt of high quality. Density in shoulder: 8.59. The fracture was fine gray at the exterior with a small area of open coppery appearance extending to the point of attachment with the web. Tensile properties were 35,800 psi., and 42.1 per cent elongation. This longitudinal section is taken through the plane of the web. While there is a considerable amount of microporosity, it is fairly fine and not characterized by marked angularity. Lead particles shown faintly on the photograph are of large size. Structures of this type generally are associated with good tensile properties. Compare with Fig. 13, which illustrates the effect of higher gas content of the melt.

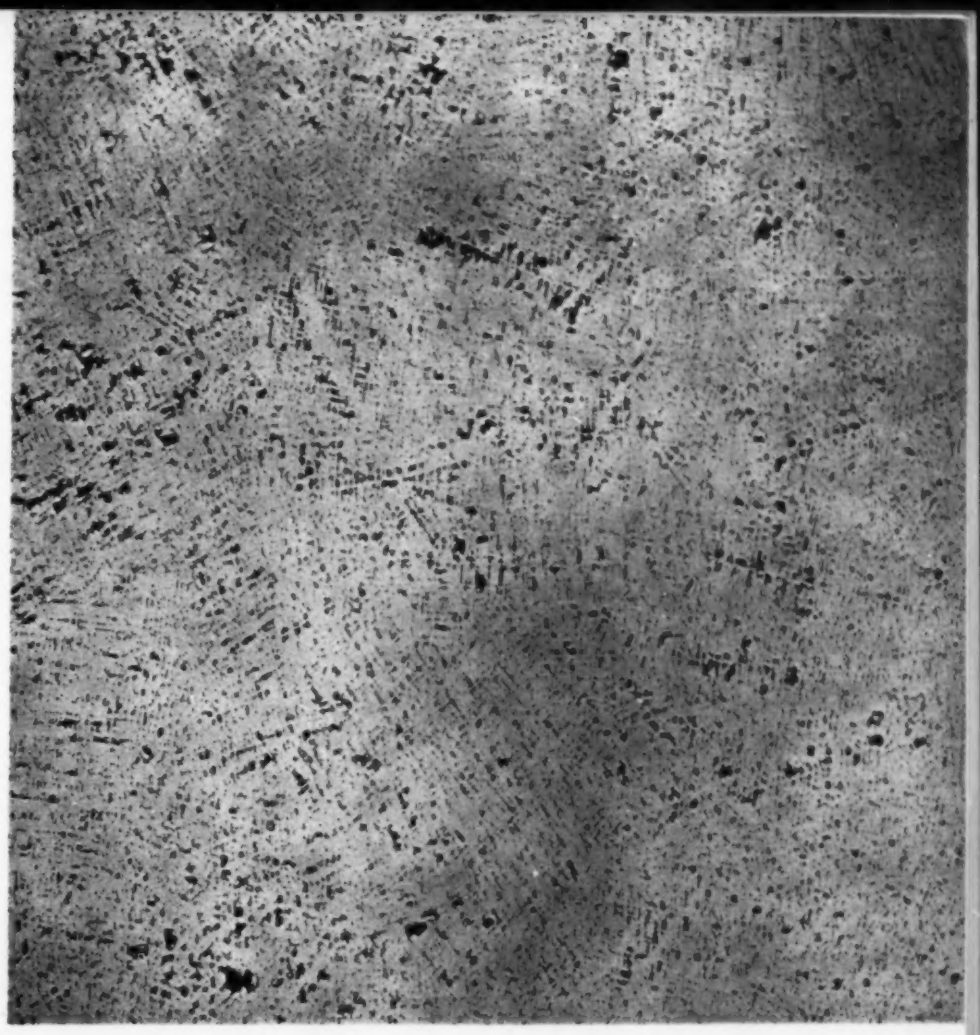


Fig. 13—X13. Unetched. Longitudinal section adjacent to the fracture of a Horizontal 5/8-in. web-Webbert test bar of 85-5-5-5 alloy poured at 2250° F. from a melt of poor quality. Density in the shoulder 8.51. This test bar had a very open coppery fracture, and produced mechanical properties of 31,000 psi. tensile strength and 17.7 per cent elongation. This melt was of poor quality; namely, it had a high gas content and the accentuation of the occurrence of microporosity is entirely evident. Compare with Fig. 12, which is entirely similar except that the specimen shown by Fig. 12 was poured from a melt having high quality.

ence to the content of gases and gas-forming constituents. It was also noted that it was the function or desired purpose of the test bars to reflect this melt quality. The measurement of melt quality became of vital importance in this investigation for two reasons:

1. The effect of various melting techniques was investigated, and it became necessary to measure the relative melt quality produced.

2. Since one of the purposes of this investigation was the evaluation of the ability of each test bar to reflect melt quality, it was essential that a means of measuring melt quality be adopted.

These melt-quality ratings could be compared with the individual test-bar properties in order to show how reliably these properties of a given design reflected the melt quality.

Forty-nine heats, prepared in three groups, were made. The first group consisted of 12 heats prepared to check the relative merits of various melting techniques. This first group of 12 heats, data on which are listed in Table 7, was prepared and poured into the following molds:

1. Horizontal 5/8-in. web-Webbert
2. L.T.B.-1
3. Modified 5/8-in. web-Webbert
4. Modified Crown
5. Universal

The second group of five heats was prepared and poured into the first four designs listed above.

The third group, consisting of 32 heats, data on which are listed in Table 8, was also prepared to determine the relative merits of the various melting techniques employed. These melts were poured into the following test bars:

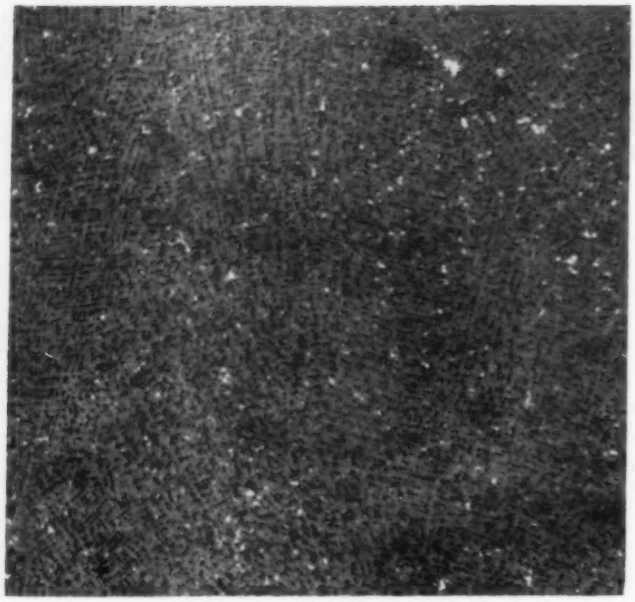
1. Horizontal 5/8-in. web-Webbert
2. L.T.B.-2
3. Modified 5/8-in. web-Webbert
4. Navy 5/8-in. web-Webbert
5. Navy 5/16-in. web-Webbert
6. Navy 5/8-in. Crown

All forty-nine melts were prepared from 300 to 350-lb. charges, and one set of each design was poured in the pouring-temperature range of 2250 to 2200° F., 2150 to 2100° F., and 2050 to 2000° F. In the first group of castings, six different melting conditions were employed and these heats were run in duplicate. Of the five heats in the second group, only two were duplicates. The third group of melts represented ten different melting conditions, including the melting conditions investigated in the first group of melts, were prepared in triplicate. In these melts, over 16,000 pounds of metal was prepared and poured under carefully controlled conditions.

For reasons indicated later, the melt quality rating of an individual heat, or of two or more heats representing one melting condition, was quantitatively evaluated by calculating the weighted average test-bar properties of all designs poured from a single heat or for a single melting condition.

These weighted averages were prepared by applying a weight of two to the tensile properties obtained at the highest pouring temperature, a weight of one to those properties obtained at the intermediate pouring temperature, while those properties obtained at the lowest pouring temperature were not considered when calculating the melt quality.

It is obvious that the melt quality ratings of each melting condition are based on the average tensile properties of 40 bars of five different designs poured from duplicate heats of Group 1, 16, and 32 bars of four different designs in heats of Group 2, and 72 bars of six different designs in the triplicate heats of Group



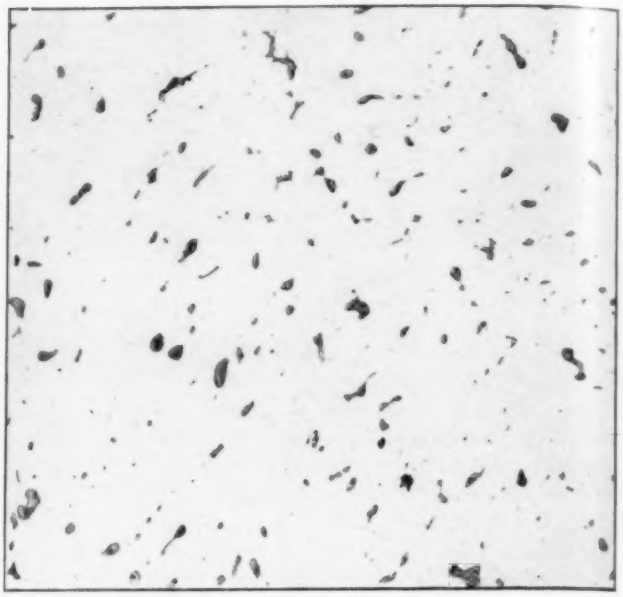
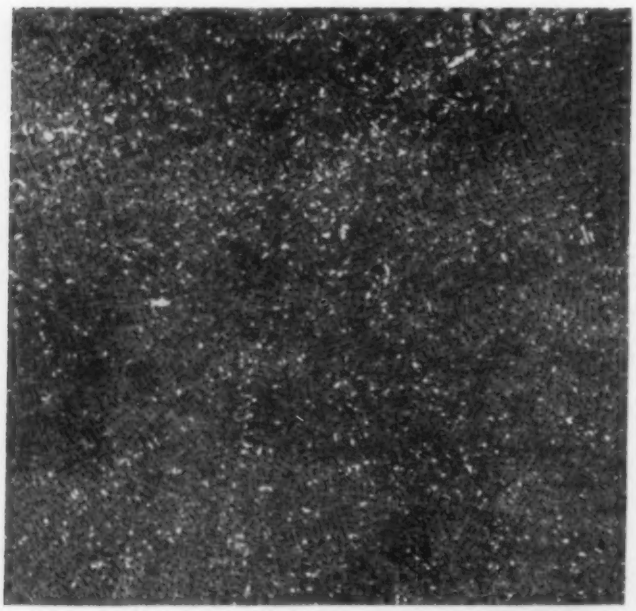
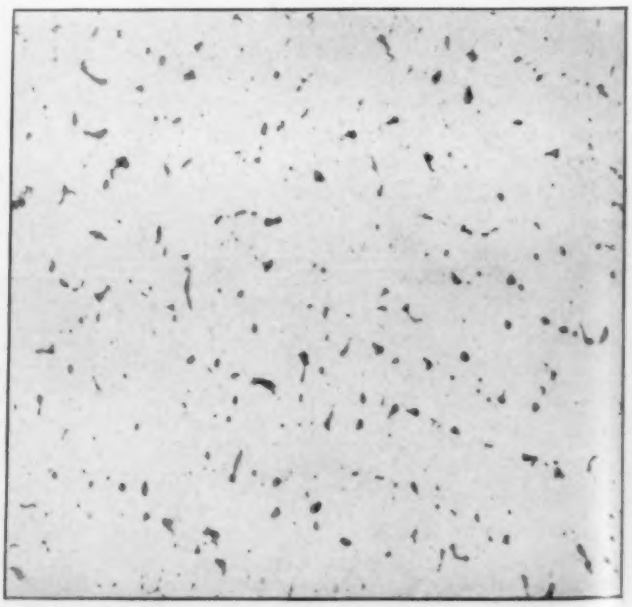
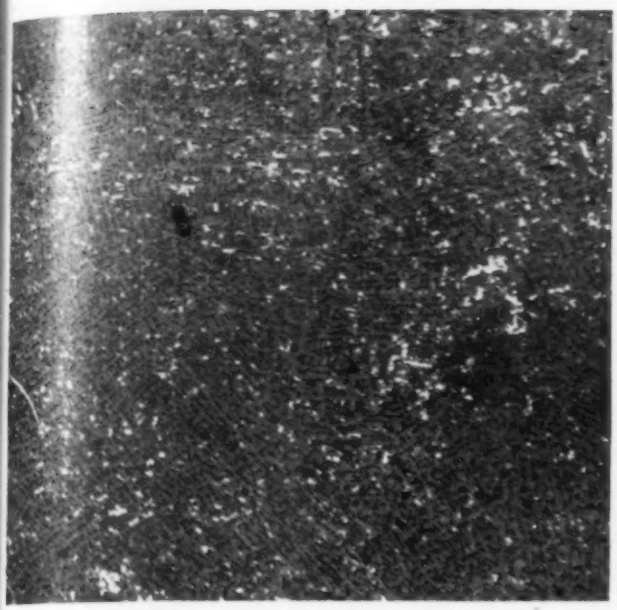


Fig. 14a (left)—Microradiograph of a cross section 0.005 in. thick cut from a Horizontal 5/8-in. web-Webbert test bar adjacent to the fracture. Density at the shoulder is 8.91. The test bar is 85-5-5-5 alloy poured at 2155° F., from a melt of high quality, into a synthetic sand mold baked at 1700° F. for about 4 hr. Tensile strength is 38,200 psi., and elongation is 50.8 per cent. The fracture was fine gray with a small area of a slightly open or coppery area adjacent to the web. The small light areas are porosity. The high degree of soundness and the rather coarse, widely separated lead particles are conducive to the good tensile properties obtained. Fig. 14b (right)—X100. Unetched. Photomicrograph of the same specimen used in Fig. 14a, in the plane of the connecting web midway between the periphery and the center of the bar. The coarse, rounded lead particles are favorable to the attainment of good tensile properties.

Fig. 15a (left)—X13. Similar to Fig. 14a in all respects and poured from the same melt at 2175° F., but into a natural green sand mold. Density at the shoulder was 8.82. As indicated by the small white areas, there is a greater amount of microporosity of a fine, rounded type, and the lead particles are more closely spaced and are, therefore, smaller. However, very good tensile strength of 37,200 psi., and 48.2 per cent elongation were obtained. But these values are slightly inferior to those possessed by the specimens shown by Fig. 14a. Fig. 15b (right)—X100. Unetched. Photomicrograph of the same specimen used in Fig. 15a, in the plane of the web and midway between the periphery and the center of the bar. The lead particles are slightly finer and more angular than those shown by Fig. 14b, accounting for the slightly lower tensile properties.







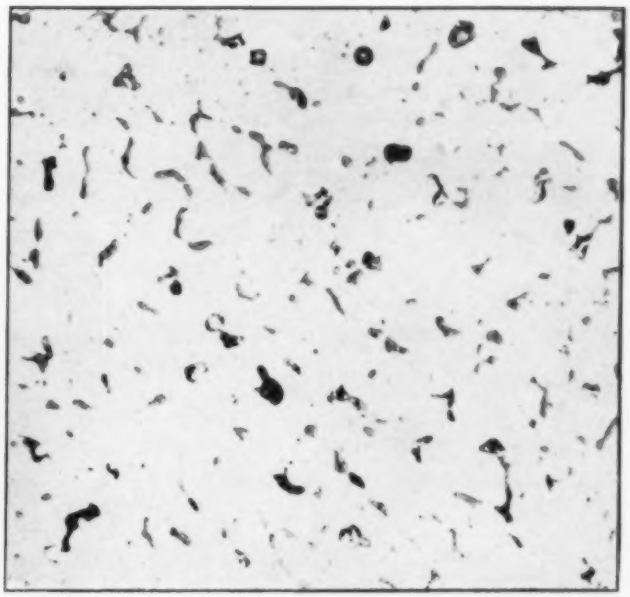
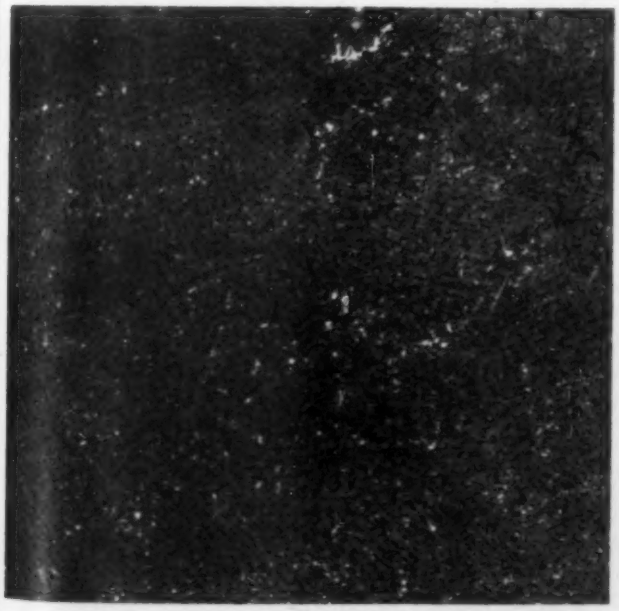


Fig. 16a (left)—X13. This microradiograph is identical to those shown by Fig. 14a and 15a in all respects, except the melt was poured at 2145° F. into a synthetic sand baked for 4 hr. at 450° F. The density in the shoulder was 8.79. Again the light areas are porous, and the small dark areas are the lead particles. Some of the lead is coarse and very angular, a shape which tends to produce low elongation values, which in this instance was 33.7 per cent with a tensile strength of 37,100 psi. Fig. 16b (right)—X100. Unetched. Photomicrograph of the same specimen used in Fig. 16a, in the plane of the web midway between the periphery and the center of the bar. The coarse, angular lead particles and to some extent the greater amount of microporosity, account for the markedly lower tensile properties obtained from the same melt as the preceding two specimens, but which were poured in different mold materials.

Fig. 17a (left)—X13. This microradiograph is identical to Fig. 14a, 15a, and 16a, except the melt was poured at 2115° F. into a green, synthetic sand, which was not dried. The density in the shoulder was 8.83. There is much angular porosity, the lead particles are closely spaced, and many of them are of angular shape. These structural conditions tend to produce low elongation values, which for this specimen was 24.7 per cent with a tensile strength of 35,200 psi. Fig. 17b (right)—X100. Unetched. Photomicrograph of the same specimen used in Fig. 17a, in the plane of the web midway between the periphery and the center of the specimen. As compared with Fig. 16b, this bar is sounder, but the lead particles are finer and at least as angular, with the result that the tensile properties are lower.



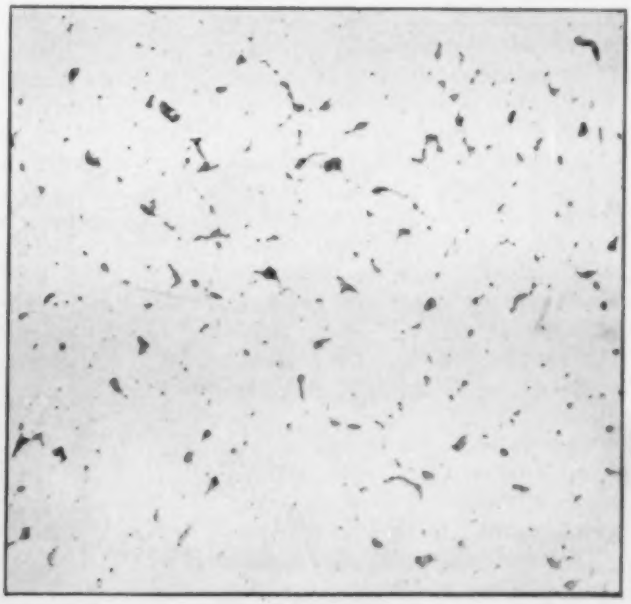


Table 5

EFFECT OF MOISTURE CONTENT AND A.F.A. PERMEABILITY OF NATURAL AND SYNTHETIC SANDS ON THE TEST-BAR PROPERTIES OBTAINED IN THE HORIZONTAL 5/8-IN. WEB-WEBBERT AND THE KEEL-BLOCK CASTINGS[a]

To Evaluate Effect of:	*Sand, Type*	*Moisture, per cent*	*A.F.A. Perm.*	*Test-Bar Design*	*Variant*	*No. of Bars Averaged*	*Tensile Strength, psi.*	*Elongation, per cent*
Moisture	Natural	5.5 to 9.5	18 to 13	Horizontal 5/8-in. w-W	5.5-6.0% Moisture	8	37,800	45.9
					7.8-9.5% Moisture	12	38,100	40.5
Moisture	Natural	5.8 to 10	17 to 13	Horizontal 5/8-in. w-W	5.8- 6.8% Moisture	12	38,100	44.7
					9.0-10.0% Moisture	8	38,175	37.2
A.F.A. Perm.	Natural	6.0	17 to 117	Horizontal 5/8-in. w-W	19-26 Perm.	8	38,400	40.1
					86-117 Perm.	8	38,400	35.8
A.F.A. Perm.	Natural	7.0	11.8 to 54	Horizontal 5/8-in. w-W	11.8-15.2 Perm.	8	37,800	37.3
					46-54 Perm.	16	38,400	37.2
Moisture	Natural	5.8 to 10	17 to13	Keel block	5.8% Moisture	2	34,200	28.2
					8.0-10.2% Moisture	3	36,250	30.6
					Standard (3)	2	37,800	52.7
A.F.A. Perm.	Natural	6.5	12.6 to 80	Keel block	12.6 Perm.	4	28,000	16.0
					80 Perm.	4	30,300	19.7
					Standard (3)	2	36,800	40.3
A.F.A. Perm.	Natural	7.0	15.2 to 138	Keel block	15.2 Perm.	6	29,100	15.6
					63-138 Perm.	6	31,000	20.2
					Standard (3)	4	38,000	42.2
A.F.A. Perm.	Synthetic	3.4	14.7 to 142	Horizontal 5/8-in. w-W[b]	14.7-22.7 Perm.	8	35,500	27.4
					85-142 Perm.	12	36,150	29.4
					Standard (3)	2	37,200	40.1
A.F.A. Perm.	Synthetic	3.4	12.6 to 138	Horizontal 5/8-in. w-W[b]	12.6 Perm.	8	36,700	32.2
					96-138 Perm.	16	36,500	28.8
					Standard (3)	4	37,900	43.1
A.F.A. Perm.	Synthetic	3.4	14.7 to 134	Keel block	14.7 Perm.	8	29,200	20.0
					134 Perm.	6	28,300	18.6
					Standard (3)	4	36,600	40.5

[a] Melts were made in a silica-lined, high-frequency induction furnace, using the C_1 method of charging, described below, employing 50 per cent ingot and 50 per cent gates and risers. The surface of the melts were skimmed three times, as described in the text. Tapping temperatures were 100° F. higher than the pouring temperatures. Melts were tapped into a preheated, clay-lined ladle, and ¼ per cent zinc and 2 oz. of 15 per cent phosphor-copper per 100 lb. of charge were added to the stream during the transfer to the ladle. Pouring temperatures ranged from 2185 to 1900° F.

[b] While the A.F.A. permeability had no apparent effect upon the test-bar properties obtained in the synthetic sand, the latter mold material produced markedly lower tensile properties in the Horizontal 5/8-in. web-Webbert than were obtained by the natural green-sand molds having an A.F.A. permeability of approximately 20 at 6 per cent moisture. On the other hand, the tensile properties obtained in the keel-block castings were approximately the same with both types of sand.

(3) Horizontal 5/8-in. web-Webbert test bars cast in a natural sand mold of 6 per cent moisture and 20 A.F.A. permeability.

3 shown in previous paragraphs.

As indicated by Fig. 18, the weighted average elongation bears a direct relationship to the weighted average tensile strength obtained. As indicated by this chart, the tensile strength decreases almost linearly with the decrease in elongation as the melt quality decreases. Thus, if the melts are arranged in the order of decreasing weighted average elongation values, the order of the weighted average tensile strength values also decreases in essentially the same manner. This method of rating melt quality has the advantage that it is simple and readily applicable. Furthermore, the results obtained by it have been quite reproducible.

The Effect of Various Melting Techniques Upon the Melt Quality Produced

The relative merits of various furnace types, furnace linings, and charging methods have been evaluated. In order to simplify the presentation of the data, the melting conditions are represented by a code which is interpreted as follows:

F_1 indicates a high-frequency induction furnace, having a capacity of about 350 lb. of metal.

F_2 indicates a gas-fired pot-type tilt furnace, having a clay-graphite crucible holding about 400 lb. of metal. The oxygen content of the furnace atmosphere was maintained at approximately 0.5 per cent.

F_3 indicates a gas-fired barrel-type firebrick-lined furnace with a burner in each end and having a total capacity of about 800 lb., but in which 350 to 400-lb. charges were melted. This furnace was mounted on the trunnions of a Detroit rocking-furnace, and when the charge was melted down, the furnace was rocked through an arc of 50 degrees with two cycles per minute. The oxygen content of the furnace atmosphere was maintained at approximately 0.5 per cent.

L_1 indicates a silica lining.

L_2 indicates a clay-graphite lining or crucible.

L_3 indicates a firebrick lining.

C_1 indicates the method of charging whereby about five ingots of the 350-lb. charge are placed in the bottom of the crucible where they are melted without further addition. As soon as these five ingots are partially melted, the power is reduced to prevent undue agitation, if the melting is being done in a high-frequency induction furnace, and one ingot or piece of scrap is melted at a time until the crucible is filled, and the entire charge completely melted.

C_2 indicates a method of charging whereby the crucible is filled with metal, the small scrap being charged first with the larger ingots and

Tensile Properties of Five Designs of Test Bars Poured from 85-5-5-5 Alloy Melted in a High-Frequency In

Furnace	Lining	Charge	Pouring Range, °F.	Horizontal 5/8-In. Web-Webbert — Tensile Strength, psi. Min.	Avg.	Max.	Elongation, per cent Min.	Avg.	Max.	L.T.B.-I — Tensile Strength, psi. Min.	Avg.	Max.	Elongation, per cent Min.	Avg.	Max.	Tensile Strength, Min.	Avg.
F_1	L_2	C_1	2250-2220	30,000	32,700	34,700	17.5	25.7	34.6	37,500	38,150	38,800	32.0	35.0	38.0	33,800	35,925
F_1	L_1	C_2	"	34,700	35,875	37,300	32.3	36.4	43.2	38,400	38,600	38,800	43.8	44.2	44.5	35,100	36,100
		Average	"		34,288			31.0			38,375			39.6			36,012
F_1	L_2	C_1	"	33,600	34,233	34,600	27.3	31.8	34.5	38,500	38,900	39,300	41.3	41.6	41.8	34,000	35,025
F_1	L_2	C_2	"	33,100	34,075	34,900	23.6	26.9	31.0	38,000	38,200	38,400	39.0	39.2	39.3	35,200	35,925
		Average	"		34,154			29.4			38,550			40.4			35,475
F_2*	L_2	C_1	"	34,400	34,933	35,300	32.0	35.3	37.9	37,000	37,550	38,100	34.8	38.0	41.3	35,200	35,425
F_2*	L_2	C_2	"	29,900	31,525	33,600	24.6	28.1	34.1	35,700	35,750	35,800	31.7	33.8	36.0	30,300	31,200
		Average	"		33,229			31.7			36,650			35.9			33,312
F_1	L_1	C_1	2150-2120	35,000	36,550	37,200	29.4	37.3	41.7	38,100	38,150	38,200	40.8	40.8	40.8	37,000	37,325
F_1	L_1	C_2	"	34,200	35,550	36,700	29.1	34.5	41.9	35,800	37,150	38,500	26.3	34.3	42.3	36,200	37,100
		Average	"		36,050			35.9			37,650			37.6			37,212
F_1	L_2	C_1	"	34,000	36,075	37,100	25.2	35.8	43.8	36,000	37,500	39,000	26.7	35.1	43.5	36,600	37,575
F_1	L_2	C_2	"	35,400	36,375	37,000	30.0	38.9	43.8	32,300	35,150	38,000	20.2	29.4	38.6	37,000	37,650
		Average	"		36,225			37.4			36,325			32.2			37,612
F_2*	L_2	C_1	"	33,800	35,425	36,700	25.3	33.9	42.8	36,500	37,350	38,200	33.0	36.4	39.7	36,400	37,275
F_2*	L_2	C_2	"	32,600	34,675	36,300	24.3	33.6	43.4	32,300	35,250	38,200	22.4	32.3	42.2	30,200	34,100
		Average	"		35,050			33.8			36,300			34.4			35,688
F_1	L_1	C_1	2050-2020	38,200	38,625	39,300	36.5	39.1	40.8	34,600	36,700	38,800	24.4	33.6	42.8	37,800	38,700
F_1	L_1	C_2	"	38,400	38,975	39,700	35.7	39.5	43.0	34,400	36,600	38,800	22.5	35.2	48.0	38,000	38,325
		Average	"		38,800			39.3			36,650			34.4			38,512
F_1	L_2	C_1	"	35,100	37,500	38,700	23.2	32.6	36.4	31,900	32,200	32,500	17.6	18.6	19.5	35,900	36,075
F_1	L_2	C_2	"	37,800	38,750	40,100	37.0	41.0	42.7	36,000	37,850	38,700	24.1	34.8	45.4	36,500	37,100
		Average	"		38,125			36.8			35,025			26.7			36,588
F_2*	L_2	C_1	"	38,700	39,450	40,500	36.7	39.3	42.4	31,200	34,050	36,900	19.4	22.7	26.0	36,900	37,775
F_2*	L_2	C_2	"	35,100	35,650	36,600	27.4	33.4	43.0	35,800	36,550	37,300	33.6	34.0	34.3	36,900	37,575
		Average	"		37,550			36.4			35,300			28.4			37,675

*Gas furnace operated to produce approximately 0.5 per cent O_2 in the furnace atmosphere.

Table 7

LLOY MELTED IN A HIGH-FREQUENCY INDUCTION OR GAS-FIRED FURNACE WITH A SILICA OR CLAY-GRAPHITE LINING USING THE C_1 METHOD OR THE C_2 METHOD OF CHARGING AS INDICATED

l — Elongation, per cent			Modified Crown Type — Tensile Strength, psi.			Modified Crown Type — Elongation, per cent			Modified Webbert Type — Tensile Strength, psi.			Modified Webbert Type — Elongation, per cent			Canadian Universal — Tensile Strength, psi.			Canadian Universal — Elongation, per cent			Average — All Designs — Tensile Strength, psi.			Average — All Designs — Elongation, per cent		
Min.	*Avg.*	*Max.*	*Min.*	*Avg.*	*Max.*	*Min.*	*Avg.*	*Max.*	*Min.*	*Avg.*	*Max.*	*Min.*	*Avg.*	*Max.*	*Min.*	*Avg.*	*Max.*	*Min.*	*Avg.*	*Max.*	*Min.*	*Avg.*	*Max.*	*Min.*	*Avg.*	*Max.*
32.0	35.0	38.0	33,800	35,925	37,800	25.2	31.8	36.0	35,600	36,175	36,700	29.1	35.3	40.7	32,800	33,025	33,200	32.0	33.5	34.2	30,000	35,195	38,800	17.5	32.3	40.7
43.8	44.2	44.5	35,100	36,100	37,000	29.1	33.7	36.8	35,800	36,300	36,500	40.2	42.4	45.5	31,700	32,250	33,200	28.3	34.8	36.3	31,700	35,825	38,800	28.3	38.3	45.5
	39.6			36,012			32.8			36,238			38.9			32,638			34.2			35,510			35.3	
41.3	41.6	41.8	34,000	35,025	36,000	28.2	31.3	32.5	35,500	36,075	36,500	38.5	40.5	42.5	32,200	33,125	33,800	33.7	36.1	38.0	32,200	35,472	39,300	27.3	36.3	42.5
39.0	39.2	39.3	35,200	35,925	36,100	31.4	33.3	36.5	35,200	35,800	36,200	38.3	40.2	41.8	31,700	32,175	32,900	27.6	31.4	34.4	31,700	35,235	38,400	23.6	34.2	41.8
	40.4			35,475			32.3			35,938			40.4			32,650			33.8			35,354			35.2	
34.8	38.0	41.3	35,200	35,425	35,800	28.4	30.0	32.4	35,800	36,075	36,300	34.5	37.9	40.4	31,400	32,400	33,300	25.0	28.9	34.2	31,400	35,277	38,100	25.0	34.0	41.3
31.7	33.8	36.0	30,300	31,200	32,000	22.3	25.4	29.0	31,700	33,700	35,000	33.4	35.3	36.6	30,300	30,650	31,400	26.1	28.4	31.8	29,900	32,565	35,800	22.3	30.2	36.6
	35.9			33,312			27.7			34,888			36.6			31,525			28.6			33,921			32.1	
40.8	40.8	40.8	37,000	37,325	37,600	32.8	37.4	39.9	35,200	36,100	36,700	32.1	39.9	44.6	34,600	34,975	36,000	31.7	38.0	42.7	34,600	36,620	38,200	29.4	38.7	44.6
26.3	34.3	42.3	36,200	37,100	38,000	30.5	33.6	36.8	36,000	36,350	36,600	43.1	43.3	43.7	34,000	34,525	35,400	33.0	35.0	36.2	34,000	36,135	38,500	26.3	36.1	43.7
	37.6			37,212			35.5			36,225			41.6			34,750			36.5			36,378			37.4	
26.7	35.1	43.5	36,600	37,575	38,400	36.2	39.2	43.6	35,900	36,350	36,900	37.6	41.2	43.2	33,300	34,075	34,600	31.9	36.9	41.3	33,300	36,315	39,000	25.2	37.6	43.8
20.2	29.4	38.6	37,000	37,650	38,200	36.1	39.1	41.3	34,800	36,050	37,400	32.5	38.7	42.1	34,100	34,825	36,900	35.0	36.4	38.6	32,300	36,010	38,200	20.2	36.5	43.8
	32.2			37,612			39.2			36,200			40.0			34,450			36.6			36,162			37.0	
33.0	36.4	39.7	36,400	37,275	37,900	28.2	33.5	38.4	35,000	36,250	37,300	30.7	36.5	41.4	31,300	33,525	34,400	23.1	32.2	36.2	31,300	35,965	38,200	23.1	34.5	42.8
22.4	32.3	42.2	30,200	34,100	36,600	20.2	29.1	35.8	34,500	35,200	36,500	33.4	39.5	43.9	31,600	32,400	33,500	31.5	33.7	37.8	31,600	34,325	38,200	20.2	33.6	43.9
	34.4			35,688			31.3			35,725			38.0			32,962			33.0			35,145			34.0	
24.4	33.6	42.8	37,800	38,700	39,500	30.3	33.7	36.6	36,800	38,025	39,300	33.5	37.0	42.6	34,600	35,800	36,900	27.6	31.6	34.6	34,600	37,570	39,500	24.4	35.0	42.8
22.5	35.2	48.0	38,000	38,325	38,800	29.6	32.0	34.0	36,500	37,550	38,800	30.2	35.0	41.8	34,400	34,500	34,600	28.1	29.2	30.2	34,400	37,190	39,700	22.5	34.2	48.0
	34.4			38,512			32.8			37,788			36.0			35,150			30.4			37,380			34.6	
17.6	18.6	19.5	35,900	36,075	36,300	23.7	24.9	26.2	38,200	38,250	38,300	37.5	38.0	38.6	31,200	32,575	33,400	20.8	22.9	25.6	31,200	35,320	38,700	17.6	27.4	38.6
24.1	34.8	45.4	36,500	37,100	37,800	24.6	28.0	31.4	36,500	37,400	37,800	31.2	36.0	46.0	30,600	33,525	35,600	18.6	30.7	41.3	30,600	36,925	40,100	18.6	34.1	46.0
	26.7			36,588			26.4			37,825			37.0			33,050			26.8			36,122			30.8	
19.4	22.7	26.0	36,900	37,775	39,400	25.5	29.8	35.8	36,600	37,925	38,800	33.4	35.4	37.2	30,200	32,833	35,600	18.0	27.8	44.2	30,200	36,407	40,500	18.0	31.0	44.2
33.6	34.0	34.3	36,900	37,575	38,300	27.6	33.3	40.5	36,200	37,050	37,900	34.6	37.9	40.2	28,100	32,225	34,300	19.5	31.1	42.1	28,100	35,810	38,300	19.5	33.9	43.0
	28.4			37,675			31.6			37,488			36.6			32,529			29.4			36,108			32.4	

Table 8

TENSILE PROPERTIES OF SIX DESIGNS OF TEST BARS POURED FROM 85-5-5-5 ALLOY MELTED IN A HIGH-FREQUENCY INDUCTION FURNACE WITH A SILICA OR CLAY-GRAPHITE LINING, A GAS-FIRED TILTING FUR
LINING, USING THE C_1 METHOD OR THE C_2 METHOD OF CHARGING AS INDICATED

Furnace	Lining	Charge	Pouring Range, ° F.	Horizontal ⅝-In. Web-Webbert — Tensile Strength, psi. Min.	Avg.	Max.	Elongation, per cent Min.	Avg.	Max.	L.T.B.-2 — Tensile Strength, psi. Min.	Avg.	Max.	Elongation, per cent Min.	Avg.	Max.	Modified ⅝-In. Web-Webbert — Tensile Strength, psi. Min.	Avg.	Max.	Elongation, per cent Min.	Avg.	Max.	Navy ⅝-In. Web-Webbert — Tensile Strength, psi. Min.	Avg.	Max.	Elongation, per cent Min.	Avg.	Max.	Tensile Str Min.	Avg
F_1	L_1	C_1	2250-2220	35,900	36,867	37,800	35.7	40.1	43.5	36,700	37,783	38,200	30.8	39.0	43.5	36,300	36,575[4]	37,000	40.8	42.9[4]	45.0	31,800	34,420[5]	35,800	25.4	30.6[5]	35.2	35,400	36,5
F_1	L_1	C_2	"	35,400	36,180[5]	36,800	36.7	40.5[5]	44.1	36,300	37,300[5]	38,400	31.3	38.1[5]	44.4	36,000	36,883	37,200	39.2	42.9	46.3	33,100	34,217	35,100	29.1	29.8	32.8	34,900	36,2
		Average	"		36,555			40.3			37,563			38.6			36,760			42.9			34,309			30.2			36,3
F_1	L_2	C_1	"	34,900	36,400[5]	37,300	33.6	37.9[5]	42.4	36,000	37,583	39,200	34.2	38.6	43.2	36,000	36,940[5]	38,300	33.6	39.3[5]	45.4	34,100	34,560[5]	35,200	28.2	29.5[5]	31.2	34,400	36,5
F_1	L_2	C_2	"	30,800	34,266[3]	37,300	20.5	29.7[3]	38.4	32,800	35,283	37,500	22.3	28.6	37.1	36,100	36,933	37,800	37.0	41.7	44.4	33,600	34,800[4]	35,500	28.6	30.7[4]	33.2	30,400	34,8
		Average	"		35,600			34.8			36,433			33.6			36,936			40.6			34,667			30.0			35,6
F_2	L_2	C_1	"	29,900	33,340[5]	35,600	18.8	30.2[5]	40.0	31,500	34,933	37,600	21.2	30.8	39.5	35,000	36,383	37,000	35.0	41.0	45.7	30,600	33,450	35,600	21.2	29.3	36.8	32,100	35,3
F_2	L_2	C_2	"	34,400	35,133[3]	35,700	33.5	35.6[3]	37.0	32,000	35,350	37,400	18.9	31.4	41.4	36,200	36,720[5]	37,000	40.0	41.0[5]	42.1	30,200	32,417	33,800	20.4	26.0	29.6	29,400	32,1
		Average	"		34,012			32.2			35,142			31.1			36,536			41.0			32,934			27.7			33,7
F_1	L_2	C_1*	"	16,300	21,833	26,000	2.0	8.9	15.0	23,800	27,150	31,200	11.0	15.8	21.6	28,200	31,633	33,100	16.0	23.8	28.0	22,300	25,783	30,100	11.5	15.2	18.4	19,000	22,9
F_2	L_2	C_2**	"	32,500	35,183	37,100	23.8	31.9	38.6	33,700	36,325[4]	38,000	22.2	31.5[4]	36.6	36,200	36,917	37,900	36.0	41.5	46.8	33,400	35,080[5]	36,800	27.2	31.8[5]	36.6	33,700	35,9
F_1	L_1	C_1(T)	"	34,900	36,250	37,200	33.5	38.7	42.7	36,000	37,033	38,300	30.5	36.8	44.0	36,300	36,908	37,600	40.1	44.3	48.2	33,000	34,350	37,300	26.0	29.3	32.0	35,200	36,6
F_1	L_1	C_2(T)	"	36,000	36,600	37,300	38.6	40.7	44.6	28,500	34,767	38,300	14.5	30.1	44.8	35,600	36,633	37,700	34.1	40.7	46.7	32,200	34,133	36,600	22.6	29.6	36.7	31,100	35,7
		Average	"		36,425			39.7			35,900			33.4			36,766			42.5			34,242			29.5			36,1
F_1	L_1	C_1	2150-2120	36,700	37,050[5]	37,400	41.0	42.0[5]	44.0	34,900	37,433	38,600	20.9	35.4	43.0	34,800	36,633	38,500	33.0	38.0	44.8	32,500	34,933	35,800	21.6	29.8	34.1	36,400	37,5
F_1	L_1	C_2	"	36,600	37,467	38,000	37.8	43.7	51.0	32,500	36,420	38,700	19.2	33.9	45.0	36,200	36,875[4]	37,300	35.9	41.5[4]	46.2	32,300	35,160[5]	36,900	19.8	31.4[5]	41.4	36,500	37,7
		Average	"		37,277			42.9			36,972			34.7			36,730			39.4			35,036			30.5			37,6
F_1	L_2	C_1	"	36,400	37,517	38,300	32.0	41.2	49.6	33,400	37,300[5]	39,300	21.6	32.5[5]	45.0	35,300	36,760[5]	37,300	29.6	37.6[5]	42.4	35,400	36,367	37,700	27.6	31.7	35.3	33,900	36,4
F_1	L_2	C_2	"	36,700	37,333[3]	38,100	37.1	42.9[3]	47.2	34,300	38,183	39,700	22.5	40.7	50.5	36,500	36,933[3]	37,300	37.6	39.1[3]	40.2	34,600	36,817	38,900	28.4	33.5	37.4	36,300	37,6
		Average	"		37,456			41.8			37,782			38.6			36,825			38.2			36,592			32.6			37,0
F_2	L_2	C_1	"	33,300	36,420[5]	37,700	26.8	39.1[5]	44.6	36,600	37,583	38,500	31.4	36.0	42.4	35,400	36,780[5]	37,400	37.3	42.9[5]	48.0	33,300	34,500	35,400	24.8	28.2	31.6	34,400	36,4
F_2	L_2	C_2	"	34,800	36,920[5]	37,700	30.5	41.4[5]	46.3	33,300	36,980[5]	39,300	18.7	35.0[5]	44.0	35,900	36,567	37,200	30.5	38.7	43.0	31,900	34,160[5]	35,600	26.4	29.6[5]	33.5	34,700	35,3
		Average	"		36,670			40.2			37,309			35.5			36,664			40.6			34,345			28.8			36,1
F_1	L_2	C_1*	"	24,200	29,483	33,600	11.1	19.7	25.9	20,000	30,233	34,100	5.0	19.0	26.7	29,200	33,267	36,000	12.2	26.4	34.8	26,400	30,117	32,200	13.0	20.5	25.6	22,300	26,1
F_2	L_2	C_2**	"	36,900	37,480[5]	37,900	39.3	42.7[5]	44.6	33,200	37,000	39,600	20.5	36.4	47.3	37,100	37,540[5]	38,200	41.4	44.8[5]	47.7	35,400	36,280[5]	37,400	31.6	33.3[5]	36.8	34,700	37,0
F_1	L_1	C_1(T)	"	36,800	37,383	38,000	34.5	41.2	43.5	37,500	38,533	39,300	32.8	40.0	49.1	35,700	36,800	37,800	36.1	40.8	47.2	33,700	35,867	38,600	24.8	31.4	38.7	37,400	37,7
F_1	L_1	C_2(T)	"	32,000	36,367	38,000	20.5	40.4	50.2	37,600	38,483	39,800	35.3	41.4	45.0	36,100	37,020[6]	37,600	30.3	39.1[6]	43.0	31,800	34,950	36,800	21.5	29.3	37.7	36,700	38,2
		Average	"		36,875			40.8			38,508			40.7			36,900			40.0			35,408			30.4			38,0
F_1	L_1	C_1	2050-2020	38,100	38,760[5]	39,200	36.6	38.4[5]	40.4	35,700	37,300	38,000	26.1	32.1	10.0	38,100	38,725[4]	39,600	36.8	38.5[4]	41.0	37,100	37,850	38,500	29.6	32.0	34.6	39,100	39,2
F_1	L_1	C_2	"	38,000	38,750[4]	39,500	35.3	40.2[4]	44.4	36,800	38,220[5]	39,400	29.0	36.4[5]	43.3	38,400	38,950[3]	39,500	35.9	38.7[3]	41.5	36,900	38,000[5]	39,000	27.3	31.9[5]	34.9	39,100	39,4
		Average	"		38,756			39.2			37,718			34.1			38,800			38.6			37,918			32.0			39,3
F_1	L_2	C_1	"	37,800	39,033	39,900	32.8	37.1	40.0	34,200	37,233	39,700	22.2	29.5	43.1	36,100	38,233[3]	39,400	25.6	34.4[3]	40.2	37,800	38,750	39,600	31.8	33.1	34.6	39,500	39,8
F_1	L_1	C_1	"	37,100	38,333	39,900	32.0	35.2	38.4	35,200	37,133	38,600	26.4	29.9	34.2	36,800	38,160[5]	39,000	33.0	33.1[5]	34.4	34,200	37,767	40,100	24.4	32.8	38.8	38,500	39,6
		Average	"		38,683			36.2			37,183			29.8			38,187			33.6			38,259			32.9			39,7
F_2	L_2	C_1	2050-2020	38,200	38,983	40,000	33.2	39.4	43.7	29,200	34,500[5]	37,900	10.0	22.5[5]	29.7	36,900	37,700[5]	38,800	33.2	34.9[5]	38.2	36,300	37,733	39,600	27.4	30.6	37.3	37,800	39,4
F_2	L_2	C_2	"	34,900	37,450[4]	39,300	26.0	33.4[4]	40.6	37,600	36,960[5]	37,200	26.2	28.7[5]	32.0	37,100	38,133	39,500	33.4	36.2	40.4	37,700	38,533	39,200	32.5	34.2	39.2	38,600	39,1
		Average	"		38,370			37.0			35,618			25.3			37,936			35.6			38,133			32.4			39,2
F_1	L_2	C_1*	"	30,200	32,067	33,400	16.3	19.6	23.9	28,500	30,367	33,200	11.8	16.1	20.7	35,400	36,183	37,100	25.2	27.9	30.4	31,300	33,400	37,100	14.0	20.3	25.6	27,600	30,9
F_2	L_2	C_2**	"	37,800	39,517	41,000	32.6	39.7	44.6	36,800	38,217	39,900	25.3	32.0	35.0	36,800	38,817	41,300	29.6	37.3	41.1	35,800	37,350	38,500	25.6	30.0	32.0	34,900	38,8
F_1	L_1	C_1(T)	"	38,100	39,183	39,900	35.8	39.7	43.2	35,300	36,925[4]	38,400	23.6	29.8[4]	39.0	36,900	38,433[3]	40,000	30.8	35.1[3]	37.4	28,100	38,850[4]	39,700	33.1	35.2[4]	39.0	38,200	39,1
F_1	L_1	C_2(T)	"	37,100	38,580[5]	40,400	32.7	38.1[5]	44.6	34,700	36,800	38,200	22.0	28.4	33.3	36,800	37,600[5]	38,700	28.2	30.7[5]	37.1	37,300	38,383	39,300	30.1	33.5	38.5	38,200	39,1
		Average	"		38,909			39.0			36,850			29.0			37,912			32.4			38,570			34.2			39,1

(T) 5 per cent turnings in the charge.
* Glass charged with the metal to maintain a glass-slag cover at all times.
** Glass added just prior to melt-down to form a glass slag and prevent excessive oxidation of the melt.
Note: Gas furnaces were operated to produce approximately 0.5 per cent O_2 in the furnace atmosphere.
[3] Average of 3 bars.
[4] Average of 4 bars.
[5] Average of 5 bars.
Note: No exponent indicates the average of 6 bars.

E LINING, A GAS-FIRED TILTING FURNACE HAVING A CLAY-GRAPHITE CRUCIBLE OR A GAS-FIRED ROTATING FURNACE WITH A FIREBRICK
HARGING AS INDICATED

-In. Web-Webbert				Navy $\frac{1}{16}$-In. Web-Webbert						Navy $\frac{5}{16}$-In. Web-Crown						Average — All Designs					
	Elongation, per cent			Tensile Strength, psi.			Elongation, per cent			Tensile Strength, psi.			Elongation, per cent			Tensile Strength, psi.			Elongation, per cent		
Max.	Min.	Avg.	Max.	Min.	Avg.	Max.	Min.	Avg.	Max.	Min.	Avg.	Max.	Min.	Avg.	Max.	Min.	Avg.	Max.	Min.	Avg.	Max.
6,800	25.4	30.6[5]	35.2	35,400	36,517	37,200	35.0	38.2	41.2	35,400	36,883	37,600	28.3	36.0	39.8	31,800	36,567	38,200	25.4	37.7	45.0
6,100	29.1	29.8	32.8	34,900	36,217	37,200	32.6	39.0	46.7	36,700	37,317	38,000	35.1	37.1	39.6	33,100	36,259	38,400	29.1	37.8	46.3
		30.2			36,367			38.6			37,100			36.6			36,439			37.8	
5,200	28.2	29.5[5]	31.2	34,400	36,580[5]	37,800	25.4	36.3[5]	42.0	32,000	35,133	37,800	20.4	28.3	37.4	32,000	36,209	39,200	20.4	34.9	45.4
5,500	28.6	30.7[4]	33.2	30,400	34,816	36,800	18.4	29.8	37.0	30,300	34,500	36,800	17.5	26.7	32.6	30,300	35,200	37,800	17.5	31.4	44.4
		30.0			35,618			32.8			34,817			27.5			35,712			33.2	
5,600	21.2	29.3	36.8	32,100	35,333	37,600	20.3	29.3	39.8	31,600	33,067	34,400	20.4	25.1	32.0	29,900	34,448	37,600	18.8	31.0	45.7
3,800	20.4	26.0	29.6	29,400	32,183	35,400	15.6	25.2	35.7	32,000	34,100	35,300	22.1	27.2	32.6	28,900	34,166	37,400	15.6	30.3	42.1
		27.7			33,758			27.2			35,584			26.2			34,314			30.6	
0,100	11.5	15.2	18.4	19,000	22,917	26,300	6.0	9.4	13.2	23,800	27,400	29,400	11.3	16.2	20.5	16,300	26,119	33,100	2.0	14.9	28.0
6,800	27.2	31.8[5]	36.6	33,700	35,900[5]	37,900	26.0	35.8[5]	41.2	30,800	34,050[4]	37,600	20.9	27.9[4]	40.2	30,800	35,633	38,000	20.9	33.9	46.8
7,300	26.0	29.3	32.0	35,200	36,633	37,800	29.0	37.2	44.4	35,000	37,233	38,000	27.6	37.4	41.1	33,000	36,401	38,300	26.0	37.3	48.2
6,600	22.6	29.6	36.7	31,100	35,750	37,400	33.6	38.8	42.7	36,900	37,350	38,000	36.8	37.8	39.1	28,800	35,872	38,300	14.5	36.8	46.7
		29.5			36,192			38.0			37,292			37.6			36,136			36.8	
5,800	21.6	29.8	34.1	36,400	37,560[5]	38,700	33.0	37.1[5]	45.6	37,400	38,483	39,200	30.2	38.5	42.8	32,500	36,998	39,200	20.9	36.6	45.6
6,900	19.8	31.4[5]	41.4	36,500	37,720[5]	38,300	32.7	37.9[5]	40.4	37,600	38,340[5]	39,100	32.2	37.4[5]	42.1	32,300	37,017	39,100	19.2	37.7	51.0
		30.5			37,640			37.5			38,418			38.0			37,007			37.1	
7,700	27.6	31.7	35.3	33,900	36,460[5]	38,600	21.0	30.9[5]	43.2	36,200	37,300[5]	37,900	26.4	32.6[5]	38.8	33,400	36,950	39,300	21.0	35.1	49.6
8,900	28.4	33.5	37.4	36,300	37,640[5]	38,900	31.6	36.7[5]	47.6	35,400	36,400[5]	38,000	25.6	29.1[5]	34.0	34,300	37,250	39,700	22.5	36.4	50.5
		32.6			37,050			33.8			36,850			30.9			37,090			35.7	
5,400	24.8	28.2	31.6	34,400	36,483	37,900	23.5	36.4	43.7	36,300	37,600	38,500	28.8	36.0	41.1	33,300	36,556	38,500	23.5	35.9	48.0
5,600	26.4	29.6[5]	33.5	34,700	35,367[3]	36,500	26.0	28.2[3]	31.3	34,900	36,050	37,900	25.2	30.0	37.3	31,900	36,070	39,300	18.7	34.2	46.3
		28.8			36,111			33.7			36,825			32.6			36,313			35.1	
2,200	13.0	20.5	25.6	22,300	26,133	31,100	7.5	12.7	19.8	26,300	30,317	33,100	12.6	18.3	22.4	20,000	29,925	36,000	5.0	19.2	34.8
7,400	31.6	33.3[5]	36.8	34,700	37,075[4]	39,000	29.2	36.0[4]	44.2	31,000	34,880[5]	36,500	18.6	27.2[5]	32.0	31,000	36,707	39,600	18.6	36.7	47.7
8,600	24.8	31.4	38.7	37,400	37,780[5]	38,200	34.5	38.0[5]	42.2	33,400	37,250	39,400	21.9	36.7	45.5	33,400	37,254	39,400	21.9	38.0	49.1
6,800	21.5	29.3	37.7	36,700	38,220[5]	38,800	25.6	36.9[5]	41.7	37,300	38,583	39,600	35.1	41.5	46.0	31,800	37,259	39,800	20.5	38.1	50.2
		30.4			38,000			37.5			37,917			39.1			37,205			38.1	
8,500	29.6	32.0	34.6	39,100	39,233[3]	39,400	34.8	36.2[3]	37.2	35,400	38,417	39,200	31.6	32.8	33.8	35,400	38,260	39,600	26.1	34.5	41.0
9,000	27.3	31.9[5]	34.9	39,100	39,400[5]	39,900	29.7	35.2[5]	42.6	37,400	38,560[5]	39,400	25.9	31.6[5]	34.5	36,800	38,608	39,900	25.9	35.1	44.4
		32.0			39,337			35.6			38,482			32.3			38,421			34.8	
9,600	31.8	33.1	34.6	39,500	39,800[5]	40,100	33.8	35.5[5]	38.0	39,000	39,150[4]	39,500	32.8	34.8[4]	36.8	34,200	38,680	40,100	22.2	33.9	43.1
0,100	24.4	32.8	38.8	38,500	39,680[5]	41,000	31.5	34.9[5]	41.2	36,700	38,017	39,400	24.5	29.3	34.4	34,200	38,138	41,000	24.4	32.4	41.2
		32.9			39,740			35.2			38,470			31.5			38,392			33.2	
9,600	27.4	30.6	37.3	37,800	39,400	40,000	23.3	33.5	39.9	33,300	36,300[5]	39,000	19.5	27.0[5]	38.2	29,200	37,462	40,000	10.0	31.3	43.7
9,200	32.5	34.2	39.2	38,600	39,117	39,800	30.8	32.9	35.7	36,500	37,775[4]	39,300	23.5	27.7[4]	33.3	34,900	38,077	39,800	23.5	32.5	40.6
		32.4			39,259			33.2			36,956			27.2			37,755			31.9	
7,100	14.0	20.3	25.6	27,600	30,917	35,000	11.2	15.8	23.6	28,900	31,800	34,800	15.3	19.8	24.0	27,600	32,220	37,100	5.0	19.9	30.4
8,500	25.6	30.0	32.0	34,900	38,850[4]	41,000	32.1	33.6[4]	36.4	37,800	38,733	39,700	28.2	31.4	33.6	34,900	38,565	41,300	25.3	34.0	44.6
9,700	33.1	35.2[4]	39.0	38,200	39,100[4]	39,800	29.5	33.0[4]	35.4	38,100	38,550	39,400	29.5	32.0	36.1	35,300	38,563	40,000	23.6	34.4	43.2
9,300	30.1	33.5	38.5	38,200	39,160[5]	40,200	32.1	35.8[5]	42.0	37,900	38,460[5]	39,100	31.3	33.0[5]	34.3	34,700	38,128	40,400	22.0	33.1	44.6
		34.2			39,133			34.6			38,509			32.5			38,327			33.7	

scrap on top, and the crucible is kept full until all of the charge is melted. This and the C_1 method represent the two extremes of the commonly used charging methods.

Method of Melt Preparation. All of the charges weighed about 300 to 350 lb. and consisted of 50 per cent ingot from a producer of secondary ingot metal and 50 per cent gates and risers from previous melts. In some heats, 5 per cent turnings were included in the charge which were melted down first, followed by the ingots, comprising 45 per cent of the charge, and the gates and risers, comprising 50 per cent of the charge. The use of 5 per cent turnings from the machine shop in the charge is indicated by the letter (T).

Thus, by using the code, the condition $F_1L_2C_1(T)$ indicates that the melt was prepared in a high-frequency induction furnace lined with a clay-graphite crucible, using the C_1 method of charging described above, and the initial charge consisted of 5 per cent by weight of turnings from the machine shop, along with 50 per cent gates and risers, and 45 per cent ingot.

When making the melts of the highest quality in the high-frequency induction or gas-fired tilt furnaces, the charges placed in the furnace by either the C_1 or C_2 methods, were melted down and the dross removed from the surface of the melt three times: (1) soon after the charge was completely molten, (2) when the melt had been heated to approximately 2150° F., and (3) just before transferring the melt when it had reached the temperature of approximately 2300° F. This procedure of melting these relatively small heats produced a metal loss amounting to about 3 per cent by weight.

The same procedure of melting was used in the gas-fired crucible tilt furnace, and in the high-frequency induction furnace, but when melting in the barrel-type rotating furnace, only the C_2 method of charging was practiced. The charge in this barrel-type furnace was covered with a glass slag when the charge melted and, of course, the melt was not skimmed until just before it was transferred to the pouring ladle.

When preparing certain melts where a low quality was desired, the high-frequency furnace with a relatively new clay-graphite lining was employed. Charging was done by the C_1 method, and before melting began, the metal was covered with a glass slag to prevent oxidation. This slag was maintained on the melt in the furnace as well as in the pouring ladle after transferring the charge to it. When a glass slag was used in the high-frequency induction furnace during the melting operation, this condition is indicated in the accompanying figures, illustrating the results obtained.

After the melts were prepared by the methods described above and the temperature reached 2300° F., they were then transferred to a preheated fire-clay-lined ladle which would hold the entire charge of 300 to 350 lb. As the melt was transferred, metallic zinc additions were made to restore the zinc content to approximately 5 per cent and 2 oz. of 15 per cent phosphor-copper per 100 lb. of melt was added to the pouring ladle. The melts were then poured into the molds from the various temperatures.

The complete tensile data on the 12 melts of Group 1 are listed in Table 7, while Table 8, pages 17 and 18, contains similar data on the 32 melts of Group 3.

Effect of Charging Method. The relative merits of the C_1 and C_2 methods of charging are illustrated by Figs. 19 to 25, inclusive. Each point on Figs. 19, 22, and 24 represent the average of 18 test-bar values; while each point on Figs. 20, 21, 23, and 25 represent the averages of 36 bars. Figures 19 and 20, prepared from data obtained from two separate but similar investigations, illustrate the relative merits of the two methods of charging into a high-frequency furnace lined with silica.

Figure 19 represents the results obtained from some of the first group of 12 melts prepared, while Fig. 20 represents the data obtained from the melts from the third group. Figure 21 shows the relative merits of the C_1 and C_2 methods of charging the same furnace, but 5 per cent turnings was included in the charge. For reasons described more fully later, the greatest weight is placed upon the data obtained at the two highest pouring temperatures, while those obtained at the lowest temperature are generally disregarded.

With this in mind, it is evident that Figs. 19 and 20 show the C_2 method of charging to be slightly superior to the C_1 method when using the high-frequency induction furnace without turnings in the charge. The superiority of the C_2 method is slight as further attested by the fact that the C_1 method appeared slightly the better when 5 per cent turnings was used in the charge.

Figures 22 and 23 illustrate the data from two separate, but similar investigations comprising melts from Groups 1 and 3, respectively. These

Table 6

EFFECT OF MOLD MATERIAL UPON THE TENSILE PROPERTIES OBTAINED IN THE HORIZONTAL 5/8-IN. WEB-WEBBERT TEST BAR[a]

Mold Condition	*Pouring Temp., °F.*	*Tensile Strength, psi.*			*Elongation, Per Cent*		
		Minimum	*Average*	*Maximum*	*Minimum*	*Average*	*Maximum*
Natural sand	2090-2180	37,200	38,600	40,100	36.8	42.6	51.5
1% oil, 1% sand baked at 450° F.	2085-2180	34,300	36,900	39,500	27.6	32.0	41.3
2% cereal baked at 450° F.	2095-2170	35,100	36,400	37,700	25.6	31.0	35.9
2% western bentonite plus 6.5% Revivo, not baked	2105-2160	37,700	38,100	38,500	31.4	33.5	35.6
2% western bentonite plus 6.5% Revivo, baked at 450° F.	2110-2145	33,400	36,250	37,800	22.6	31.4	38.1
2% western bentonite plus 6.5% Revivo, baked at 1700° F.	2110-2155	37,700	38,650	40,000	41.5	45.2	50.8

[a] A 350-lb. charge, consisting of 50 per cent gates and risers and 50 per cent new ingots of 85-5-5-5 alloy, was melted in silica-lined, high-frequency induction furnace. No cover was used. The melt was transferred to a pouring ladle after skimming. An addition of 1¼ per cent zinc, and 2 oz. of phosphor-copper per 100 lb. of melt was made during the transfer.

figures show the relative merits of the C_1 and C_2 methods of charging in a high-frequency induction furnace lined with clay-graphite. It will be noted that the C_1 method is somewhat superior to the C_2 method.

Figures 24 and 25 show data again from the two groups of melts, illustrating the relative merits of C_1 and C_2 methods of charging in the gas-fired tilt furnace, having a clay-graphite crucible. It will be noted that the C_1 method of charging is substantially superior to the C_2 method.

To summarize, the C_1 and C_2 methods produce similar results or possibly the C_2 method of charging is slightly the better of the two when melting in a high-frequency silica-lined induction furnace, with or without turnings in the charge. On the other hand, the C_1 method appears to be substantially the better of the two when charging the high-frequency clay-graphite lined induction furnace or the gas-fiilled tilting furnace having a clay-graphite crucible to hold the metal charge.

The data obtained from the melts from Group 1 and Group 3 are in reasonably good agreement, indicating that the employed method of rating melt quality is fairly consistent.

Since the test bars used to make these comparisons were poured under identical conditions, it is reasonable to presume the differences, small but consistent, are caused by differences in melt quality, i.e., gas content. Furthermore, the conclusions are based on the averages of many heats, and many test bars of various designs. It must be concluded then, that the C_2 charging method produces similar or possibly very slightly lower gas contents than the C_1 method of charging in the silica-lined, high-frequency induction furnace.

On the other hand, the C_1 method of charging is superior when melting in a clay-graphite lining, or crucible, in either a high-frequency induction furnace or a gas-fired tilt furnace. Possibly the carbon in contact with the melt is a source of the gas, and the greater exposure of the melt to oxidation when using the C_1 method helps to overcome this disadvantage. Thus, more strongly oxidizing conditions are required to produce the best melt quality. This hypothesis is, however, in the realm of speculation, but it is offered as a reasonable explanation of the results obtained.

The Effect of Furnace Design. Using a melting technique which would produce the best melt quality as described previously, the following four different furnaces were compared:

1. High-frequency induction furnace with a silica lining,
2. High-frequency induction furnace with a clay-graphite lining,
3. A gas-fired tilting-crucible furnace with a clay-graphite crucible,
4. A rocking, gas-fired barrel furnace lined with firebrick.

Only the first three furnaces were used to prepare the Group 1 melts in duplicate, using the C_1 and C_2

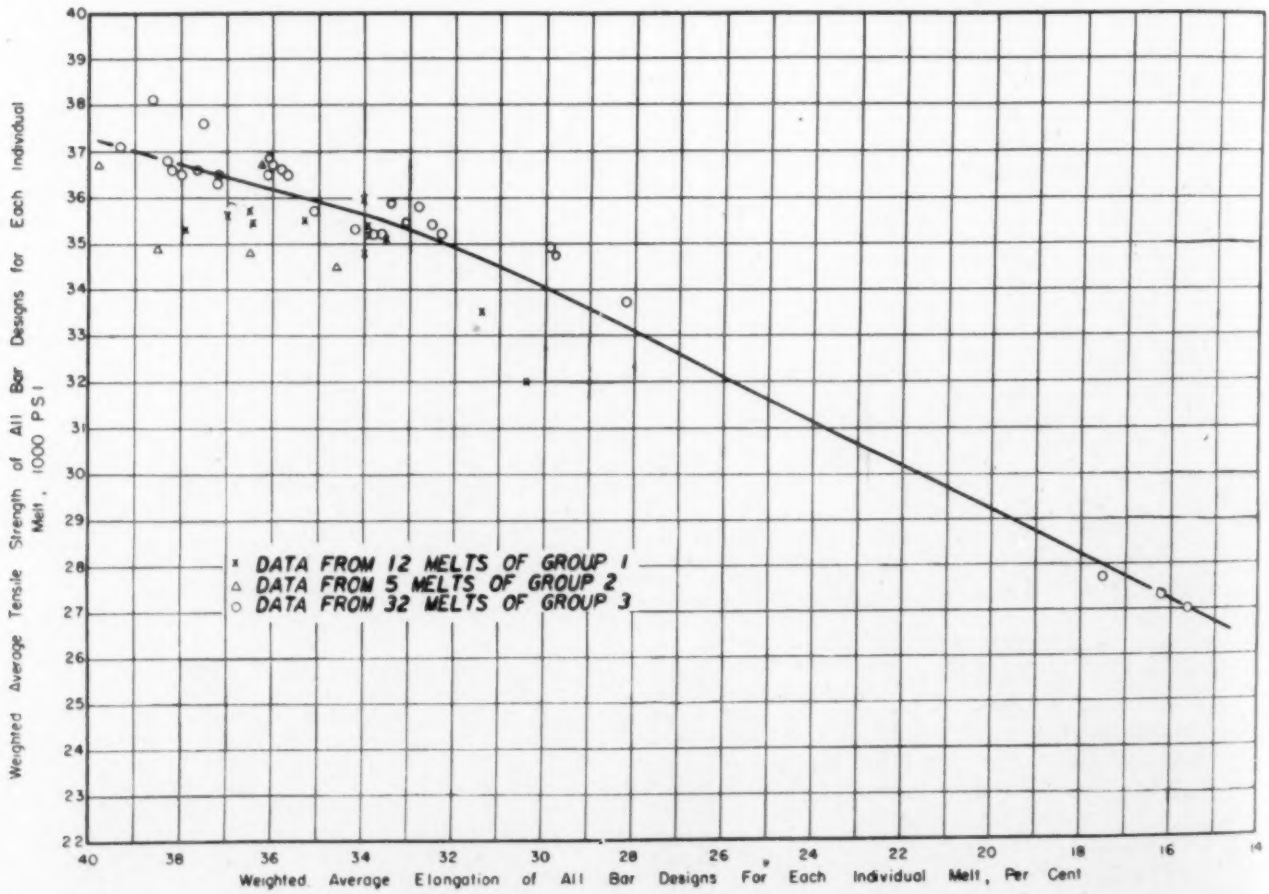


Fig. 16—Relationship between weighted average elongation and weighted average tensile strength of all test-bar designs investigated. Tensile properties at the highest pouring temperature were given a weight of 2; those at medium pouring temperatures a weight of 1; and those at lowest pouring temperature were discarded.

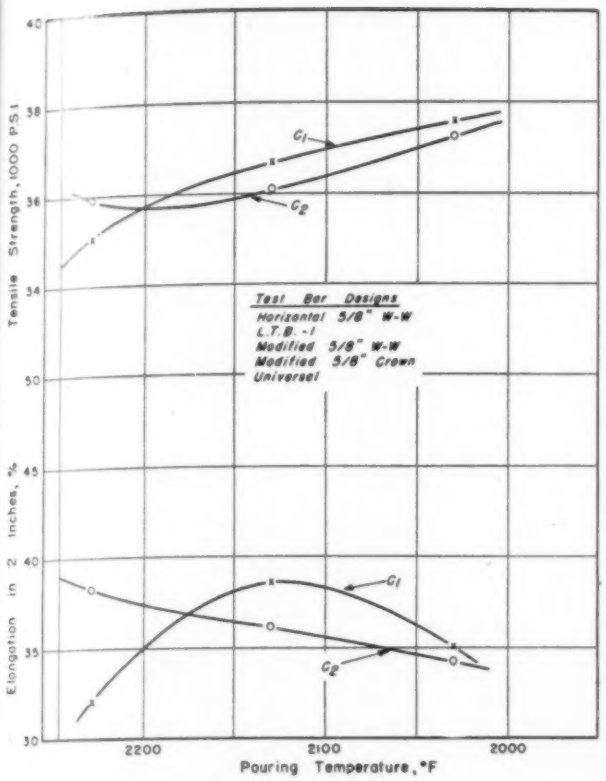


Fig. 19—Effect of charging method on tensiles of bars of 5 designs; high-frequency furnace with silica lining.

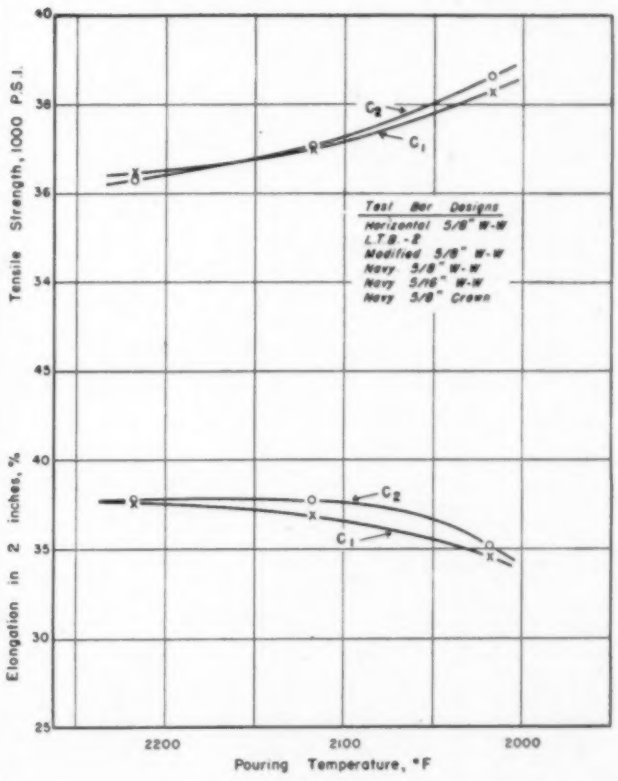


Fig. 20—Effect of charging method on tensiles of bars of 6 designs; high-frequency furnace with silica lining.

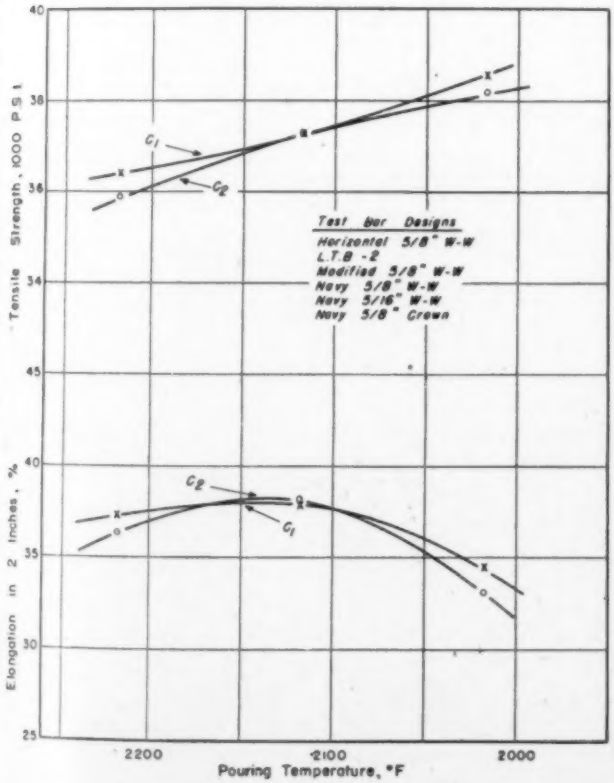


Fig. 21—Effect of charging method on tensiles of bars of 6 designs; high-frequency furnace, silica lining; 5 per cent turnings in charge.

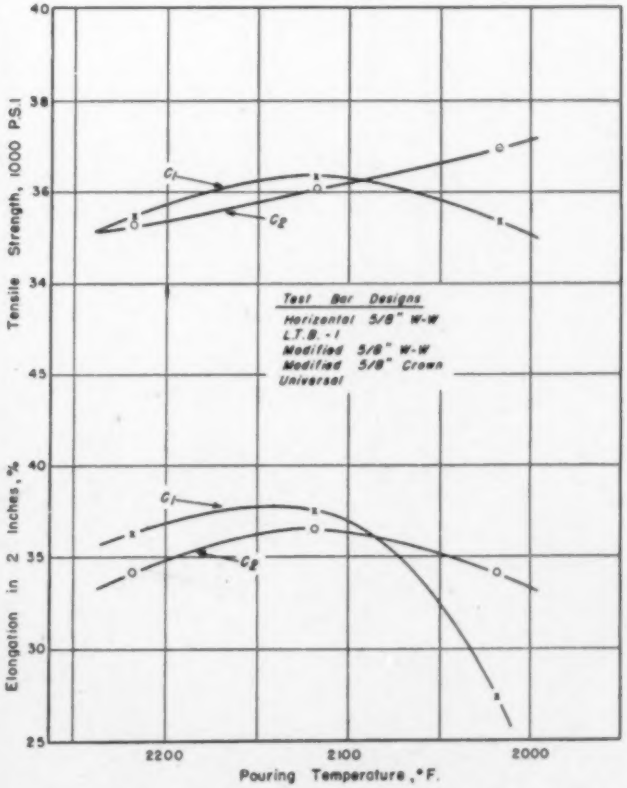


Fig. 22—Effect of charging method on tensiles of bars of 5 designs; high-frequency furnace with clay-graphite lining.

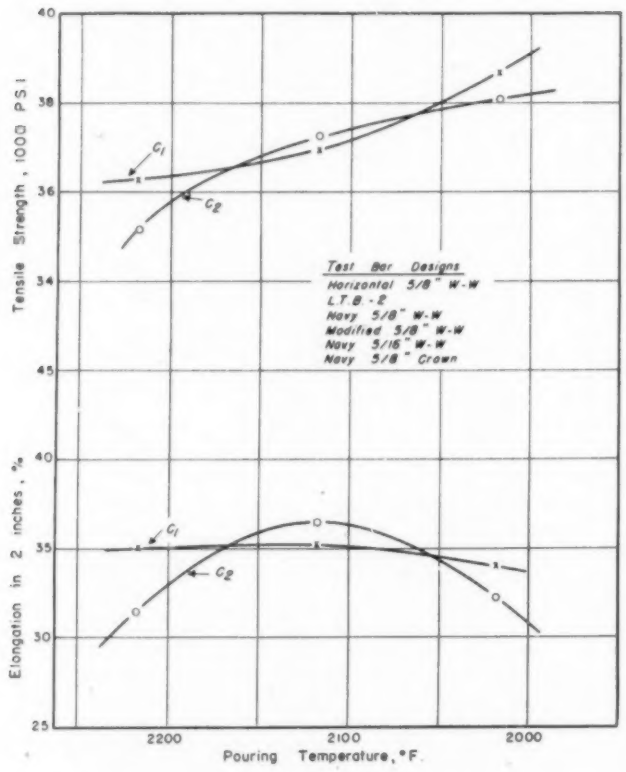


Fig. 23—Effect of charging method on tensiles of bars of 6 designs; high-frequency furnace with clay-graphite crucible; no turnings in charge.

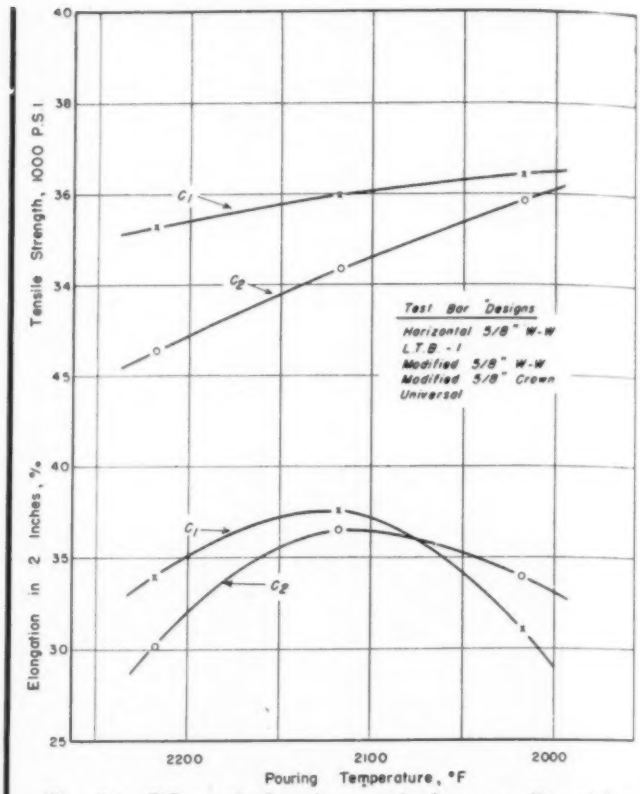


Fig. 24—Effect of charging method on tensiles of bars of 5 designs; gas-fired tilt furnace with clay-graphite crucible.

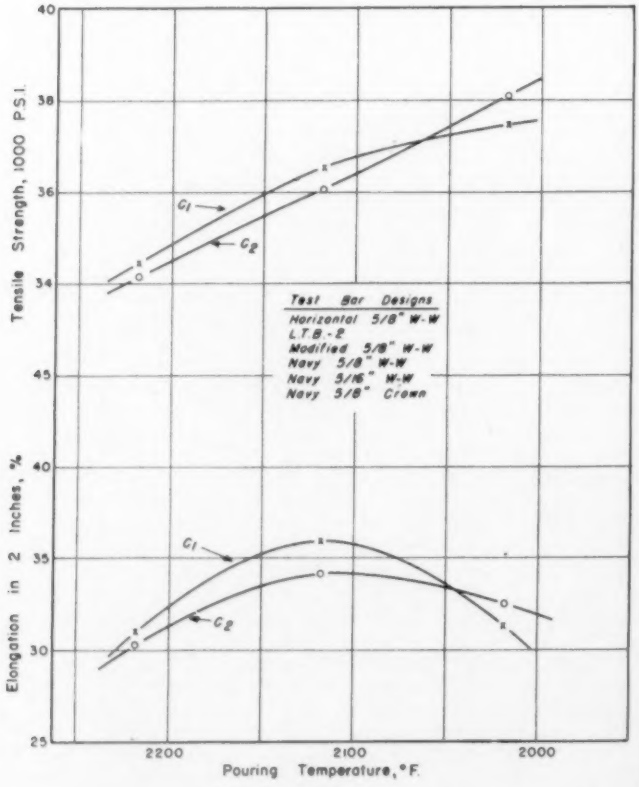


Fig. 25—Effect of charging method on tensiles of bars of 6 designs; gas-fired tilt furnace with clay-graphite crucible.

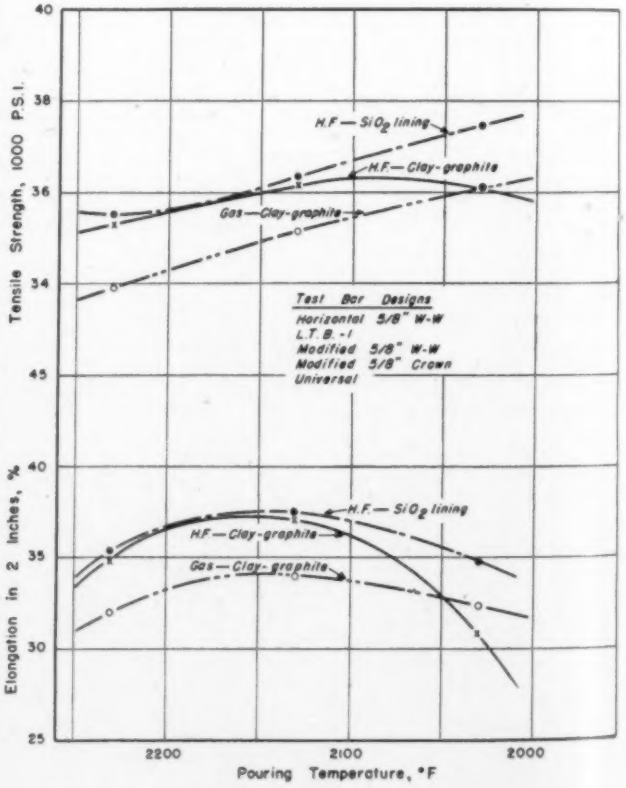


Fig. 26—Effect of furnace type and furnace lining on tensiles of bars of 5 designs; both charging methods used, duplicate melt averaged.

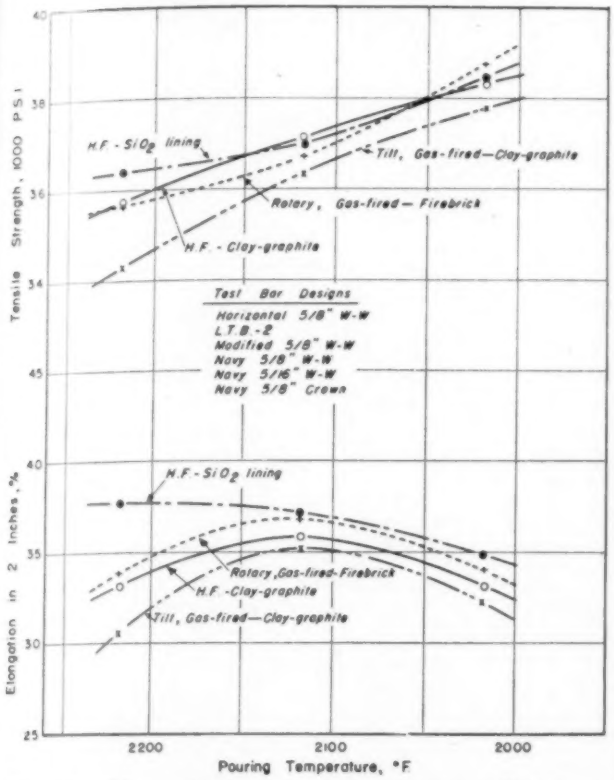


Fig. 27—Effect of furnace type and furnace lining on tensiles of bars of 6 designs; both charging methods used; no turnings in charge.

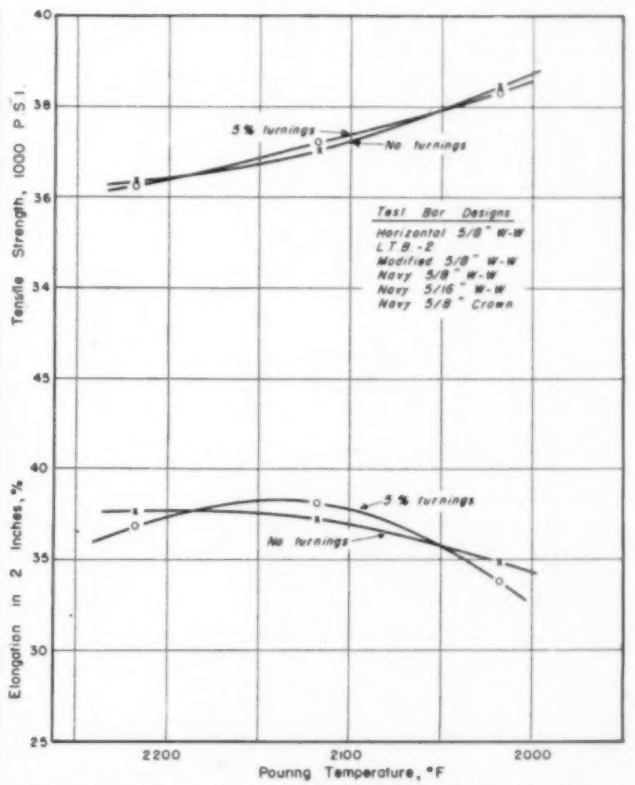


Fig. 28—Effect of charging method with and without turnings on tensiles of 6 test-bar designs; high-frequency furnace with silica lining.

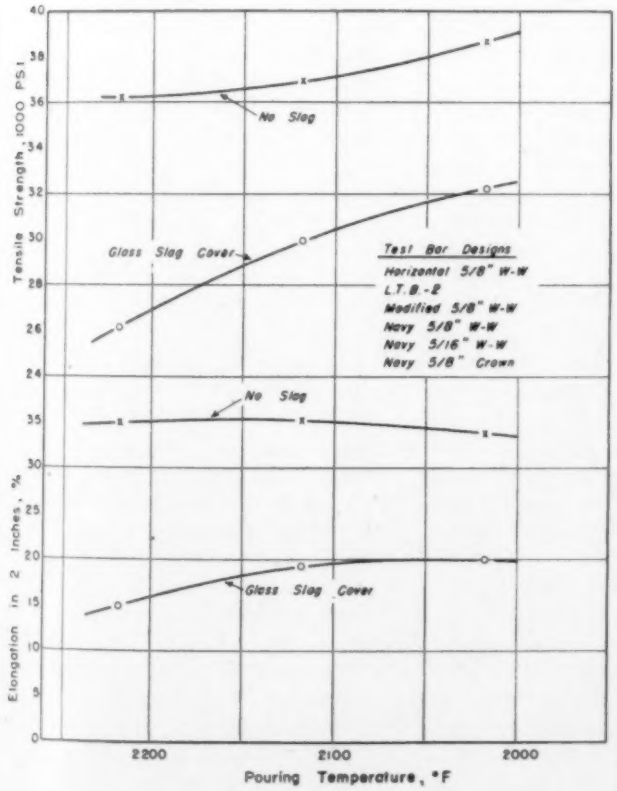


Fig. 29—Effect of glass slag cover on tensiles of 6 test-bar designs; high-frequency furnace with clay-graphite linings; no turnings used.

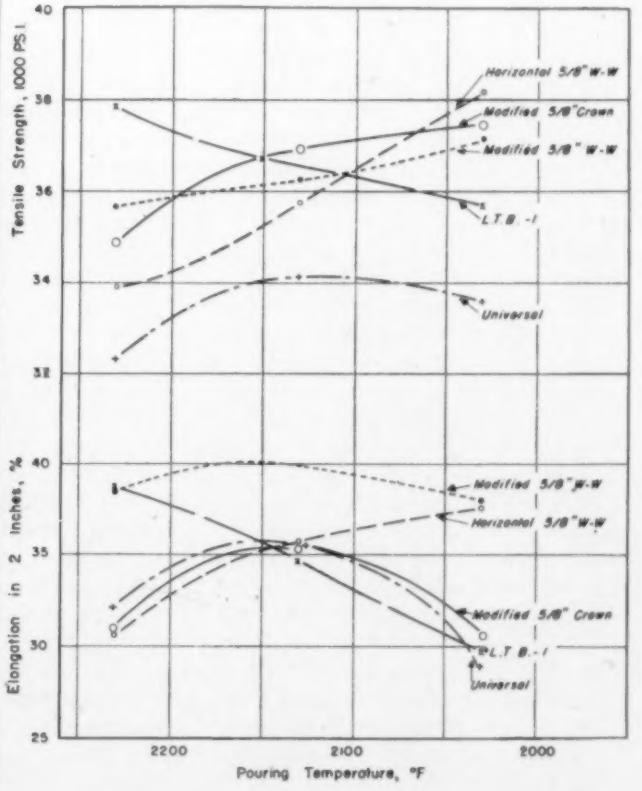


Fig. 30—Effect of test-bar design on test-bar properties; average values on 12 melts of Group 1, prepared under various conditions.

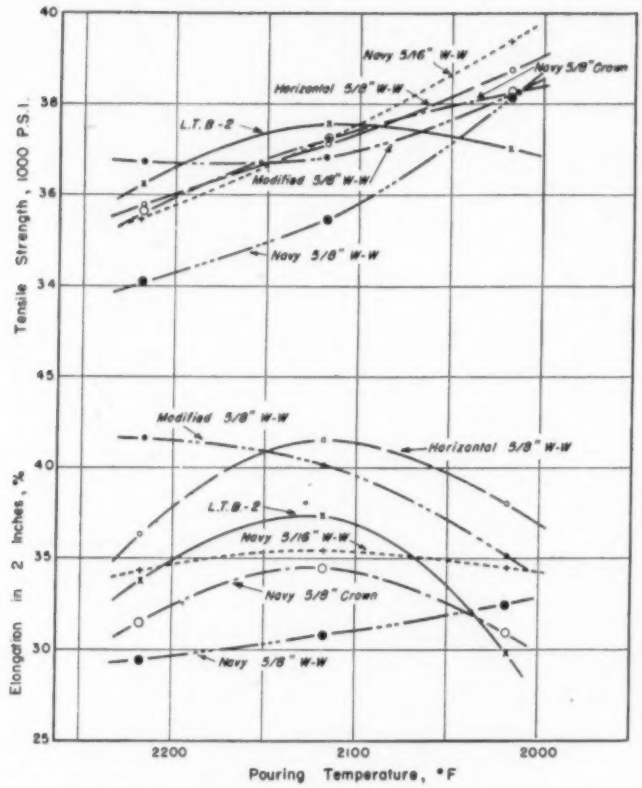


Fig. 31—Effect of test-bar design on tensile properties Each point represents average of 54 test bars from 27 melts of Group 3 prepared under various conditions. Melts prepared under a glass slag are excluded.

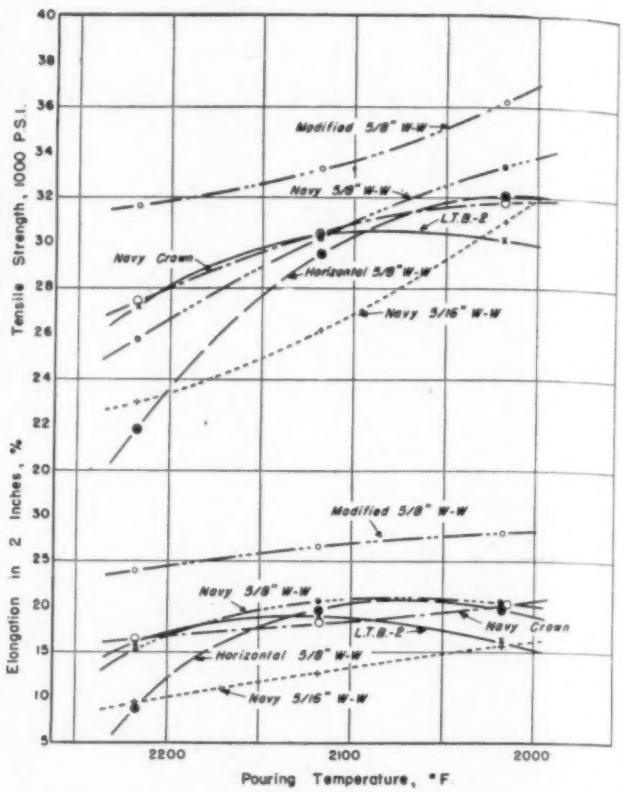


Fig. 32—Effect of test-bar design on tensile properties of bars cast from poor quality melts under a glass slag in a clay-graphite lined high-frequency induction furnace. Each point is average of 6 test bars from 3 melts.

methods of charging the furnace.

On the other hand, Group 3 melts were prepared in triplicate again using the C_1 and C_2 methods of charging, with the exception that the gas-fired barrel-type furnace was charged by the C_2 method only. The relative merits of these four furnaces are illustrated by Figs. 26 and 27. The results illustrated by Fig. 26 are from the first group of melts, while Fig. 27 represents the third group of melts, and each point represents the average of 36 and 72 test-bar values, respectively, excepting these representing condition $F_3L_2C_2$ on Fig. 27, which are the averages of 36 bars only.

Reference to the elongation values on Figs. 26 and 27 will show that the high-frequency induction furnace with a silica lining is somewhat superior to the same furnace with a clay-graphite lining. According to Fig. 27, the rotary gas-fired barrel-type furnace is slightly superior to the high-frequency furnace with a clay-graphite lining and therefore occupied second place. Both Figs. 26 and 27 show that the gas-fired tilt furnace with a clay-graphite crucible is substantially inferior to the other furnaces.

Since Figs. 26 and 27 are based on a very large amount of data, considerable confidence may be placed upon the results shown by them. It is quite possible, however, that relatively small differences in melting technique might change the relative order of these four furnace types. Thus, while the gas-fired tilting furnace with a clay-graphite crucible definitely appears inferior to the other type of furnaces, it is quite possible that this could be compensated, partly or entirely, by a melting technique which was somewhat different from that employed in these investigations.

In fact, the results illustrated by Figs. 26 and 27 are the averages obtained by both the C_1 and C_2 methods of charging. If only the C_1 method had been used, the high-frequency, clay-graphite-lined furnace would appear slightly better than it does, while the gas-fired tilt furnace having a clay-graphite crucible would appear substantially better than it does in Figs. 26 and 27.

Again, the relative merits of these furnaces must be translated into the relative gas contents of the melts produced in them. It is logical to speculate that the melting procedure which produces the best conditions of oxidation is superior. There is the further probability of carbon absorption from the clay-graphite crucible or lining with the subsequent evolution of gas by Reactions C or D described on page 3.

In view of the good results obtained with the gas-fired, firebrick-lined rotating furnace, this conjecture appears especially logical. Recalling that the gas-fired furnaces were operated at 0.5 per cent oxygen in the furnace atmosphere, it may be speculated that the silica-lined, high-frequency induction furnace was best because of the absence of carbon and the strongly oxidizing conditions. The same is true of the gas-fired, barrel-type rotating furnace. Although a glass cover was used, the furnace rotation produced good opportunities for oxygen absorption.

The clay-graphite-lined, high-frequency induction furnace occupied third place because, although conditions were strongly oxidizing, this

only partially offset the ill effects of carbon absorption. The gas-fired tilt furnace not only had the disadvantage of the contact of the melt with the carbon in the clay-graphite crucible, but the oxidizing conditions were less pronounced. Hence, this furnace appeared in fourth place. It should be emphasized that these are speculations only.

The Theory of Cases in Copper-Base Alloys. It may be well to point out here the theoretical reason why strongly oxidizing conditions produce the melts of highest quality, i.e., lowest gas content. The only limitation on oxidizing conditions is that, when excessively oxidizing melting conditions are employed, loss in alloy is quite marked. Not only do the zinc losses become excessive, but there will also be loss of tin and an over-all poor metal recovery.

Within this extreme, however, strongly oxidizing conditions produce a high oxygen content in solution in the melt, and it makes no practical difference whether this oxygen is present as atomic oxygen or as an oxide so long as it is in solution in the melt. The more of the readily oxidizable elements such as phosphorus or zinc present and the higher their concentration, the lower the maximum oxygen content which is in equilibrium with them in the melt, thus

$$2P+5\,O\ (or\ an\ oxide)=P_2O_5$$

$$K=\frac{C_{P_2O_5}}{C_P^2 C_O^5}$$

$$C_O^5=K'C_{P_2O_5}\times\frac{1}{C_P^2}$$

where K is a constant for a given temperature, and C is the concentration of the constituent indicated by the subscript. This equation, based on the law of mass action, shows that, as the concentration of phosphorus in the melt increases, there is a decrease in the maximum equilibrium oxygen or oxide content in solution in the melt. Similar relationships exist for the oxygen or oxide content in equilibrium with the zinc, tin, hydrogen, carbon, and other elements having an affinity for oxygen.

It is evident that the higher the content of strongly reducing elements, such as phosphorus and zinc, the lower the maximum soluble oxygen or oxide content in the melt, and consequently, the higher the maximum content of hydrogen, carbon, or other gases or gas formers in the melt. As a consequence of the higher soluble oxygen or oxide content which can be attained, the maximum concentration of hydrogen and other reducing gases or gas formers is less in phosphorus-free or zinc-free alloys, still less in binary copper-tin alloys, and still less in commercially pure copper melted under oxidizing conditions.

It will be evident then that, because of the zinc content of 85-5-5-5 alloy, the maximum oxygen content which the melt can contain in solution is relatively low. This condition produces a fairly narrow range in conditions whereby the oxygen content of the melt will be high enough without producing an excessive metal loss. If this condition is attained, the high oxygen or oxide content in solution in the melt will be high enough to keep the content of hydrogen and possibly other gases or gas formers low.

Furthermore, a phosphorus addition to the ladle during the transfer then reduces this equilibrium oxygen content, so that Reactions A, B, C, or D, shown on page 3, do not proceed to the right when the metal solidifies. The technique, then, is to produce low hydrogen or other gas or gas former contents by maintaining as high as practical oxygen content during melting, then reducing this oxygen with phosphorus before pouring. As a result of the low hydrogen and the low oxygen contents, they do not again reach equilibrium proportions even though they are concentrated in the remaining liquid during solidification.

The above brief account of the theory of gases in copper-base alloys, though not rigorously proved, is in conformity with all the observations made during the course of this investigation.

The Effect of the Use of Machine Turnings in the Charge. Where machine shop turnings are available, this scrap is quite frequently used to make a portion of the charge. Figure 28 shows the average test-bar properties of all designs when poured from melts prepared in a high-frequency induction furnace with a silica lining, with and without 5 per cent machine turnings in the charge. The results shown are the averages of triplicate melts, each point representing 36 bars of six different designs. It will be noted that, in this instance, at least, the presence of the machine turnings in the charge had no appreciable effect upon the average test-bar results obtained.

Some previous, more limited data, not listed here, had indicated that machine turnings in the charge may produce slightly lower test-bar properties. Apparently, the important factor is the cleanliness of the machine turnings. If harmful impurities are permitted to contaminate them, adverse effects may be produced. If, however, contamination is not a factor, the machine turnings can be successfully remelted in the high-frequency induction furnace without damage to the melt quality produced.

The Effect of Melting Under a Glass Slag Cover. Six heats were prepared in the high-frequency induction furnace having a clay-graphite lining. Three of these heats were prepared by using the technique which gave the highest melt quality; namely, no cover was used, and the melt was skimmed three times during the heating-up period as described previously.

The other three heats differed in that a glass slag was charged into the furnace before the metal was melted, and this cover was maintained throughout the entire melting operation, and even in the pouring ladle. The C_1 charging method was identical for both series of heats and zinc and phosphorus additions were also identical, as described previously.

Figure 29 shows the average results obtained by the two different melting techniques. It will be noted that the glass slag cover has had a very adverse effect on the average test-bar properties obtained. The melt is obviously of high gas content, as shown by the very open, coppery fractures not illustrated here, and by the low test-bar properties obtained. Many melts were prepared using a glass slag cover before the three melts described here were prepared. The melt quality obtained was invariably low, and the results illustrated by Fig. 29 are typical.

Apparently, the removal of the glass slag before transferring the melt, as was done in many heats not described here, provides some opportunity for melt oxidation and partial gas removal. It is not known whether or not the adverse effects of the glass

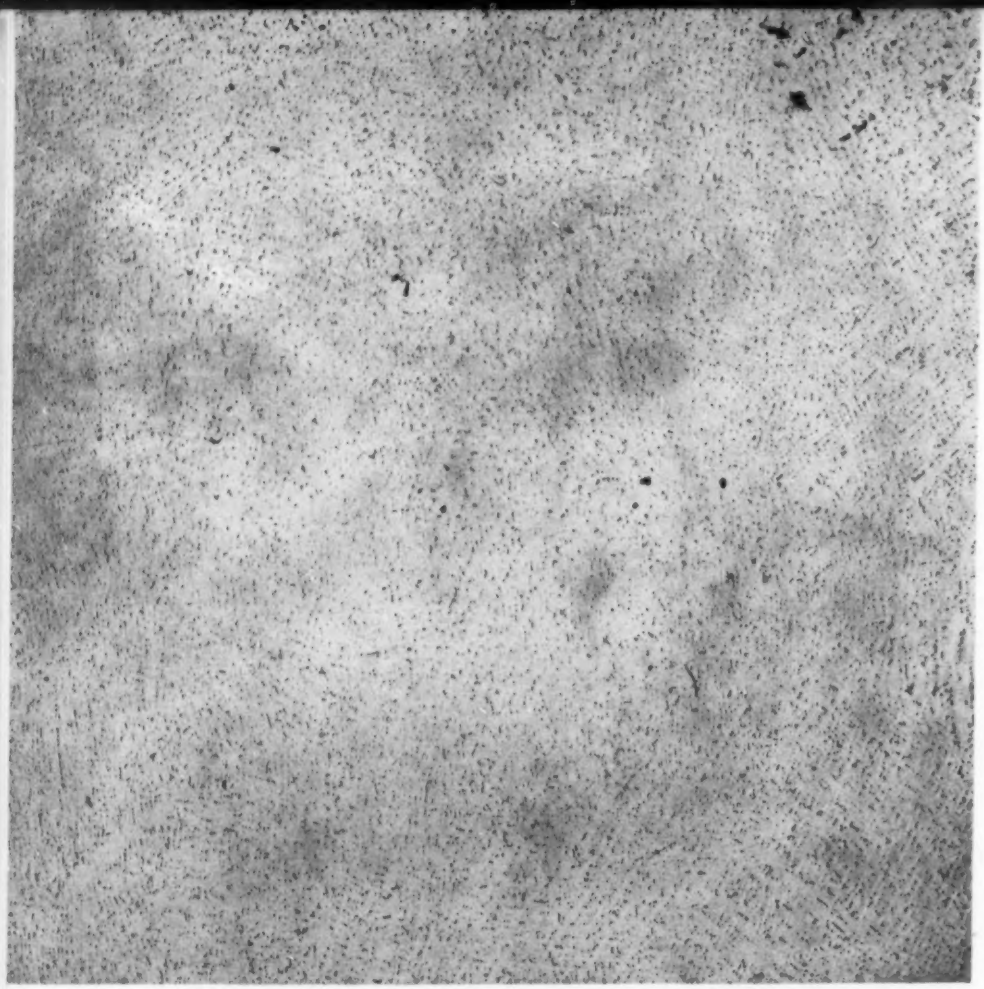


Fig. 33 (left)—Unetched. Longitudinal section at X13 of a Horizontal 5/8-in. web-Webbert bar poured at 2100° F. from a melt of high quality. Density in the shoulder is 8.85. This specimen was poured from the same melt as the one shown by Fig. 12, and is similar except that it was poured from a temperature of 2100° F. instead of 2240° F. The mechanical properties are 36,800 psi. tensile strength and 37.7 per cent elongation. The decreased pouring temperature has greatly decreased the amount of microporosity, but the lead particle size is much finer; the net effect of these two changes in structure, one favorable, the other unfavorable, is a slight reduction in tensile properties.

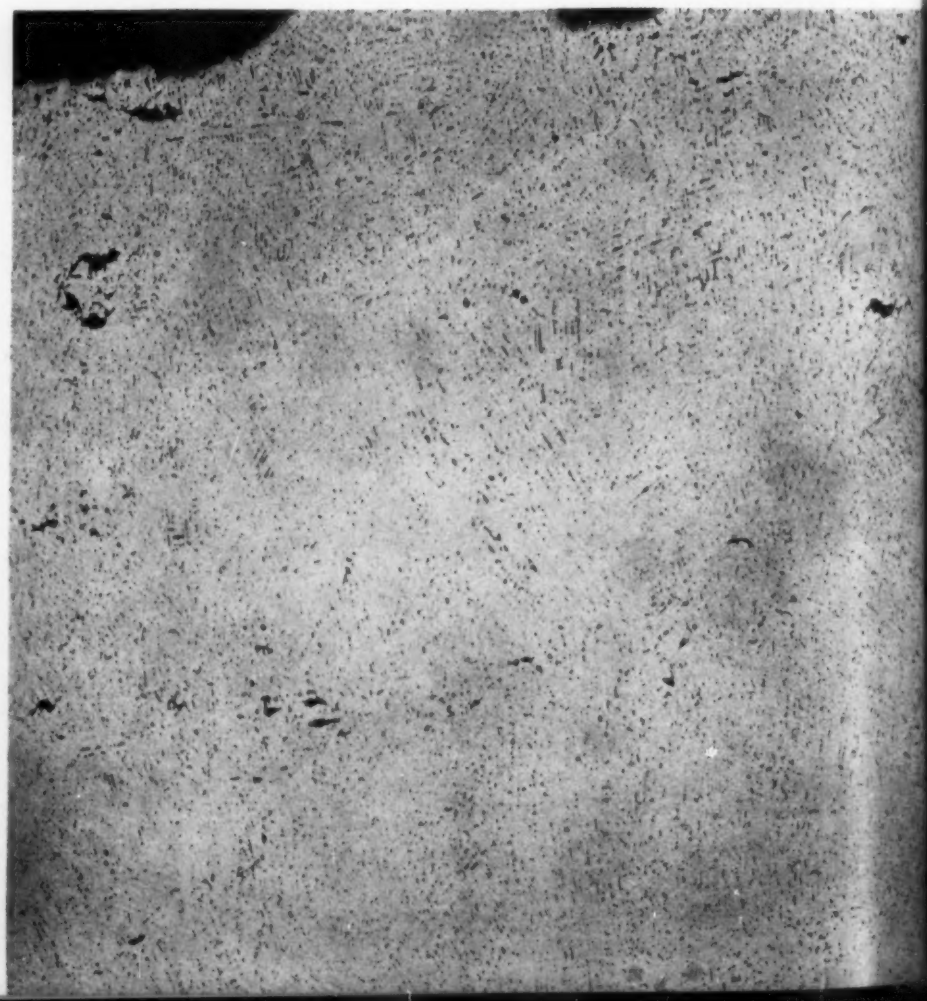


Fig. 34 (right)—Unetched. Longitudinal section at X13, adjacent to the fracture of a Horizontal 5/8-in. web-Webbert bar poured at 2000° F. from a melt of high quality. Density in the shoulder is 8.91. The lead particles are fine, but not as fine as those shown by Fig. 33. The low pouring temperature has produced a high degree of soundness. The voids shown as large dark areas are probably small ruptures produced when the bar was broken. Mechanical properties are 40,200 psi. tensile strength and 44.3 per cent elongation. As indicated by Figs. 30 and 31, this elongation value is higher and the lead particles coarser than average for this pouring temperature. This bar is similar to the one shown by Fig. 12 and 33, except that it was poured at a lower temperature.

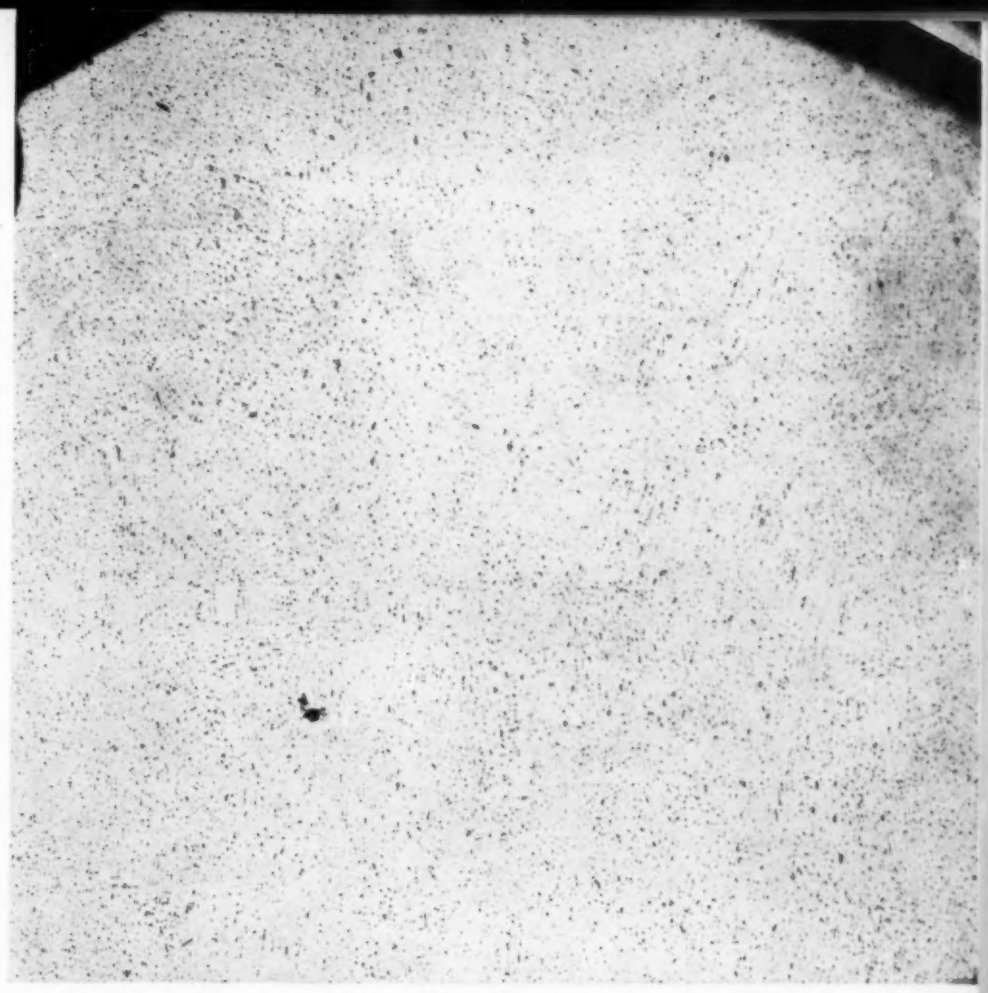


Fig. 36 (right)—Unetched. Longitudinal section at X13 near the fracture of an L.T.B. test bar at 2100° F. from a melt of high quality. Density is 8.90, and the mechanical properties are 38,600 psi. tensile strength with 29.1 per cent elongation. This bar is entirely similar to the one shown by Fig. 11, except the pouring temperature was 2100° F. instead of 2250° F. Main effect of decreasing the pouring temperature of this test-bar design is the occurrence of localized or centerline shrinkage at the low pouring temperatures. The local shrinkage, not shown, but apparent on the fracture, greatly reduces the tensile properties obtained. There may also be an adverse effect, not shown here, of low pouring temperature on the lead particle size. When the grains become equiaxed, the lead particles frequently become angular, resulting in low elongation values.

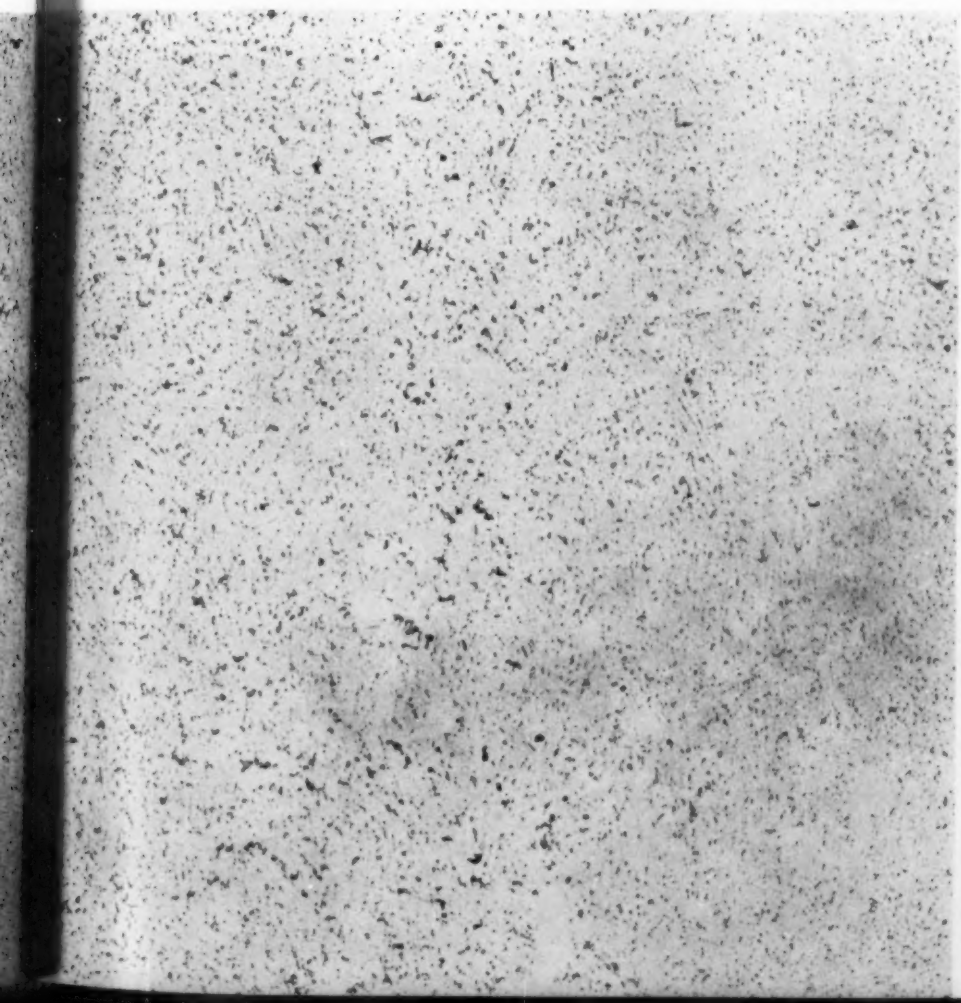


Fig. 35 (left)—Unetched. Longitudinal section at X13 near the fracture of a Horizontal 5/8-in. web-Webbert test bar poured at 2000° F. from a melt of poor quality. The specimen was poured from the same melt and is otherwise entirely similar to the one shown by Fig. 13, except the pouring temperature was 2000° F. instead of 2250° F. Greatly improved soundness produced by the lower pouring temperature is evident. However, the very angular lead particles have produced low elongation values despite the soundness. Compare with Fig. 34, which also shows a high degree of soundness, but this specimen was poured from a high-quality melt, and the lead particle shape is more favorable, i.e., less angular. Density is 8.80 and the mechanical properties are 38,900 psi. tensile strength with 33.9 per cent elongation.

slag are more pronounced when melting in a clay-graphite crucible than when melting in a silica-lined high-frequency induction furnace.

It is now generally accepted, as described previously, that copper-base alloys should be melted under slightly oxidizing conditions. All of the results obtained on this experimental program confirm this contention. It is obvious that a glass cover maintained over the melt at all times greatly reduces the contact with air and prevents suitable oxidation. The result is that the melts are gassy, the test bars obtained contain a very considerable amount of microporosity as evidenced by the coppery, open fracture, and their tensile properties are low.

VII.—Average Effect of Pouring Temperature on Tensile Properties of Various Test-Bar Designs

The average effect of pouring temperature on test-bar properties of any given test-bar design depends not only upon the design but also upon the melt quality employed. While the general influence of the melt quality upon the effects produced by variations in the pouring temperature is known, there are variations in the effects of pouring temperature that are at the present time inexplicable.

The general effects of pouring temperature upon the test-bar properties of five different designs poured from the first group of melts are shown by Fig. 30. The average melt quality from this group was good but slightly inferior to the average melt quality obtained by the third group of melts having the highest melt quality. The data on the third group of melts are represented by Fig. 31 for melts of high quality, and by Fig. 32 for melts of very low quality. Considerable study of these three figures is required.

In general, the tensile properties of all designs are increased with decreasing pouring temperature, but the rate of increase is very much greater with melts of very poor quality.

Several of the designs show a maximum elongation value at the intermediate pouring temperature or a gradual decrease in elongation with decreasing pouring temperature when they are cast from melts of good quality. When these same test-bar designs are poured from melts of low quality, the elongation values are very much lower, of course, but as the pouring temperature decreases, the elongation values increase very substantially, particularly as the pouring temperature is dropped from 2235 to 2135° F.

Effects of Pouring Temperature on the Microstructure and Soundness

Apparently, the microporosity resulting from melts of low quality, namely, high gas content, is very pronounced at the highest pouring temperature, especially in Webbert-type bars. This defect is less pronounced at the intermediate pouring temperature with the result that markedly better tensile properties are obtained. With a still further decrease in pouring temperature from 2135 to 2035° F., the microporosity defect decreases still further, but other adverse effects may partially counterbalance the improvements obtained in soundness.

A comparison of Figs. 12, 33, and 35, will show the effects of decreasing the pouring temperature on the structure, fracture, and density of Horizontal ⅝-in. web-Webbert bars poured from melts of good quality, while Figs. 13 and 35 show similar effects on bars poured from melts of poor quality. These two groups of photographs should be compared with Figs. 11 and 36, which show the effects of decreasing pouring temperature on the structure and soundness of the cast-to-shape L.T.B. bar poured from a high-quality melt. In contrast to the Horizontal ⅝-in. web-Webbert bars, the L.T.B. is quite sound even at the highest pouring temperature.

With melts of good quality, most of the designs show lower elongation at a pouring temperature of 2035° F. This apparently arises from one or more of several causes. The cast-to-shape bars, such as the Crown-type or the L.T.B. test bars, are frequently characterized by some localized shrinkage at the lower pouring temperatures, resulting in lowered properties. There is also a marked tendency for the grain size to decrease markedly at the lower pouring temperature, resulting in angular lead particles at the intersection of three or more fine equiaxed grains. These angular lead particles as illustrated by Figs. 16 and 17 have an adverse effect on the tensile properties obtained. There is also some evidence that, as the pouring temperature decreases, the lead particles become finer, which also has an adverse effect on the test-bar properties.

Again, Figs. 12, 33, and 34, Figs. 13 and 35, and Figs. 11 and 36, should be compared to observe the effects of decreasing pouring temperature on lead particle size and shape. Figure 35 is especially interesting because the combination of poor melt quality and low pouring temperature has produced sound bars having very angular lead particles and relatively low elongation values.

The lead particles are essentially discontinuities which have very little strength in themselves. As they become finer, they became more continuous and have a greater adverse effect on the elongation values. These same effects of discontinuities produced by temper carbon particles in malleable iron and by sulphide particles in some cast steels have also been noted[4]. These adverse effects of finer or more angular lead particles obtained with decreasing pouring temperature are partially or completely offset by the improved soundness unless localized shrinkage occurs, as it frequently does, in the cast-to-size bars such as the L.T.B.

These two opposite effects then combine to determine the average effects of pouring temperature on the test-bar properties of the various designs as illustrated by Figs. 30, 31, and 32. These general effects of pouring temperature on the soundness and lead particle size and distribution are illustrated by the figures referred to previously.

Test Bars Poured at High Temperature Reflect Melt Quality Most Reliably

In view of the above description and data on the effects of pouring temperature on the test-bar properties obtained, it is quite obvious that the melt quality, namely, the gas content of the melt, is most markedly reflected by the test-bar properties obtained at the highest pouring temperature. On the other hand, the test-bar properties obtained at the lowest pouring temperature depend upon many factors other than melt quality, and these factors are more difficult to control.

Some of them have been mentioned; as for example, as the pour-

ing temperature decreases, the size and shape of the lead particles become smaller or more angular, the fine equiaxed grains obtained at low pouring temperatures sometimes produce coarse, angular lead particles, or localized shrinkage occurs, especially in cast-to-shape bars, such as the L.T.B. Consequently, the melt quality rating of a single heat or of several melts made under the same conditions was obtained by using the weighted average elongation and tensile values.

By this procedure, the properties obtained at the highest pouring temperature near 2225° F. were given a weight of two; those obtained at the intermediate pouring temperature of approximately 2125° F. were given a weight of one; while those obtained at the lowest pouring temperature were not considered for the reasons described above. In effect, this average value is the tensile property which would be obtained at about 2200° F., i.e., ⅔ of the way between 2135° F. and 2235° F.

It may be worth while, at this point, to make some speculations regarding the characteristics which a test bar should have to reflect melt quality to a maximum extent. It was noted at the outset that the keel block was not suited to 85-5-5-5 alloy. Somewhat inadequate investigations indicated that, because of the heavy mass in this design, there was a tendency for slow and rather nondirectional solidification.

As a consequence, the microporosity and lead particles were coarse and angular, resulting in low properties even with melts of sufficiently high quality to produce excellent properties in other designs. This design is well fed, but this partial advantage is offset by the slow cooling which produces defects that overshadow the microporosity defects resulting from high gas contents in the melt.

The cast-to-shape Crown-type and the L.T.B. bars have strongly directional solidification in the test section which produces bars substantially free of microporosity, regardless of gas content, but centerline shrinkage may occur because of the poor feeding, especially at low pouring temperatures.

As a rule, these designs are erratic and not particularly reflective of melt quality because the defect normally accentuated by the gas in the melt, i.e., microporosity, does not occur. Instead, melts of high gas content produce fine, rounded pinholes which are entrapped in the test section in an erratic manner. These characteristics are generally unfavorable, but the L.T.B. apparently is slightly superior to the Crown bars, as will be described later.

Apparently the Webbert-type bars represent the best compromise between good feeding, nondirectional solidification, and too heavy sections on the one hand and rapid radial solidification with poor feeding on the other. Because microporosity may be reduced by either decreasing the gas content of the melt or by providing better feeding, the thickness of the web is quite critical. However, this phase of the problem has not been completely investigated.

VIII.—Correlation of the Test-Bar Properties of Individual Designs with Melt Quality Rating

As previously indicated, the primary purpose of test bars is that of a device for measuring melt quality. It is a characteristic of all castings that their mechanical properties will vary somewhat although they are poured from the same melt under identical conditions. There are, then, some uncontrolled factors which influence the mechanical properties of all castings including test bars. A perfect correlation, therefore, of the test-bar properties with melt quality is not to be expected, but a good test bar should reflect melt quality reliably.

The extent to which the various test-bar designs reflect melt quality has been graphically represented by Figs. 37 to 45, inclusive. This correlation has been carried out on all 12 melts of Group 1, all five melts of Group 2, and all 32 melts of Group 3 described previously. Although, as indicated earlier, not exactly the same series of test bars were poured from each group, the melt quality ratings have been obtained by using the weighted average tensile properties of all designs as described above.

As previously stated, the properties obtained at the highest pouring temperature were given a weight of two, those at the intermediate temperature a weight of one, while the values obtained at the lowest pouring temperature were not included in these calculations. Thus, the 49 melts prepared in Groups 1, 2, and 3, were rated in the decreasing order of melt quality. Since each test-bar design contained two test bars per mold, the average properties obtained on these two bars from a given pouring temperature are plotted against the melt quality rating.

Thus, Fig. 37 shows the relationship of the melt quality to the average test-bar properties of the two Horizontal ⅝-in. web-Webbert test bars from a single casting poured from the highest temperature. Figure 38 shows the similar relationship for the two bars poured from the intermediate temperature. Figures 39 and 40 show a similar pair of charts for the L.T.B. test bar.

Figures 41, 42, 43, and 44 show similar charts for the Navy 5/16-in. web-Webbert. The first two, Figs. 41 and 42, show the relationship of the average test-bar properties of two bars cast from the high and intermediate temperatures, respectively.

Figures 42 and 43, on the other hand, show the individual test-bar values. These two charts show a considerable scatter in the individual test-bar values, and it is quite evident that, on the basis of individual bar values, it would be entirely possible to reject a satisfactory or good heat and accept a rather poor heat. Because the individual values do show this scatter, all the other charts are based on the average of the two bars. This is also in conformity with the ASTM Specification B145-44T, which permits the average of two bars to be used.

Figure 45 shows the correlation of the average of two bars of the Navy 5/16-in. web-Webbert poured at the lowest pouring temperature. In general, it will be noted that the correlation becomes less perfect as the pouring temperature decreases. This is to be expected, and it is caused partly by the method of obtaining melt quality ratings and in part by the more erratic properties obtained at the lowest pouring temperature because of their dependence upon many factors in addition to melt quality.

In general, the Navy 5/16-in. web-Webbert produces the best correlation of the test-bar properties with melt quality. The Horizontal ⅝-in. web-Webbert and the L.T.B. test bars show a very definite trend, the test-bar properties decreasing with melt quality, but there is more scatter in the data. On the other hand,

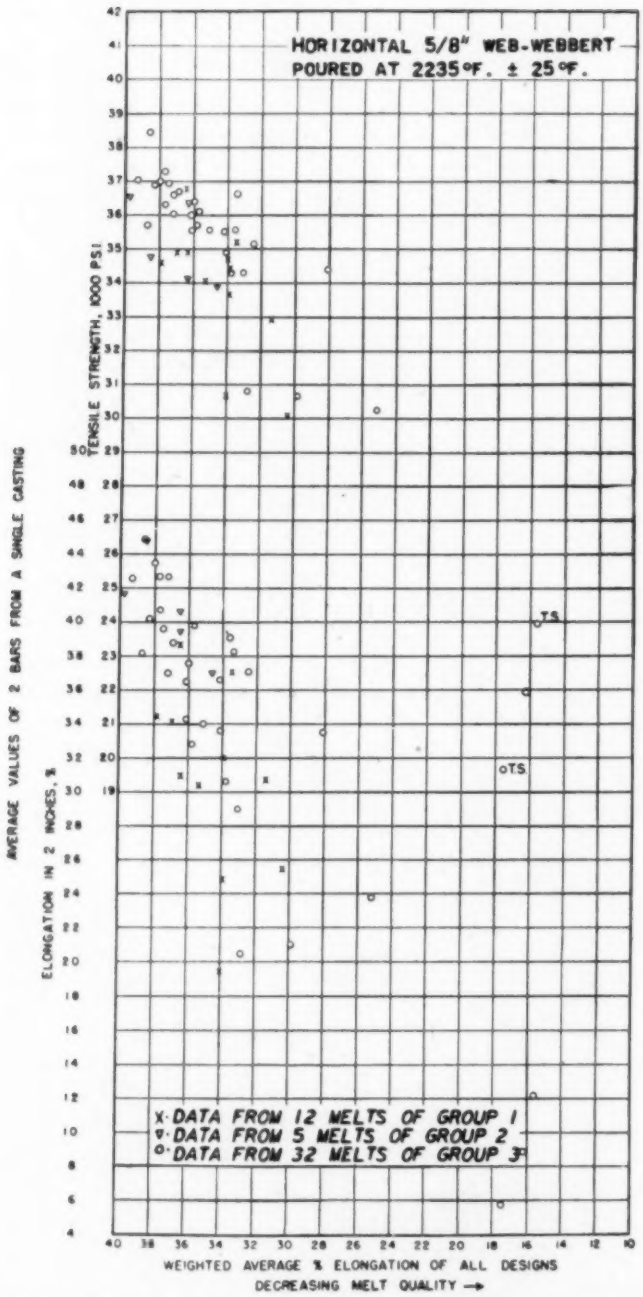


Fig. 37—Average tensile values of two Horizontal 5/8-in. web-Webbert bars from a single casting poured at 2235° F.

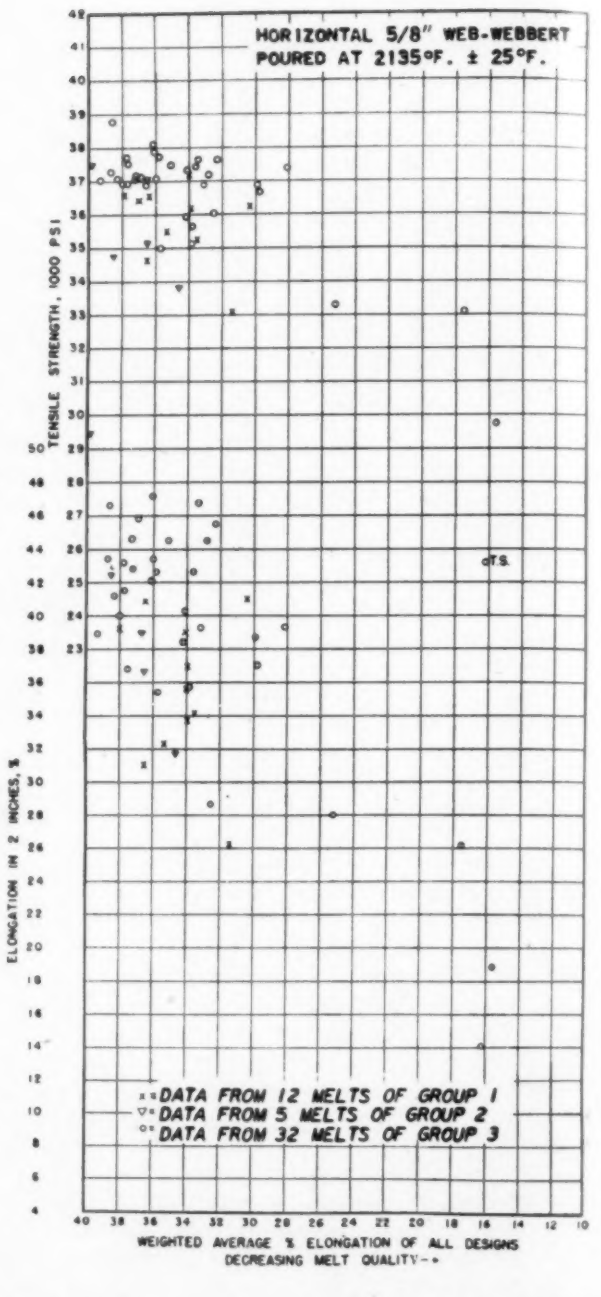


Fig. 38—Average tensile values of two Horizontal 5/8-in. web-Webbert bars from a single casting poured at 2135° F.

more data are available on these two designs and this may account in part for the greater scatter obtained.

Similar charts, not shown here, were prepared for the Modified 5/8-in. web-Webbert, Modified Crown, Universal, Navy 5/8-in. web-Crown bars. All of these showed rather poor correlation of test-bar properties with melt quality. In general, those data obtained with the Navy 5/8-in. web-Crown, Navy 5/8-in. web-Webbert, and Universal bars were low as well as erratic.

Those data obtained with the Modified 5/8-in. web-Webbert were very high, and while the lowest melt quality produced moderately low properties, the higher range in melt qualities produced bars having similar and very high tensile properties. Thus, all of these designs did not differentiate between melts differing considerably in melt quality, but did differentiate between melts of widely divergent melt qualities.

IX.—Summary and Conclusions

As employed in this paper, the term "melt quality" is synonymous with melt composition and refers specifically to the content of gas and gas formers in the melt in the pot. The various types and kinds of gas evolution during solidification have been outlined. In general, this gas evolution during solidification produces a defect known as microporosity.

If all other factors are maintained constant, this microporosity in the test bars increases with increasing gas content of the melt which in turn produces decreasing tensile properties. As a result, a reasonably fair

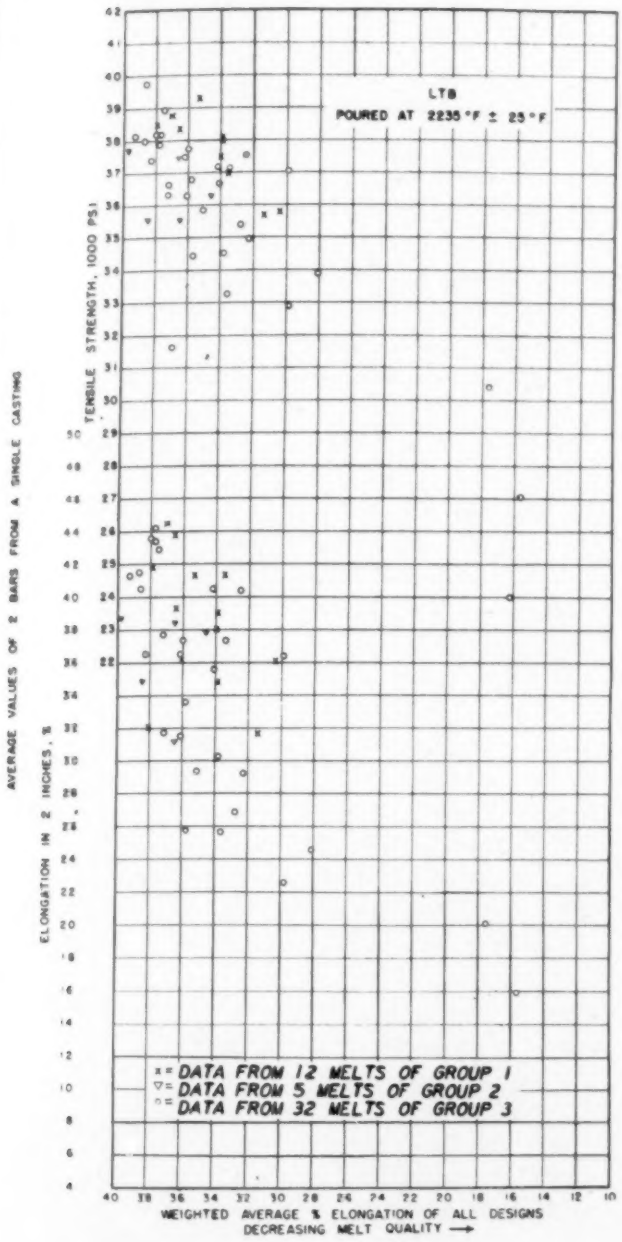


Fig. 39—Average tensile values of two L.T.B. bars from a single casting poured at 2235° F.

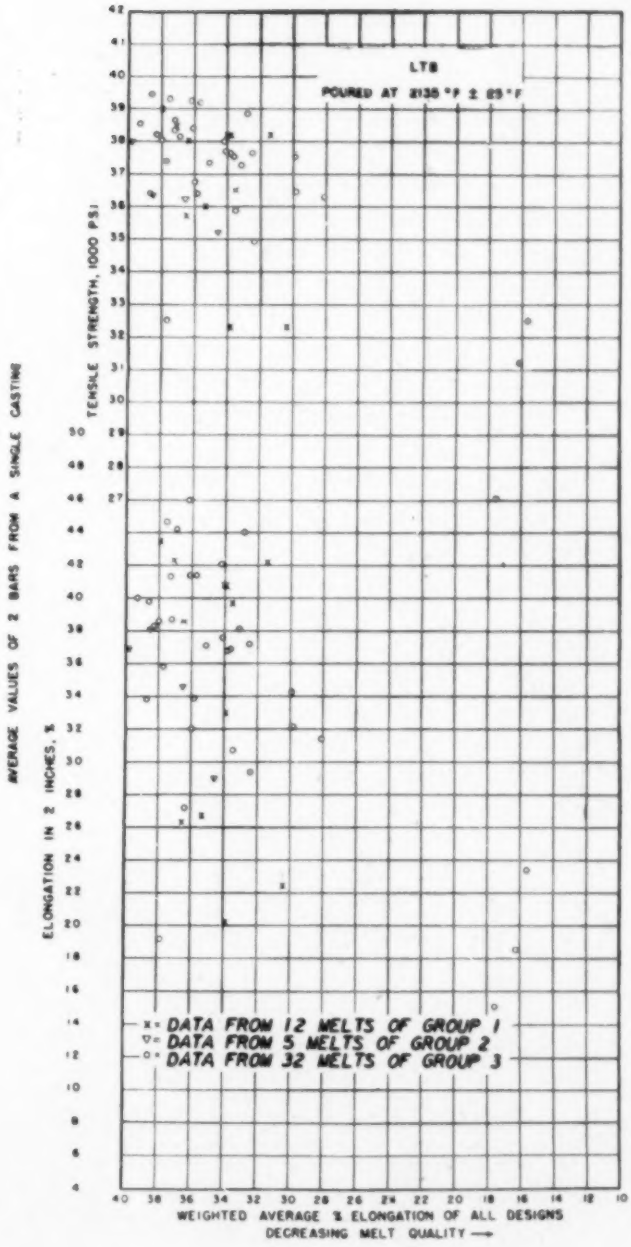


Fig. 40—Average tensile values of two L.T.B. bars from a single casting poured at 2135° F.

correlation of melt quality with the average tensile properties of the test bars is obtained, provided a standardized test-bar procedure is employed.

A study has been made of test bars, test-bar practice, test-bar design, and of the effects of various melting practices on the test-bar properties.

The results obtained may be summarized under six main headings as follows:

Effect of Test-Bar Design

1. Of the eleven test-bar designs investigated, the British D.T.D. bar, shown in Fig. 8, was discarded early in the investigation because the low properties obtained indicated that factors other than melt quality were predominant. The properties obtained, however, were equal to or superior to those reported by British investigators.

2. The keel block casting, shown in Fig. 4; the Universal, Fig. 1; the Navy 5/8-in. web-Webbert; and Navy Crown, Fig 7, also produced very low properties, and as a result, it was concluded that these test bars are not suited for 85-5-5-5 alloy.

3. Using the weighted average tensile properties of several designs of test bars as a measure of melt quality, the tensile properties obtained on the Universal, Fig. 1; the Modified Crown, Fig. 9; the Navy 5/8-in. web-Crown, Fig. 7; the Modified 5/8-in. web-Webbert, Fig. 9; and the Navy 5/8-in. web-Webbert, have produced a rather poor correlation with melt quality.

4. On the other hand, the Horizontal 5/8-in. web-Webbert, Fig. 3; the

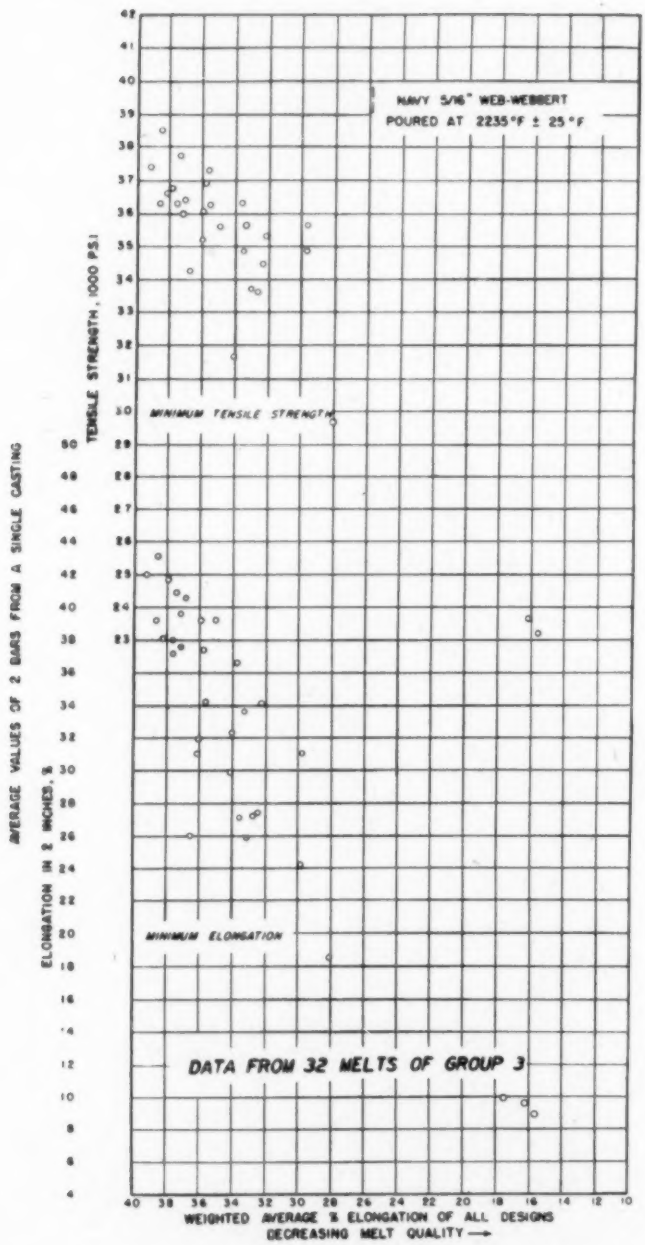


Fig. 41—Average tensile values of two Navy 5/16-in. web-Webbert bars from a single casting poured at 2235° F.

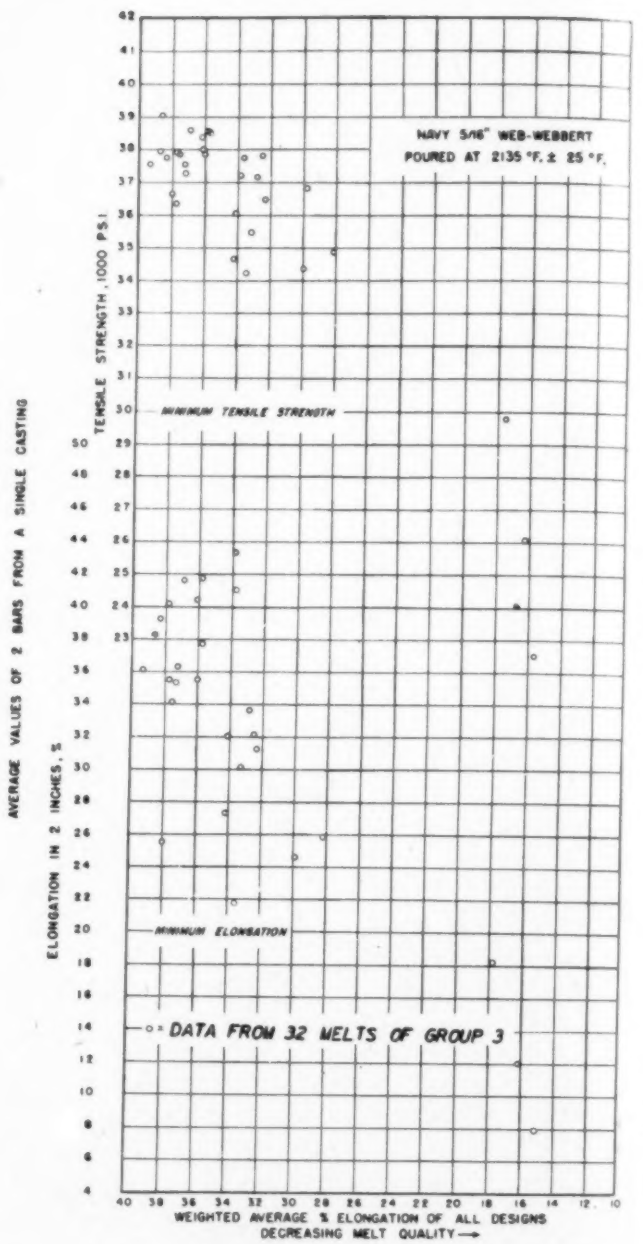


Fig. 42—Average tensile values of two Navy 5/16-in. web-Webbert bars from a single casting poured at 2135° F.

L.T.B., Fig. 2; and the Navy 5/16-in. web-Webbert, Fig. 6, all show a fair correlation of the average tensile properties of the two bars with melt quality. Of these three designs, the tensile properties of the Navy 5/16-in. web-Webbert appear to produce slightly the best correlation with melt quality.

5. The individual test-bar values, however, show a considerable scatter, and it would be possible to accept a melt of rather poor quality or reject one of the rather high quality on the basis of single-bar values of any of the test-bar designs investigated.

Relation of Structure, Fracture, and Test-Bar Properties

1. As the gas content of the melt in the pot increases, with all other factors the same, the quantity of microporosity in the Webbert-type test bars increases with a consequent lowering of tensile properties.

2. The presence of microporosity in the Webbert-type bars is evidenced by the occurrence of an open, coppery appearance of the fracture. The greater the amount of microporosity in the test bar, the greater the area and the coarser the coppery or open portion of the fracture, making it possible to estimate the amount of microporosity by the examination of the fracture appearance.

3. In cast-to-shape bars, such as the L.T.B., Fig. 2, increasing gas content in the melt produces an increasing amount of fine, rounded, gas porosity in the test bar which is not readily observed on the fracture, but which, in general, causes a reduction in tensile properties.

4. In addition to the amount of microporosity in Webbert bars or small rounded pinholes in cast-to-shape bars, other factors which determine the properties of the test

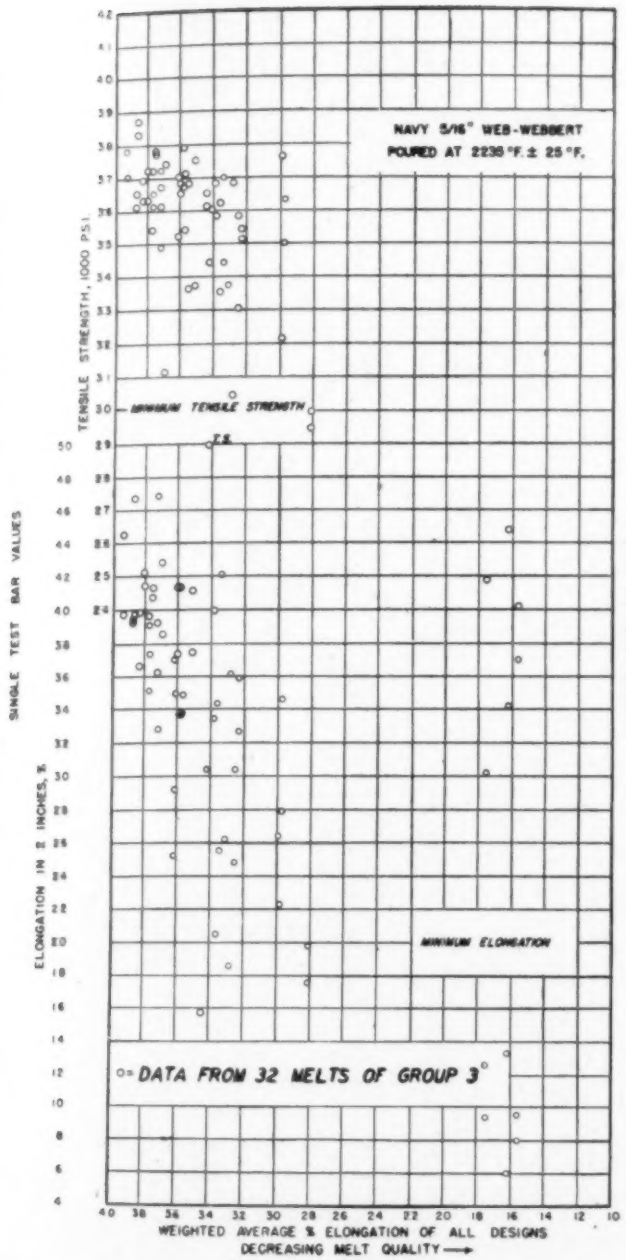


Fig. 43—Single Navy 5/16-in. web-Webbert tensile test bar values; bar poured at 2235° F.

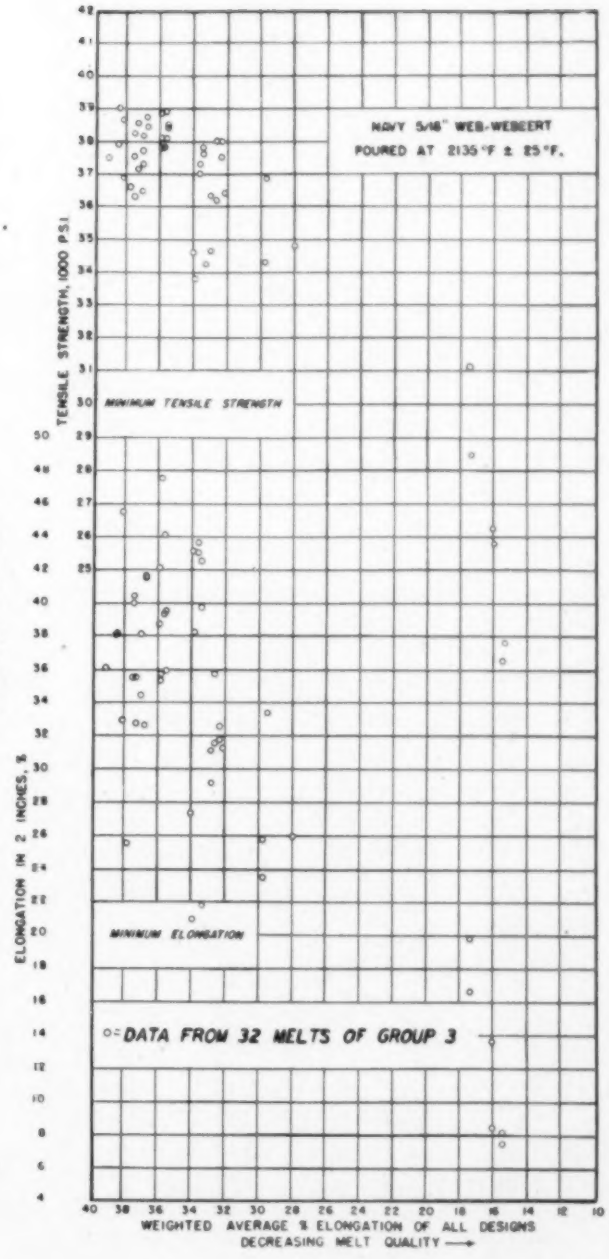


Fig. 44—Single Navy 5/16-in. web-Webbert tensile test bar values; bar poured at 2135° F.

bars apparently are size, shape, and distribution of the lead particles. In general, coarse, rounded lead particles are the most favorable, while very fine lead particles or coarse angular lead particles generally have an adverse effect on tensiles.

5. Localized shrinkage also causes low properties of certain cast-to-shape designs, such as the L.T.B., when poured at a low temperature.

6. In general, unfavorable lead particle size, shape, and distribution in all designs, and the occurrence of localized shrinkage in the cast-to-shape designs become more pronounced as the pouring temperature is lowered. On the other hand, the unsoundness in the form of microporosity in Webbert bars or rounded pinholes in cast-to-shape bars and which are caused by the gas content of the melt, becomes more pronounced as the pouring temperature is increased.

7. The interrelation of these two effects of pouring temperature, as outlined in Item 6, appears to determine the over-all effect of pouring temperature on the bar properties.

8. Because the microporosity caused by the gas content of the melt is most pronounced at higher pouring temperatures, melt quality is most accurately reflected by the test-bar properties when the test bars are poured at a temperature near 2200° F.

9. Because of the effect of pouring temperature noted in Item 6 in the foregoing, the over-all effect of pouring temperature on the tensile properties of test bars will depend upon the quality of the melt.

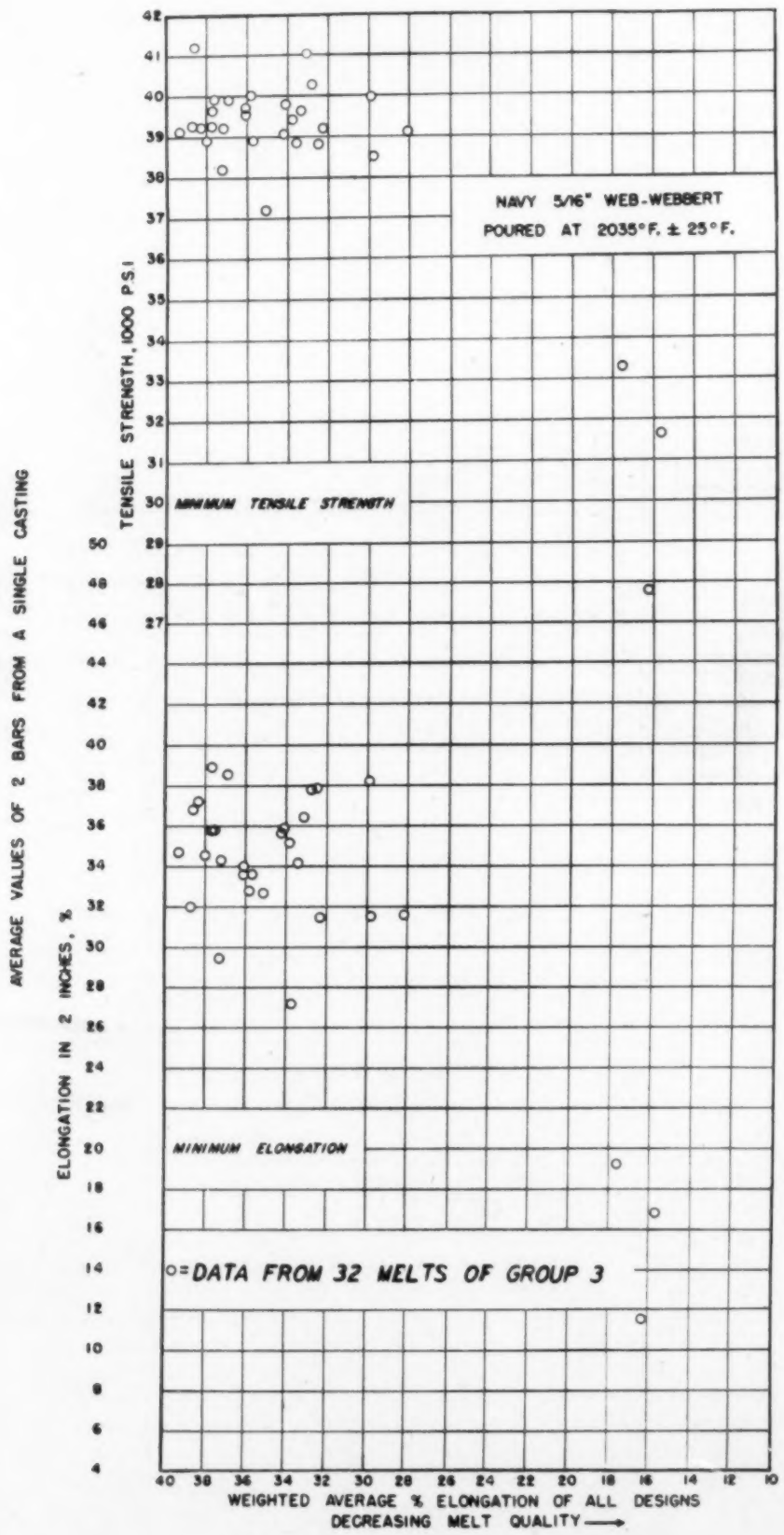


Fig. 45—Average tensile values of two Navy 5/16-in. web-Webbert bars from a single casting poured at 2035° F.

10. With a poor quality melt, the tensile properties of test bars of all designs will, in general, increase with decreasing pouring temperature. With melts of high quality, most designs produce maximum tensile properties at a relatively high pouring temperature near 2200° F. Intermediate melt qualities produce more complex relations between pouring temperature and the test-bar properties obtained.

Effect of Mold Material

1. Wide variations in moisture content or in A.F.A. permeability of natural or synthetic sands have only very slight effects on the test-bar properties obtained in several of the different designs of the test bars.

2. From Item 1, it was concluded that gas absorption in the mold was not an important factor determining the tensile properties of test bars of 85-5-5-5 alloy.

3. The tensile properties of the Horizontal 5/8-in. web-Webbert test bar were adversely affected by the use of a synthetic sand, whereas the keel block casting was unaffected by the use of this type of sand.

4. It should not be concluded from these observations that wide variations in permeability or moisture content of the green sand would be acceptable in commercial foundry operations.

Effect of Some Variations in Test-Bar Practice

1. At least during a period of high atmospheric humidity, poor test-bar properties were obtained from a high-quality melt by pouring it from a height of about 10 in. from the mold surface. Thus, a high-quality melt was converted to low quality by gas absorption during the pour.

2. In general, erratic properties were obtained in the Horizontal web-Webbert test bar by employing a slow pour or by pouring the metal from a height of about 10 in.

3. If the mold of the Horizontal 5/8-in. web-Webbert is tilted so that one test bar is lower than the other, the lowest tensile properties will be obtained in the lowest bar.

4. Open risers generally produce less localized shrinkage and higher tensile properties than blind risers in the L.T.B. test bar. Presumably this observation would also hold for the two-bar design.

Effect of Melting Practice on Melt Quality or on Test-Bar Properties

Two methods of charging were employed. The C_1 method consisted of preparing a molten heel and then melting down one ingot or piece of scrap at a time. The C_2 method of melting consisted of keeping the furnace full until the entire charge was melted. The gas-fired furnaces were operated to produce approximately 0.5 per cent oxygen in the furnace atmosphere. Using the weighted average tensile properties of several designs of test bars as a measure of melt quality, the following conclusions on the effect of melting practice on melt quality may be made:

1. A silica lining in the high-frequency induction furnace produced slightly higher melt quality than when a clay-graphite lining was employed.

2. A gas-fired, rotating barrel-type furnace lined with firebrick produced a melt quality nearly comparable to that obtained in the high-frequency induction furnace lined with silica.

3. A gas-fired, tilting-type pot furnace, having a clay-graphite crucible, produced melt qualities substantially inferior to those obtained in either the induction or in the barrel-type furnace.

4. The C_2 method of charging possibly may be very slightly superior to the C_1 method when melting in the silica-lined high-frequency induction furnace.

5. The C_1 charging method is substantially superior to the C_2 method when melting in a gas-fired clay-graphite crucible furnace or in a high-frequency induction furnace lined with clay graphite.

6. The C_1 method was not practical in the gas-fired, rotating barrel-type furnace.

7. The C_1 method of charging required a slightly longer time to melt the charge, and slightly higher metal losses were suffered.

8. The best melt qualities were obtained by exposing the melt to oxidation, and the lowest melt qualities were obtained by maintaining a glass cover on the melt prepared in a high-frequency induction furnace lined with clay graphite.

9. In general, all of these observations on the effects of melting methods on the melt quality confirm the theory of gas in copper-base alloys as outlined in the text.

General Conclusions on Test-Bar Practice

1. Test-bar properties are not indicative of the quality of the castings poured from the same melt, but may be indicative of the melt quality in the pot provided test-bar practices are standardized and controlled. Thus, test-bar properties are indicative of only one important factor determining casting quality.

2. The melt quality will be most accurately reflected by test-bar properties of 85-5-5-5 alloy if the following are adhered to:

A. A specific test-bar design is selected. On the basis of existing information, the Navy 5/16-in. web-Webbert or the Horizontal ⅝-in. web-Webbert appear the most suitable.

B. The gate dimensions, sprue size, and method of making the pattern is fixed to avoid variations in these factors.

C. The mold is horizontal to prevent variations in the flow of metal in the mold.

D. The melt is poured rapidly enough to maintain a full sprue at all times, and the pouring lip is held low. If necessary with a full pot, the mold should be set high enough enough to make this possible.

E. The same sand as used for the other castings is employed.

F. Good moisture control is employed, and the use of water and mold coatings is avoided.

G. The test bar is poured from various melts at the same temperature, preferably near 2200° F.

H. The melt quality is judged on the basis of the average tensile properties of at least two sound bars and not on individual bar values.

X.—Acknowledgments

Dr. Clamer desires to extend grateful acknowledgment for the assistance rendered him in preparation of this study to the following: members of the Metallurgists' Advisory Committee of the Non-Ferrous Ingot Metal Institute for general supervisory assistance on the project; George P. Halliwell, Chairman, H. L. Smith and J. F. Ednie, members of the Research Sub-Committee of the Metallurgists' Advisory Committee; Leighton M. Long, Leighton M. Long & Associates, Toledo, Ohio, for consulting service on foundry problems; C. W. Barnhart, who conducted the metallographic work; J. R. Doig, who performed the microradiograph investigations; and to J. G. Kura, Research Engineer; L. W. Eastwood, Assistant Supervisor, and C. H. Lorig, Supervisor of the Process Metallurgy Division, Battelle Memorial Institute, Columbus, Ohio, under whose direction the project has been carried on.

XI.—Appendix

The invitation to present the 1946 American Foundrymen's Association Foundation Lecture before the Association's Golden Anniversary Convention was accepted with the understanding that Dr. Clamer would act as spokesman for submission of a report on a research project conducted at Battelle Memorial Institute, Columbus, Ohio, under the sponsorship of the Non-Ferrous Ingot Metal Institute. At the time this invitation was accepted, it was the intent that the research project would cover the study of gases and oxides in copper-base alloys for sand castings. General supervision of the project was delegated by the Institute to its Metallurgists' Advisory Committee, of which Dr. Clamer is Chairman.

It was realized at the start that it was essential, before this project could be intelligently undertaken, that a standard test-bar design and standardized practice for making it would have to be developed and adopted. Many sporadic studies had previously been made and much had been published on the subject, but there was no record that the problem had previously been attacked in a comprehensive, scientific manner.

This is clearly set forth in the paper by A. J. Smith, appearing in the 1944 Proceedings of the American Society for Testing Materials. Dr. Smith, in his paper, has traced the historical records leading to the adoption of test bars now in use. He summarizes as follows: "The bars have been brought to fruition as a hodge-podge of ideas which can be defended on neither historical nor theoretical grounds."

The Metallurgists' Advisory Committee of the Non-Ferrous Ingot

Metal Institute thereupon decided to make such an investigation the first step of the Battelle project. This first step, as it developed, proved to be of greater magnitude than originally contemplated. Two years have been devoted to this initial phase of the project, notwithstanding the fact that it was narrowed down to the study of test-bar design and test-bar practice for a single alloy; namely, 85 copper, 5 tin, 5 lead, 5 zinc. It is, therefore, only on this preliminary stage of the project that Dr. Clamer was able to report.

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FLUIDITY TESTING of FOUNDRY ALLOYS

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Fluidity testing of foundry alloys is a subject of importance to the foundry industry because it is a method which can be used to evalute the casting characteristics of alloys and which may be employed for quality control in the shop. Many investigations on fluidity have been made but few investigators have employed the same test piece. An objective of the Committee on Fluidity Testing of the American Foundrymen's Association is to determine if a single, standard fluidity test which is reasonably suitable for all foundry alloys can be adopted by the industry so that data from many sources can be compared directly. Accordingly, published work on fluidity has been appraised by the author and certain recommendations are offered.

Two REASONS for fluidity testing are: First, fluidity of new or unfamiliar alloys may be determined so that the person responsible for casting operations can choose the one with the best fluidity among several alloys which otherwise may be equally suitable for the specific application.

Some evidence exists[105] which indicates that many alloys have approximately equal fluidity when they are heated an equal degree above their respective liquidus temperatures. However, this point has not been proved conclusively and more data are necessary for the information of foundrymen. There is a definite need for fluidity testing in the laboratory to aid the development of better foundry practice.

Second, fluidity testing on the shop floor is a means of quality control to which some foundries have not given careful consideration. Some foundrymen believe that fluidity testing is strictly a laboratory expedient and that it has no place in a production shop. Certainly, in foundries where no castings having thin sections or intricate shapes are poured, or where misrun castings are never encountered, it would be a waste of time to train personnel for reliable fluidity testing.

Likewise, in certain non-ferrous foundries where temperature measurement is not difficult, fluidity testing assumes minor importance. However, in shops where metal is used at temperatures above 2200° F. (1205° C.) for the pouring of castings which are susceptible to misruns, it is believed that routine fluidity testing can be employed with profit. In support of this statement, it may be said that a fluidity test has been used with marked success to prevent the occurrence of misrun steel castings in the foundry of the U. S. Naval Research Laboratory.

After a period of observation, it was learned that if the length of the fluidity spiral poured with metal taken from the direct-arc furnace is less than 20 in., intricate castings are apt to be misrun; that if the length is more than 26 in., misruns do not occur, and that if the length is 20 to 26 in., misruns occur occasionally. These observations apply to normal operation in the pouring of miscellaneous shop castings at the U. S. Naval Research Laboratory, and the limiting values would be different in another foundry where ladle preheating, ladle capacity, pouring time, and other conditions would be different.

Fluidity Defined

Fluidity is the term used in this paper to express the ability of metal to flow. It is the resultant effect of several physical conditions which govern the solidification of molten metal as it flows in a mold. Fluidity is not a true physical property of the metal and is difficult to define exactly.

A foundryman can express it most easily by means of a comparison. For example, metal is said to have adequate fluidity for a specific casting application if it flows freely through the gates and fills the mold completely so that a casting is formed without misruns or cold shuts.

Fluidity Evaluated. Many troublesome variables would have to be considered if an attempt were made to evaluate fluidity on a truly physical basis, as heat transfer, fluid flow,

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viscosity, mechanics of solidification, film formation, surface tension, gas evolution, and suspended non-metallic materials are operative in determining this complex quantity.

Acordingly, the only convenient method of evaluating fluidity is an empirical one. That is, carefully standardized test conditions must be established and rigorously reproduced from test to test. Even when this is done, the values obtained are only relative and can be used only by observed correlation with casting practice for control purposes or for comparison of one metal with another on an arbitrary basis.

Often Investigated

Many investigators have studied fluidity and have reported their findings in the literature of foundry technology. Figures 1 through 5 illustrate some of the test pieces which have been used. A comprehensive bibliography[1-60] on fluidity of metals covering the period from 1902 to 1935 was prepared by the British Iron and Steel Institute.

The preponderance of attention in recent years has been given to ferrous alloys, and the work of British investigators has been particularly noteworthy. These reports[66, 67, 75, 76, 77, 78] and other valuable contributions on fluidity [61-65, 68-74, 79-105] since 1935 are also listed in the bibliography at the end of this paper.

Thus, most of the factors concerning the condition of the metal and their effects upon fluidity have been considered at one time or another. The unfortunate feature of all this work is that many of the investigators have adopted totally new test pieces or have so altered previously used test pieces that it is not possible to compare data.

Progress usually comes from changes and alterations and these changes have served a useful purpose. However, it appears desirable to agree upon a standard test piece and testing procedure so that foundrymen can speak a common language of fluidity and can compare notes on the fluidity of their alloys. In this manner, a rational correlation between measured fluidity and casting practice can be made in a relatively short period. Adoption of such a standard throughout the foundry industry is the objective of the Fluidity Committee of the American Foundrymen's Association.

It is not meant to imply that one test piece and one testing procedure are necessarily best for all foundry alloys, but it is believed that some concessions, made in the interest of standardization, would be beneficial to the interests of all. Therefore, this paper is written to present a survey and critical appraisal of published information concerning the variables of fluidity testing.

Aspects Considered

In the discussion of these variables which follows, consideration is given to sensitivity and reproducibility for laboratory tests and to simplicity for shop practice. Emphasis is placed on the requirements of fluidity tests for high-melting-point alloys, but special requirements for certain of the lower-melting-point alloys are not overlooked.

Ideal Conditions for Fluidity Testing. In practically all of the tests which have been used to evaluate fluidity of molten metals, the same basic method has been followed. Metal is caused to flow into a chan-

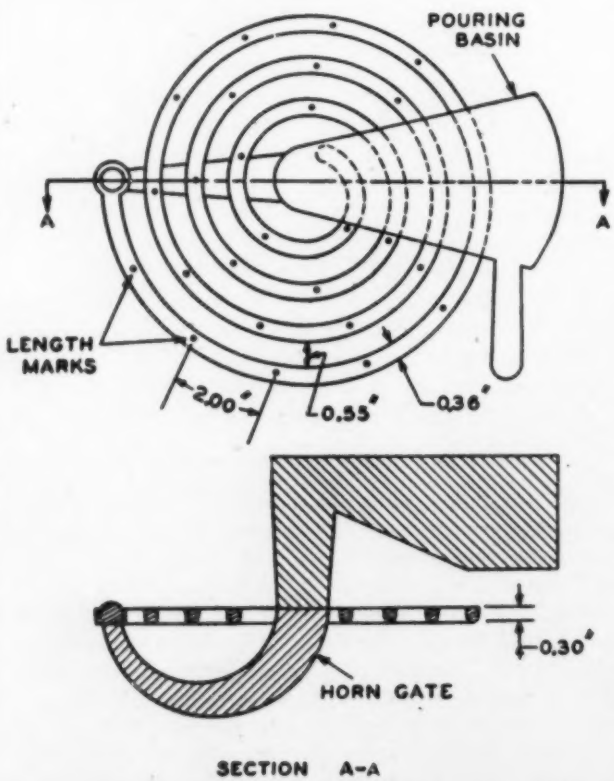


Fig. 1—Fluidity pattern assembly (Saeger, Krynitsky).

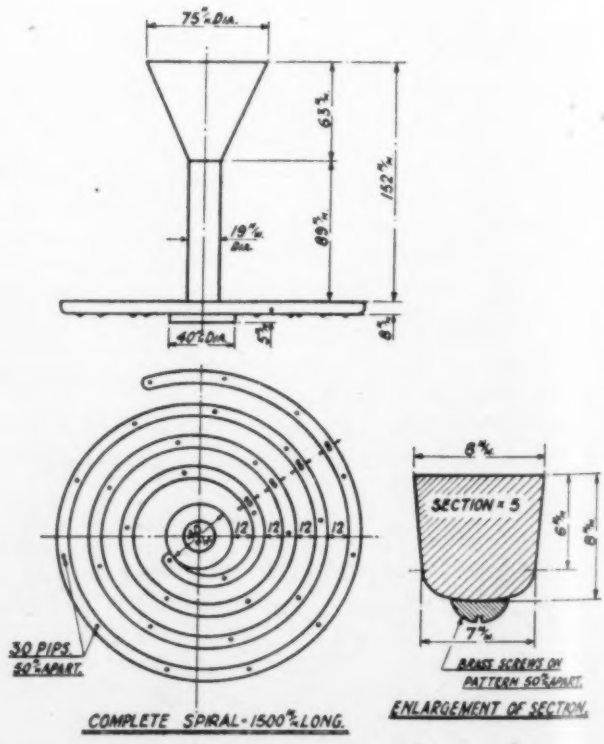


Fig. 2—Spiral for fluidity test (Gréaves).

nel of relatively small cross-sectional areas for a distance determined by the rate at which the mold extracts sufficient heat to cause solidification of the metal and to stop movement of the stream. The length of flow is taken as the measure of fluidity. As with any test, sensitivity and reproducibility are necessary if the test is to be useful.

For a fair degree of sensitivity, the test piece must be designed with a definite relationship among the factors of force acting to drive the metal through the flow channel, the dimensions of the flow channel, and the thermal characteristics of the mold material. Regardless of the fluidity of the metal, if the proper relationship is not obtained, the metal in one instance may be forced through the full length of the channel or, in another, may be frozen without having flowed an appreciable distance. In either extreme case the test is meaningless.

If the fluidity test piece is designed for adequate sensitivity, duplication of test conditions within close limits from test to test is still a major problem. Ideally, the conditions for optimum reproducibility are as follows:

1. The force exerted upon the metal at the point of entry to the flow channel should be constant.

2. No other variable force such as that of gravity, due to variations in pouring rate or metal head, or of expanding gases should be imposed upon the metal during its period of flow.

3. The conditions of heat transfer must remain constant from test to test.

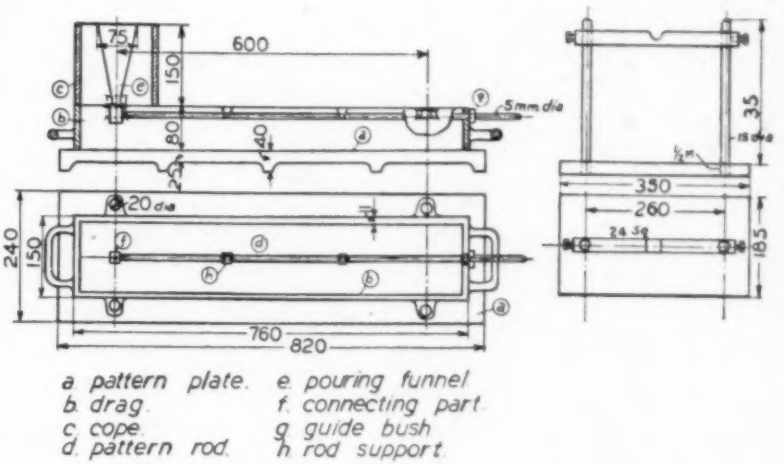


Fig. 4—Spoon test for determination of the running capacity of liquid metal (Ruff).

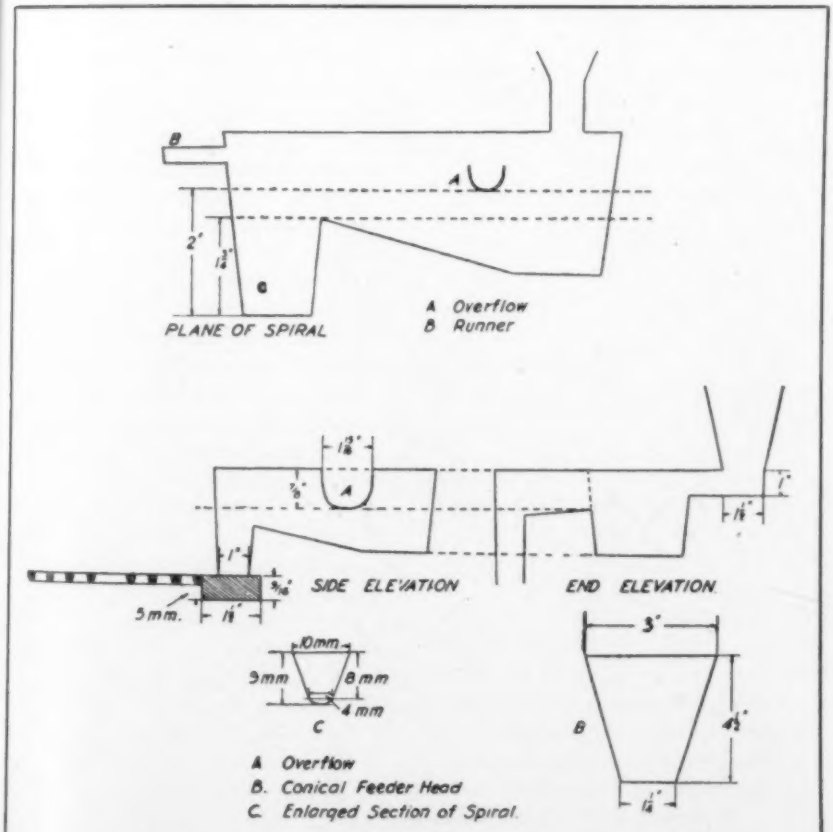


Fig. 3—Fluidity test piece (Andrew, Percival, and Bottomley).

In practice these ideals are difficult to attain and some compromise must be made. These compromises are discussed in the paragraphs which follow.

Shape and Dimensions of the Flow Channel. In one of the earliest investigations on fluidity, West[1] used a casting in the form of an elongated wedge. This type of test casting is still used in some foundries. However, most of the reported data were obtained by the employment of flow channels with uniform shapes and dimensions throughout their lengths. Square, rectangular, circular, half-circular and trapezoidal sections have been used.

General Indications

Since the practical value of a fluidity test is mainly relative, it appears that the cross-sectional shape makes little difference except for the ease with which the pattern is molded. From this standpoint, a circular section is not satisfactory. Unless a straight rod is used, a circular section must be molded with a parting line and the danger of mismatches is always present.

Because the complete impression of the flow channel can be made in one-half of the mold, and because

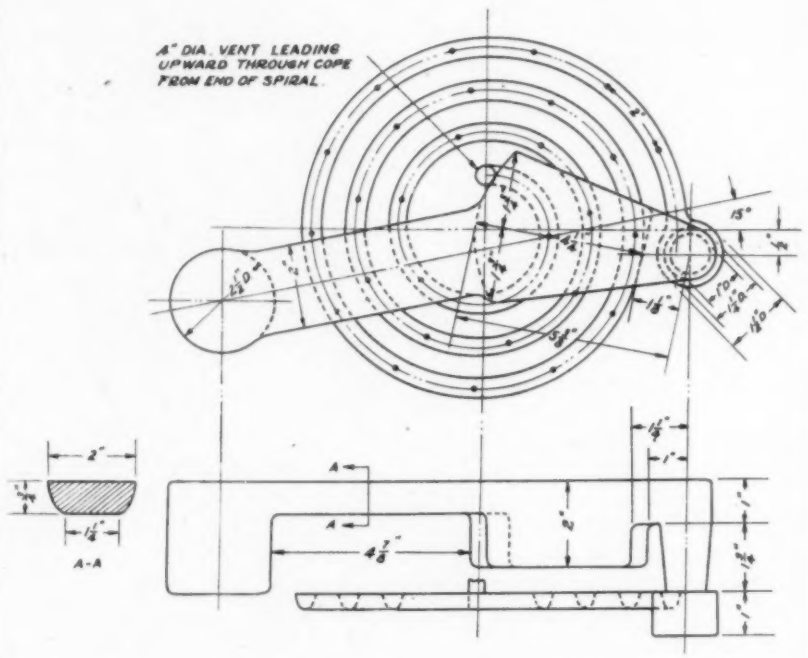


Fig. 5—Fluidity spiral test piece (Taylor, Rominski, and Briggs).

of the ease with which the pattern may be drawn from the sand, a half-round or trapezoidal section is believed to be most practical for the flow channel.

Dimensional Variations

Cross-sectional dimensions of the flow channel can be varied within certain limits in accordance with the height of the down gate and the thermal characteristics of the molding material. Saeger and Krynitsky[28] used a rounded, trapezoidal flow channel with an area of approximately 1/12 sq. in. The approximate cross-sectional area of the flow channel used by Andrew, Percival and Bottomley[67] is 1/11 sq. in. Ruff's[65] channel is circular and has a cross-sectional area of approximately 1/30 sq. in.

Taylor, Rominski and Briggs[89] made their tests with a flow channel having a rounded, trapezoidal shape and an area of approximately 1/7 sq. in. Greaves[66] also used a rounded, trapezoidal flow channel with a cross-sectional area of approximately 1/9 sq. in. Although the foregoing investigators made their studies upon ferrous alloys in sand molds, it has been reported to the writer that satisfactory fluidity tests have also been made upon aluminum and copper-base alloys with the test piece of Taylor, Rominski and Briggs.

Ruff[65] did not use the fluidity test proposed by Saeger and Krynitsky[28] because he believed that the flow channel was too large to avoid turbulent flow of the metal. Whether or not turbulent flow is deleterious to the test is debatable. However, in view of the apparently satisfactory results of several investigators who used the larger flow channels, it is believed desirable to establish the cross-sectional area of a standard fluidity test channel between 1/12 and 1/7 sq. in.

Length of the flow channel must be such that solidification of the most fluid metal to be tested occurs before the stream reaches the full length of the cavity. This, in turn, is determined by the cross-sectional area of the flow channel and by other factors which have been stated previously, namely, the thermal nature of the mold material and the pressure head exerted upon the metal.

Of the test pieces mentioned in connection with the discussion of the cross-sectional area of flow channel, those used by Saeger and Krynitsky[28], Greaves[66], and Taylor, Rominski and Briggs[89] are all approximately 60 in. long. That used by Ruff[65] is approximately 24 in. long. If the recommendations regarding cross-sectional area are to be followed, it would appear necessary to establish a length of flow channel of approximately 60 in.

Spiral Flow Channels vs. Straight Flow Channels. In most of the test pieces designed for measurement of fluidity a spiral configuration of the flow channel has been adopted for reasons of simplicity in molding. Ruff[65] used a straight flow channel but the length is relatively short, 24 in. as compared with a length of approximately 60 in. for the other test pieces which have been mentioned.

Molding Errors

If the test piece is to be molded in a two-part flask, it is almost essential that the pattern be mounted on a board. Obviously, if the board is more than 60 in. long, as is required when the previously recommended flow channels are made straight, much more opportunity exists for distortion during handling and storage of the pattern or during ramming of the sand than if the flow channel is formed as a spiral.

Even though the cross-sectional dimensions are assumed to be as large as ½x½ in., a distortion of the board of as little as 0.025 in. causes a variation of 5 per cent in the cross-sectional area. Surely, if only one condition of the test is subject to errors of this order, close control of other unavoidable errors would not be sufficient to make the test reproducible within reasonable limits.

Another point in favor of the spiral flow channel is the matter of levelling of the mold. In any laboratory determination where the highest precision is desirable, test molds should be levelled carefully, regardless of the design of the mold. However, for quality control in the shop, it appears desirable to make the test as insensitive to levelling as possible. In a case where one or more turns of the spiral are filled with molten steel, variations in pressure head caused by a slight inclination of the mold may be very nearly self-compensated.

For control purposes at the U. S. Naval Research Laboratory, the spiral-type fluidity mold is levelled only by sight. It is placed upon a thin layer of sand in the bottom of a wheelbarrow. When it is desirable to make a fluidity test, the mold is

wheeled to a convenient position and be poured without further preparation. This freedom would not be possible if a test piece with a straight flow channel were employed.

Other advantages of the spiral flow channel are that standard flask equipment may be used and the mold is small enough to be handled with ease.

Material of the Mold for Fluidity Testing. Several of the arguments set forth in the foregoing paragraphs have been made on the assumption that green sand is the most desirable molding medium. This point deserves some discussion.

If a choice is to be made between a permanent mold of cast iron or steel and a sand mold for fluidity testing, there appears to be little room for argument. Permanent molds have the one advantage of offering a better opportunity for constancy of dimensions of the flow channel. However, if scaling and pitting from repeated use are to be avoided when high-melting-point alloys are tested in the mold, a mold dressing must be applied.

Effect of Dressings

Since heat is carried away rapidly by a metallic mold, variation in the thermal characteristics of the dressing causes a profound effect upon the time necessary to extract enough heat from the flowing metal to cause it to solidify, and therefore upon the indicated fluidity. Moreover, the presence of any gas-forming material in a mold dressing is a probable cause of erroneous test results since a permanent mold has no permeability and entrapped gas causes unpredictable retardation or acceleration of the metal flow.

The choice between green and dry sand molds or baked core-sand assemblies is more difficult to make. An obvious difficulty with core assemblies is warpage of one-half of the mold in relation to the other half which occurs during baking. Unless the cores are baked as assembled in supporting frames which are retained in position until the test mold has been used, this warpage is apt to cause "run-outs" and unreliable test results.

Such difficulties are not experienced in the use of straight flow channels if corrosion-resistant straight rods are baked in the sand and are removed later, as was done by Kron and Lorig[79] who used a modification of Ruff's method for their fluidity studies.

Green sand molds are more economical than baked sand molds, and several investigators have expressed the view that they are entirely satisfactory if certain precautions are observed. First, the bond strength of the sand must be adequate so that sand is not eroded by the flowing metal. Second, the sand must be sufficiently permeable so that the steam formed by the rapid heating of the water in the molding sand can escape easily to the atmosphere and does not cause pressure to be exerted upon the metal.

Sand Moisture Content

In this connection, it is deemed advisable to provide an ample vent at the end of the flow channel as a factor of safety to prevent the development of gas pressure ahead of the flowing metal. Whether or not normal variations in the moisture content of a sand have appreciable effects upon the chilling action of the mold has been considered by several investigators. Curry[22] reported that variations of from 3 to 7 per cent appeared to make little difference in the measured fluidity.

Some differences were ascribed to variations of the moisture content of the molding sand by Saeger and Krynitsky[28].

Greaves[66] stated that moisture content of the sand seems to have little effect upon chilling of the metal unless it has flowed over a considerable area of the mold cavity. Taylor, Rominski and Briggs[89] found no significant differences in fluidity as the moisture content of the sand was changed.

If one is inclined to prefer a baked sand mold on the basis of freedom from moisture, the possibility of water absorption during storage should not be overlooked; Berger[51] summarized the situation in this regard by saying that it is easier to keep the moisture of green sand constant than to maintain dryness of cores and dried sand molds.

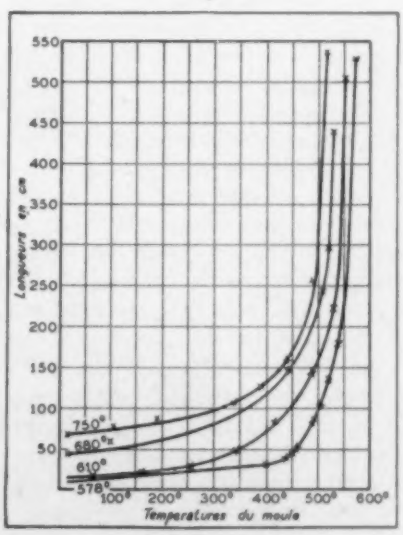
Ruff[65] conducted a study of metal flow in narrow channels to determine the effect of surface conditions of the mold cavity. For purposes of standardization of the fluidity test, a detailed treatment of these factors does not seem necessary. However, it appears desirable, for the sake of uniformity, to specify the grain size of the sand within narrow limits.

Silica sand with an A.F.A. grain size number of 80 ± 5 should be suitable if it does not contain abnormal quantities of silt and so bonded that the permeability is not less than 75 by the A.F.A. permeability test.

Temperature of Mold. The rate of heat flow from the metal varies with the initial temperature of the fluidity mold and it is to be expected that fluidity measurements would also vary with the temperature of the mold. However, Courty[29] has shown that fluidity values of aluminum-base alloys determined in cast iron molds are not increased appreciably even though the molds are heated considerably above that temperature which can be expected from climatic variations (Fig. 6). Therefore, it is believed necessary to specify only that the standard fluidity test mold be at room temperature at the time of test.

Gating Methods. Ideally, the system of gating should be so designed that metal is available at the entry of the flow channel under a constant pressure head and without appreciable inertia. Such an ideal could be achieved only by having a reservoir of liquid immediately adjacent to the flow channel with a quick-opening valve at the point of its entry, and by having provision for maintenance of a constant pressure head on the reservoir. Successful operation of a quick-opening valve in molten metal at this position in

Fig. 6—Castability of an aluminum alloy vs. mold temperature (Courty).



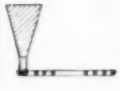
Method of Gating	A		B		C		D		E	
										
Material	Steel	Malleable Cast Iron	Steel	Malleable Cast Iron	Steel	Malleable Cast Iron	Steel	Malleable Cast Iron	Steel	Malleable Cast Iron
Number of experiments	10	6	8	6	8	6	37	13	5	6
Deviation of values, %	25	5	6	3	7	7	3	2	3	6
Maximum spiral length, Mm.	275	1091	780	1435	615	1480	940	flowed to end	745	flowed to end
Running quality ratio of steel to malleable cast iron (max.)	1 : 4·0		1 : 1·8		1 : 2·4		1 : 1·6		1 : 2·1	
Mean spiral length, Mm.	147	933	647	1261	548	1230	753	1365	616	1314
Running quality ratio of steel to malleable cast iron (mean)	1 : 6·3		1 : 1·9		1 : 2·2		1 : 1·8		1 : 2·1	

Table 1—Spiral tests of steel and malleable cast iron for different gates (Ruff).

the mold is a practical impossibility. Therefore, ideal gating conditions must be approached as nearly as possible by other means.

Ruff[65] conducted some experiments in which several details of gating were varied simultaneously. Table 1, taken from his paper, is reproduced here. His test results indicate that a horn gate is not suitable for introduction of steel into the flow channel.

Carapella and Shaw[104] were also concerned about the influence of a horn gate upon fluidity tests conducted with magnesium alloys. They have reported that better sensitivity was obtained by replacing the horn gate with a vertical down gate centered over a small basin from which entry to the flow channel was made.

Metal Flow

The cross-sectional area of the sprue must be considerably larger than that of the flow channel. In this way, the velocity of the metal is low before it enters the channel and turbulence is avoided to some extent. It has been shown by Courty[29] that less than one second is required for a aluminum-silicon alloy to run 50 cm. in a chill mold containing a spiral flow channel with a cross-sectional area of approximately 1/12 sq. in.

Although the duration of flow is probably somewhat longer in a sand mold, it is still short enough to require that the sprue be filled as rapidly as possible after pouring is started. In the absence of a valve mechanism, the pressure head causing flow varies during sprue filling.

Therefore, if the sprue is comparatively small so that it is filled quickly, inconsistencies in test results from this cause are held to a minimum. It is recommended that the ratio of cross-sectional areas of down gate to flow channel be approximately six to one, although it is recognized that the ratio can be changed somewhat from this value without affecting the test perceptibly.

From Ruff's[65] work and that of other investigators, it is apparent that a large basin at the bottom of the down gate has no useful purpose in fluidity testing. It seems desirable only to have an enlargement of the sprue as a part of the channel pattern to prevent a mismatch when the sprue is molded.

Pressure Head. If the variations of the pressure head which occur during filling of the sprue be neglected and if it be considered that the sprue is kept exactly full during the period of test, the force acting to cause flow in the channel is determined by the height of the sprue and by the density of the metal. After the dimensions of the flow channel and the material of the mold have been selected, the height of the sprue must be fixed within certain limits so that the test is reasonably sensitive and even the most fluid metals which may be encountered will not flow completely through the channel.

For example, from Ruff's[65] preliminary experiments with Saeger and Krynitsky's[28] test piece (Table 1), it is apparent that the length of spiral was not sufficient with methods of gating *D* and *E*. Even with methods *B* and *C*, the malleable iron flowed very nearly to the end of the 60-in. (1525-mm.) spiral. Under these conditions, the test is not satisfactory.

Height of Sprue

The sprue should be short for convenience of molding and for prevention of sand erosion and turbulence of the metal. It is recommended that the height of the effective head be established within the range of 1 to 2½ in. if sand molds are used with flow channels having dimensions within the limits previously recommended.

All of the investigators who used flow channels of that approximate size also used pressure heads within these limits except Greaves[66]. He employed a sprue approximately 6 in. high, but did not make any provision for the maintenance of a relatively constant pressure head.

From a comparison of the densities of aluminum and magnesium with those of iron-base and copper-base alloys the question of the necessity for changing the height of the sprue is raised if one standard test piece is to be applicable to all alloys Density is less effective in changing the conditions of test than might be expected.

Carapella and Shaw[104] used a test piece for measuring the fluidity of magnesium alloys which is similar to that developed by Saeger and Krynitsky[28] except that they substituted a simple sprue for the horn gate. Detailed dimensions are not given in their paper, but the effective height of the sprue is apparently unaltered. However, their plotted data indicate that the test is both sensitive and reproducible. Further, as has been stated, it is known by the writer that the test piece designed by Taylor, Rominski and Briggs[89] has been used satisfactorily for the measurement of fluidity of aluminum alloys.

Pouring Basins and Overflow. It is recommended that a pouring basin be provided in the standard fluidity test mold for two reasons. First, the kinetic energy of the metal being poured is largely dissipated in the pouring basin before it enters the

sprue. Second, a pouring basin is an excellent location for temperature measurement of the metal in laboratory tests. The second reason will be discussed in a subsequent section of this paper on measurement of the temperature of the metal.

Energy Dissipation

A few investigators of fluidity have taken elaborate precautions to dissipate the kinetic energy of the metal being poured before it enters the sprue. A clever device which was developed for this purpose is that used by Courty[29] who measured the fluidity of aluminum alloys in the test mold shown in Fig. 7. The sprue is stoppered by means of two fusible lead plugs. The time of melting of the plugs is regulated by the thickness.

Since lead has a much higher density than aluminum, and since the metals have practically no solubility for each other, the lead collects on the bottom of the comparatively large basin below the sprue and does not interfere with the test. This arrangement is well designed for the intended use. However, it could not be considered for a standard fluidity test because it does not appear practical for alloys with high melting points and is too complex for use as a quality-control test in the shop.

Some investigators of fluidity have poured metal directly into the sprue of the test mold. This practice does not seem to be desirable because the kinetic energy of the metal being poured is transmitted directly to that entering the flow channel. Moreover, it is apt to cause excessive and uncontrolled turbulence. If the test were to be reasonably reproducible under these conditions, metal would have to be poured from exactly the same height and at exactly the same rate, each time.

These variables are difficult to control and it is advisable to follow the methods employed by a number of the more recent investigators, namely, to design the test mold with a pouring basin which discharges into the sprue at a fixed height above the plane of the flow channel and which overflows into another basin at a small, fixed height above the plane of entry to the sprue.

Thus, if a high rate of pouring is employed so that the sprue fills quickly and remains filled, the pressure head is kept constant within the limits set by the difference between the two discharge levels. The shortest interval between discharge levels in any of the work reported is ¼ in., which is the distance employed by Andrew, Percival and Bottomley[67], and Taylor, Rominski and Briggs[89]. The latter test mold has been used widely enough to indicate that this dimension is ample although it obviously could not be much less than ¼ in.

If the amount of metal to be used for a fluidity test is to be kept reasonably small, the pouring basin can not be large. However, from the standpoint of dissipation of kinetic energy of the metal being poured, it is desirable to employ a fairly large basin. Moreover, it is essential that the pouring basin be larger than a certain minimum size so that temperature measurements can be taken therein. Size of the pouring basin is discussed further in the section on temperature measurement.

Skimmers. It is well known that the presence of slag and dross has a pronounced effect upon the apparent fluidity of a metal. The effect is more noticeable with some nonferrous alloys than with iron-base alloys. Accordingly, few of the investigators who have worked with irons and steels have given attention to skimming devices.

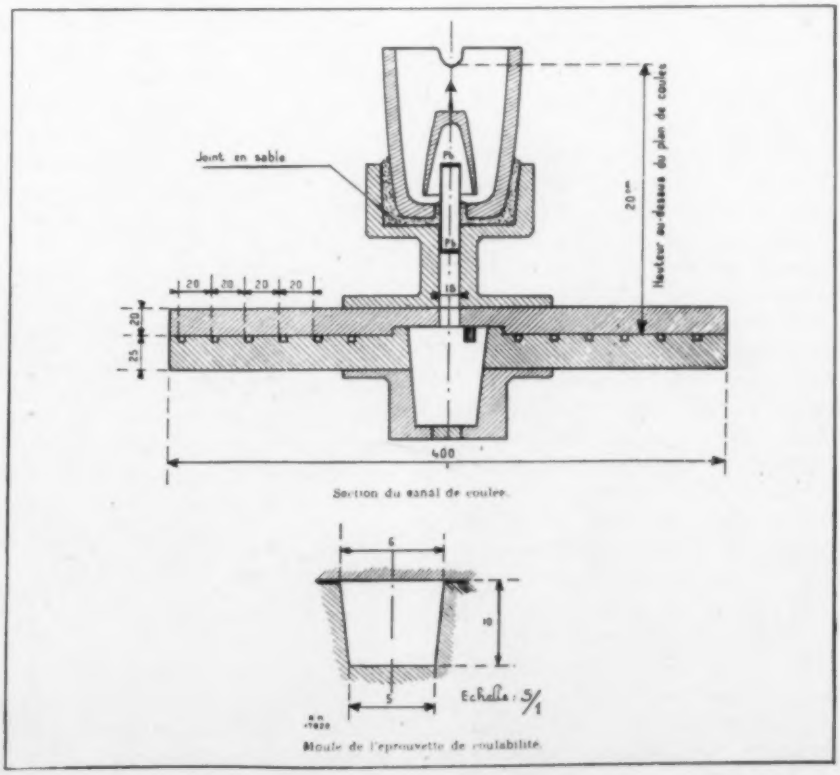
However, if the intention of specifying a standard fluidity test for all of the commonly cast metals is to be realized, it appears essential that some means be provided in the gating system to prevent floating, nonmetallic material from entering the flow channel.

Strainer Cores

A strainer core can be inserted in the sprue in the manner employed by Curry[22] and others. However, unless the openings through the core have a total cross-sectional area which is sufficient to allow rapid filling of the sprue, it is believed that some of the reproducibility of the test is lost. Since the core must have some strength and, therefore, be relatively large in comparison with the total area of the openings, the cross-sectional area of the sprue must be considerably larger above the core than below it.

Therefore, the sprue fills to a constant level less rapidly with a strainer core than without it, even though no constriction is caused by it. Moreover, the core acts to chill the metal somewhat, and it is desirable to keep

Fig. 7—Fluidity mold incorporating fusible stoppers in sprue (Courty).



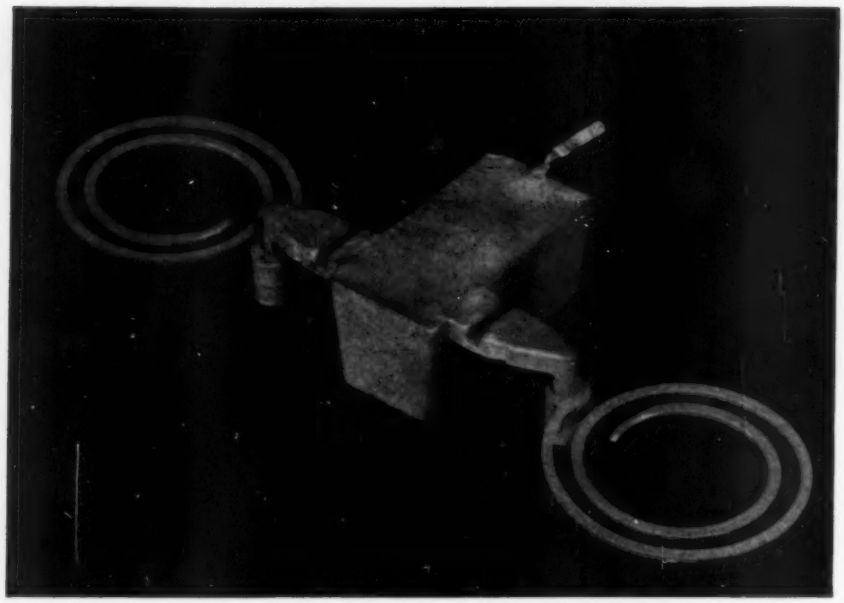


Fig. 8—Cast pair of spirals (Mott, Schaefer, and Cook).

the chilling action to a minimum.

Since it has been recommended that a pouring basin be made in the mold, a modification of the skimming device used by Mott, Schaefer and Cook[105] is believed to be satisfactory for skimming dross and slag from the metal. A photograph of a fluidity casting, taken from their report, is reproduced in Fig. 8. The suggested modification is that the section between the skimmer and the sprue be reduced to allow it to fill more completely before metal enters the sprue.

Temperature Measurement and Transfer of Metal to Test Mold. For fluidity testing in the laboratory, accurate measurement of the true temperature of the metal is an important part of the procedure because temperature has such a profound effect upon fluidity irrespective of other conditions which may alter it. Some of the inconsistencies in reported data probably are due to errors of temperature measurement. Even if the temperature of the metal is measured accurately in the furnace or in the ladle before the sample is poured into the mold, the metal can be chilled considerably as it is transferred.

For example, if metal is poured directly from a crucible or lip-pour ladle, the pouring lip may be at any temperature within a wide range depending upon a number of conditions; if the metal is dipped with a sampling spoon, the chilling effect of the spoon is dependent upon its mass, the degree of preheat, and the time required for the transfer. Thus, for laboratory tests, it is desirable to measure temperature of the metal in the fluidity mold itself.

Temperature Measurement Technique

Mott, Schaefer and Cook[105] have used a method which appears to be reliable. Their technique is to place the hot junction of a platinum, platinum-rhodium thermocouple in the pouring basin of the mold, Figs. 8 and 9, so that the temperature of the metal can be taken just before it enters the sprue. For this technique to be successful it is required that the couple respond rapidly and the indicator or recorder be equally responsive.

These investigators used a spot-welded couple with the junction no larger than the elements. This couple was pulled through a single-bore, fused-silica sheath with thin walls so that the thermal lag was not more than 4 or 5 sec. A rapidly responding indicator was used with the couple. It is necessary to have metal flowing into the pouring basin for a period which is long enough for the couple to respond to the true temperature of the metal before it enters the sprue.

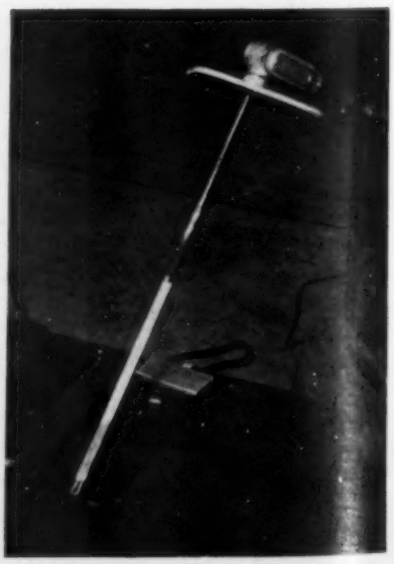
This condition can be controlled by establishing suitable dimensions for the pouring basin and by regulating the rate of pouring with a pouring cup. Mott, Schaefer and Cook used a pouring basin which is roughly cubical in shape with the length of each edge being approximately 4½ in.

Temperature measurements are not required for quality-control tests of fluidity in the shop. It should be emphasized again that useful information can be gained from this type of fluidity testing only by correlation of test results with observations of fluid performance of the metal on the casting floor for each shop.

Therefore, the important thing to be observed in shop practice is that the metal is transferred to the test mold in exactly the same manner for each test. For example, the procedure at the U. S. Naval Research Laboratory for making fluidity tests of steel in the arc-furnace for control purposes is as follows:

1. The bath is rabbled just before a sample is taken.
2. A 10-lb. sampling spoon, initially dry and at room temperature, is slagged for a period of 10 sec. as the melter counts, "one-thousand and one, one-thousand and two etc."
3. A spoonful of metal is dipped from a fixed position with respect to the electrodes and is withdrawn quickly, but not hurriedly, from the furnace.
4. The metal is deoxidized with a strip of aluminum unless the final

Fig. 9—Mold and thermocouple assembly prepared for pouring (Mott, Schaefer, and Cook).



additions have been made to the furnace prior to time of the test.

5. The spoon is inverted just above the pouring cup which is centered over the pouring basin. The average deviation from the standard time-cycle for steps 3, 4, and 5 is approximately 3 sec.

Sampling from Bottom-Pour Ladles. When it is desirable to make fluidity tests from a bottom-pouring ladle, some provision must be made to deal with the extraordinary force of the metal as it comes from the nozzle.

Investigations

Kron and Lorig[79] developed a test mold with which they were able to pour fluidity tests from a bottom-pouring ladle containing 27 tons of steel (Fig. 10). The flow channel of this mold is a modification of that developed by Ruff. Only a few tests were made with the mold and the data are insufficient to support a recommendation for its adoption.

Taylor, Rominski and Briggs altered the pouring basin of their laboratory test mold and placed a baffle core between the pouring basin and the sprue so that samples of steel could be taken from a bottom-pouring ladle. Other details are the same as those of the test piece for laboratory work or tests on the melting floor where a sampling spoon is used to transfer the metal to the mold. A photograph of the pattern for this test mold is shown in Fig. 11. Satisfactory tests were obtained with the large test piece and it is believed to be generally acceptable for use with a bottom-pouring ladle.

Shop Tests vs. Laboratory Tests. The idea has been emphasized that one well designed test piece should be suitable for both shop tests and laboratory tests of fluidity. The flow channel, sprue, and pressure head of molten steel should be identical in all cases. However, for the pouring basin, two or possibly three designs are necessary because of the differences in requirements of pouring methods.

For laboratory tests, the pouring basin must be sufficiently large so that accurate temperature measurements can be made of the metal as it enters the mold. The pouring basin of a mold which is to receive metal from a bottom-pouring ladle should also be large. By contrast, a smaller pouring basin is required for shop tests if metal is to be removed from a furnace with a sampling spoon.

It is not suggested that shop tests can be made with the same control and care with which a laboratory test is conducted. However, it is believed that the results of fluidity testing obtained from molds having the same pressure head of molten metal, flow channel and sprue, although with pouring basins sufficiently different in design to satisfy the requirements stated, would be suffiicently comparable so that operating and laboratory personnel could exchange information and improve practice for the benefit of the industry.

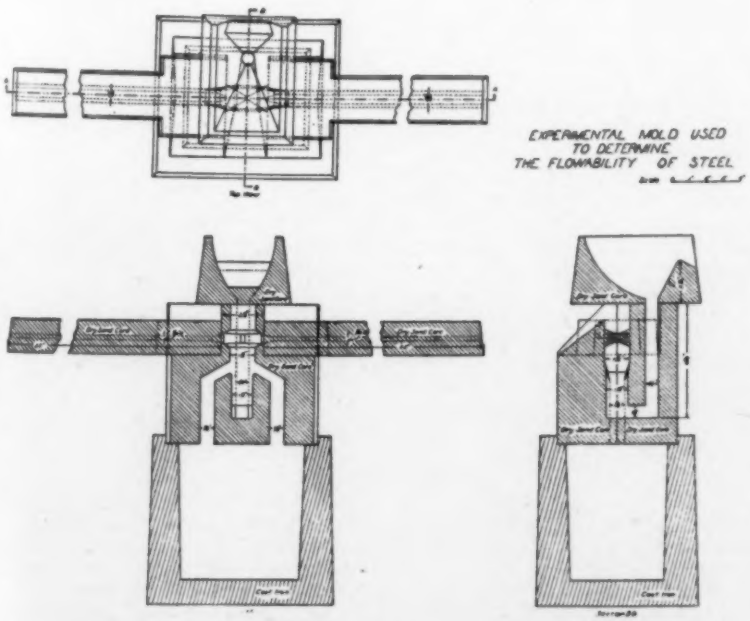


Fig. 10—Detail drawing of experimental mold used to determine the flowability of steel in foundry trials (Kron and Lorig).

Summary of Recommendations

A review of available information indicates that a standard fluidity test piece should have design details and provisions for operation similar to the following:

Flow Channel.

Half-circular or trapezoidal cross-section.

Cross-sectional area should be established at some value within the range of 1/12 to 1/7 sq. in.

Length of approximately 60 in.

Spiral configuration with a vent at the end.

Mold Material.

Green sand without a mold wash.

A.F.A. grain size of 80 ± 5.

Minimum A.F.A. permeability of 75.

Gating Arrangement.

Vertical sprue with a cross-sectional area approximately six times that of the flow channel.

Receiving well at the base of the sprue only slightly larger in diameter than the sprue.

Pressure Head.

Exact distance established within the range of 1 to 2 in. from plane of discharge from pouring basin to sprue.

Variation regulated to ¼ in. by discharge levels from pouring basin.

Pouring Basin.

At least two designs are necessary.

One should be sufficiently large to dissipate kinetic energy of metal being taken from a bottom-pouring ladle and to allow response of pyrometric equipment for temperature measurement in laboratory tests.

Other basin should be small enough so that the sample can be taken with a spoon.

Provided with an overflow to maintain a constant pressure head within the limits mentioned.

Skimmers.

Baffle core placed between pour-

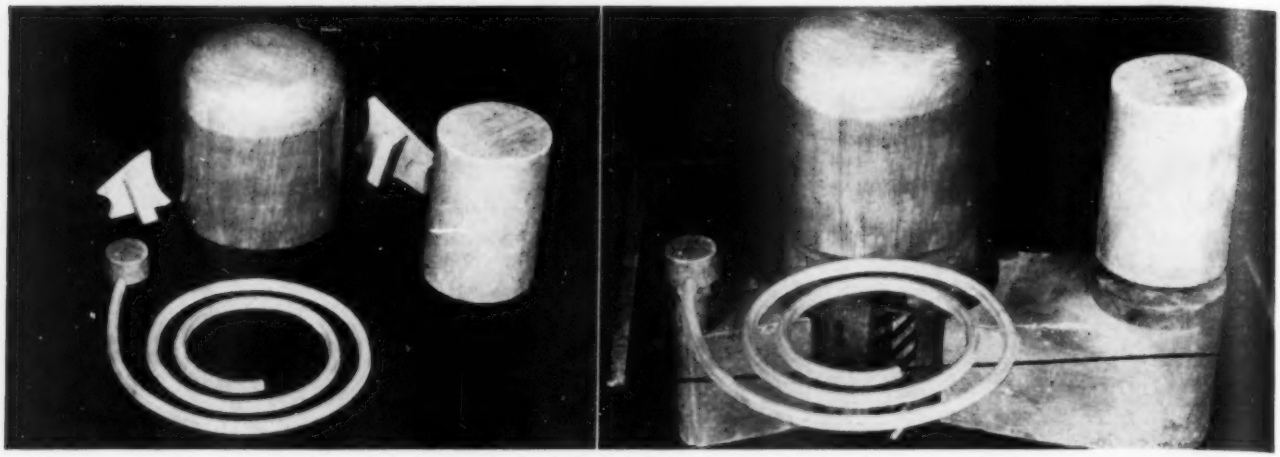


Fig. 11—Pattern for large fluidity test piece—(left) spaced properly for ramming in drag—(right) individual parts separated by glass plate (Taylor, Rominski, and Briggs).

ing basin and sprue to skim nonmetallic material from metal entering sprue.

Temperature Measurement for Laboratory Tests.

Lightly sheathed platinum, platinum-rhodium thermocouple inserted in pouring basin.

High-speed indicator or recorder used.

Fluidity Sampling from Bottom-Pour Ladles.

Mold designed so that all details except pouring basin are the same as the test mold for samples taken with spoons or from lip-pour ladles.

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MANGANESE BRONZE

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• Official Exchange Paper of the Institute of Australian Foundrymen presented at the Fiftieth Annual Convention of the American Foundrymen's Association, May 10, 1946, in Cleveland.
In this paper the author presents fundamental concepts of the physical metallurgy of Manganese Bronze.

THE alloys known as manganese bronze may, as a first approximation be regarded as derivatives of 58 per cent copper, 42 per cent zinc brass. Alloying elements may be aluminum, iron, manganese, tin, nickel and lead. It, therefore becomes clear why manganese bronzes are alternatively known as special brasses, high tensile strength brasses and mangenese brasses—on account of their high zinc content.

However, the name manganese bronze has also been applied to other alloys, e.g. alloys of copper and manganese containing 4 per cent manganese and the high tin bronzes to which was added iron and manganese. These latter alloys will have further reference made to them in the historical note. Again some alloys referred to as manganese bronze contain little or no manganese.

Chemical Composition

Today, the alloys referred to as manganese bronze are represented in chemical composition by the following broad specifications which, while it conveys the name of the constituents and their approximate amounts, does not really furnish specific information on the composition required to produce alloys of required mechanical properties. Here then is the general range of composition:

Composition

Element	*Per Cent*
Copper	55-65
Aluminum	0.2-5.0
Iron	0.2-3.0
Manganese	Trace-3.5
Tin	1.5 max.
Nickel	0-4.5
Lead	0-1.5
Zinc	Balance

In addition to copper and zinc, aluminum, iron and manganese are almost invariably present and when improved resistance to corrosion is required, e.g. resistance to sea water, tin is an essential alloying ingredient. In microstructure, two groups are recognized, the duplex alpha and beta with iron-zinc compounds and the essentially beta alloys with iron-rich compounds.

Historical Notes

Credit is due to P. M. Parsons for the invention or discovery of manganese bronze.

As a starting point in the history of these alloys it is noted that iron had been added to brass since early times—probably in many cases its inclusion was accidental. In 1779, Keir suggested a brass containing 3 per cent iron, and later two alloys known as Aitchs Metal and Sterrometal were introduced, but iron, while it was known to have a hardening and strengthening effect on copper-base alloys, had the disadvantage that it decreased the ductility.

Some beneficial effects were obtained by the introduction of manganese instead of iron using a cupromanganese obtained by reducing pyrolusite (manganese dioxide) in the presence of copper. No doubt the beneficial effects were due to the manganese functioning as a deoxidiser.

In 1876 P. M. Parsons added iron and manganese to copper-base alloys for which he was granted British Patent No. 482. The following sentence from it, is a more or less concise summary of the patent.

"Alloys of copper and tin or zinc, or both may be rendered more homogeneous, closer in texture, harder and stronger by adding a proportion of ferromanganese, spiegeleisen or other carburet of iron containing manganese."

Now examination of the above shows that the subject matter of the patent could, for convenience be divided into two groups each containing iron and manganese:

1. Those containing zinc very much in excess of tin.

2. Those containing practically copper and tin, the latter being of the order 8-10 per cent.

It was soon discovered that the manganese brasses, that is group (1) were entirely unsuitable for sand castings. They could only be cast into metal molds from which very satisfactory mechanically worked products could be produced.

Marine Applications

The manganese bronzes or group (2) could be readily cast into sand and became widely used. Thus the first vessel to have manganese bronze fittings was the S.S. Gleneagles which in 1877 had babbitted bearing brasses of copper, tin, manganese, iron alloy fitted to the engine.

In 1879 a comprehensive range of manganese bronze and manganese brass products including rolled sheets, forgings, rods, wire, castings for high pressure work and a cycloidal propellor for a torpedo boat were exhibited before the Institution of Civil Engineers at South Kensington Museum.

The first naval vessel to be fitted with a manganese bronze propellor was H.M.S. Colossus. This was in

1879 and in 1881 the S.S. Garth Castle of the mercantile marine, underwent trials with a manganese propellor. Results were so successful that two months later, manganese bronze superseded gunmetal for all future propellors for the British Admiralty.

Now as the copper-zinc-iron-manganese alloy could not be satisfactorily sand cast, it may be assumed without incurring great error that the early manganese bronze propellors were essentially an alloy of copper, tin, iron and manganese and on account of their high-tin content their good performances in sea water were to be expected.

The copper-zinc-iron-manganese alloys were again investigated from the aspect of sand casting them, no doubt economic considerations were of sufficient importance to prompt this as the high tin alloys contained tin which is more costly than zinc and their copper content was higher than the zinc-bearing alloys.

In 1888 Parsons was granted a second British Patent which covered the addition of aluminum to the alloys described in his previous patent so the manganese bronze became an alloy of copper, zinc, aluminum, iron and manganese—which is in general use today.

American Patents

In America, Parsons took out a patent in 1878. This was a duplicate of his early patent covering the addition of iron and manganese to copper base alloys and he apparently overlooked patenting the addition of manganese to the alloys. In America in 1879 Cowles was granted a patent for the addition of aluminum to these alloys.

The B. H. Cramp Co. of Philadelphia marketed a series of alloys in 1890 known as Vesuvius Metals. They were of the Manganese Bronze type and while they obtained some good results, the compositions as a whole were apparently not suitable as they were abandoned. In 1893 the same Company obtained from the Manganese Bronze and Brass Co. in England, of which Company, Parsons was the founder, the exclusive right to manufacture manganese bronze in America.

From then onward, there does not appear to have been any major set backs, but as may be expected in the case of a new multicomponent alloy, there soon appeared numerous alloys differing in minor points of composition but for which their sponsors claimed major improvements in physical properties. However, by 1914 compositions were, to some degree sorted out and reduced to a few standard ones.

Table 1

ALLOYS—BRITISH SPEC. 208—5 CLASSES

Alloy in per cent or Property	*1*	*2*	*3*	*4*	*5*
Copper	54	54	54	54	54 min.
Tin	Total	Total	Total	Total	Total
Manganese	5 max.	5 max.	5 max.	10 max.	13 max.
Iron					
Aluminum	Indiv.	Indiv.	Indiv.	Indiv.	Indiv.
Nickel	3 max.	3 max.	3 max.	3 max.	5 max.
Lead (max.)	0.5	0.50	0.50	0.50	0.50
Zinc	Balance	Balance	Balance	Balance	Balance
0.15 per cent Proof Strength, psi.	26,900	33,600	39,200	44,800	56,000
Ultimate Strength, psi.	62,700	71,700	76,200	85,100	100,800
Elongation, per cent	25	18	15	15	12

Commercial Casting Alloys

The casting alloys are divided into two classes:

1. Tin-bearing
2. Tin-free

Although the general statement of the specification covering these alloys does make provision for the tin bearing alloys, it has been thought better to divide them as indicated. Of the tin bearing alloys that most widely cast appears to be Admiralty Propellor Brass or BS. STA. 7—C.X. 3. Its composition and mechanical properties are as follows:

Element	*Per Cent*
Copper	55-57
Tin	0.7-1.4
Aluminum	0.05-0.5
Iron	0.5-1.5
Manganese	0.05-0.5
Lead	0.2 max.
Nickel	0.2 max.
Zinc equivalent	46 max.
Property	*Value*
Yield Point, psi	35,840
Ultimate Strength, psi	73,920
Elongation in 2 in., per cent	15

The tin-free alloys or more correctly, the alloys usually cast free of tin are covered by British Specification 208 of which there are five classes. Composition is not stated precisely, but certain limits and sums of alloying elements are specified. Details are given in Table 1.

The selection of the individual alloying elements mentioned, tin, manganese, iron, aluminum and nickel is left to the manufacturer. He is, however, obliged to state the ones he is using.

Some Detailed Compositions and Mechanical Properties. At this stage, it is intended to set out the composition of some alloys of nominated mechanical properties. It has been thought advantageous to do this at this stage so that readers may have something more concrete to consider rather than generalities.

Subsequently, the relationship of amount of alloying elements to structure and properties will be discussed at some length. Timms has listed the composition and mechanical properties of a series of alloys. They are shown in Table 2.

The mechanical properties claimed are minimum figures. Table 2 does not, by any means, cover the range of all the casting alloys.

Alloy Constitution

The Problem of Constitution. In general, there are three qualities or perhaps more, generally groups of qualities of an alloy about which information is required. They are:

Chemical Composition. This may be determined by chemical analysis, the amount and kind of alloying elements are determined.

Physical Properties. Determination of these is effected by tests on the metal made with the aid of instruments designed for the purpose. For example, the mechanical properties, "yield point," ultimate strength, and ductility are deter-

Table 2

COMPOSITION AND MECHANICAL PROPERTIES OF SOME COPPER-BASE ALLOYS

Alloy in per cent or Property	*1*	*2*	*3*	*4*	*5*	*6*	*7*
Copper	58.4	58.7	58.7	59.75	59.0	62.5	61.75
Aluminum	0.6	0.8	0.8	1.75	2.0	3.0	4.5
Manganese	1.2	1.2	1.75	2.0	2.0	2.5	2.25
Iron	0.8	0.8	0.8	1.5	1.5	2.0	1.5
Zinc	38.5	38.0	37.0	35.0	35.0	30.0	30.0
Tin	0.5	0.5	0.75	—	—	—	—
Equivalent Copper	56.8	56.6	56.7	55.5	54.5	54.64	50.5
Equivalent Zinc	43.2	43.4	43.3	44.5	45.5	45.36	49.5
Yield Point, psi	26,900	31,400	35,800	40,300	44,800	53,800	58,200
Tensile Strength, psi.	62,700	67,200	78,400	85,100	89,600	98,600	112,000
Elongation, per cent..	30	25	25	20	18	18	15

mined with a tensile testing machine.

Structure. The microstructure of the alloy may be ascertained by the use of a microscope and the amount of a particular phase estimated usually with an eye-piece divided into squares.

The problem of determining the relationship between these three qualities in general and more specifically the effect of variation of the five or six component elements has for convenience, been called the problem of constitution.

The first detail to concern one—and detail is a rather inadequate word to describe such a large problem—is the method of simultaneously considering the effect of five elements or six elements upon the structure of the alloy.

Constitution Diagram

In alloy systems, the general method of attack is basically the phase rule, and accordingly the binary constitutional diagram is used for alloys having but two-component metals, and for alloys of three-component metals the solid model is the formal method, but for convenience in representation in publication, and to further illuminate the problem, isothermal sections, and sections of constant composition, are used.

While three-component systems present their quota of difficulties, that of four and five-component systems are still more difficult. Therefore, recourse is made to the principle of equivalence—in the case of manganese bronze, Guillet zinc equivalent. The principle of equivalence is used in elucidating structural problems in alloys other than manganese bronze.

For example, in the metallurgy of high speed steel, use is made of the tungsten equivalent in selecting a constant tungstein section of the iron-carbon-tungsten system to represent the composition of commercial high speed steel, and again, in ferrous metallurgy the amount of alloying element required to close the gamma loop forms a basis for compiling a list of chromium equivalents.

Accordingly, it is now purposed to discuss in some detail the following aspects of constitutional approach:

1. Use, application and limitation of Guillet zinc equivalents using a two-component interpretation.
2. Description of a proposed method using zinc equivalents with a three-phase interpretation.
3. Provisional elucidation of constitution embodying the zone theory of solids and Hume-Rothery's electron concentration.
4. Probable effect of elements based on a comparison of the three-component equilibria copper, zinc and one third element, the third element considered in rotation.

Of necessity (2) and (3) and (4) will be brief as these methods are subject to further investigation by the writer.

Zinc Equivalents

Zinc Equivalents with Binary Interpretation. As indicated previously, this is the most usual method of considering the constitution of manganese bronze. It has already been said that manganese bronzes are derivatives of 58 per cent copper, 42 per cent zinc, brass. This brass has a duplex alpha solid solution and beta solid solution structure containing about 40 per cent of the alpha solution and 60 per cent of beta.

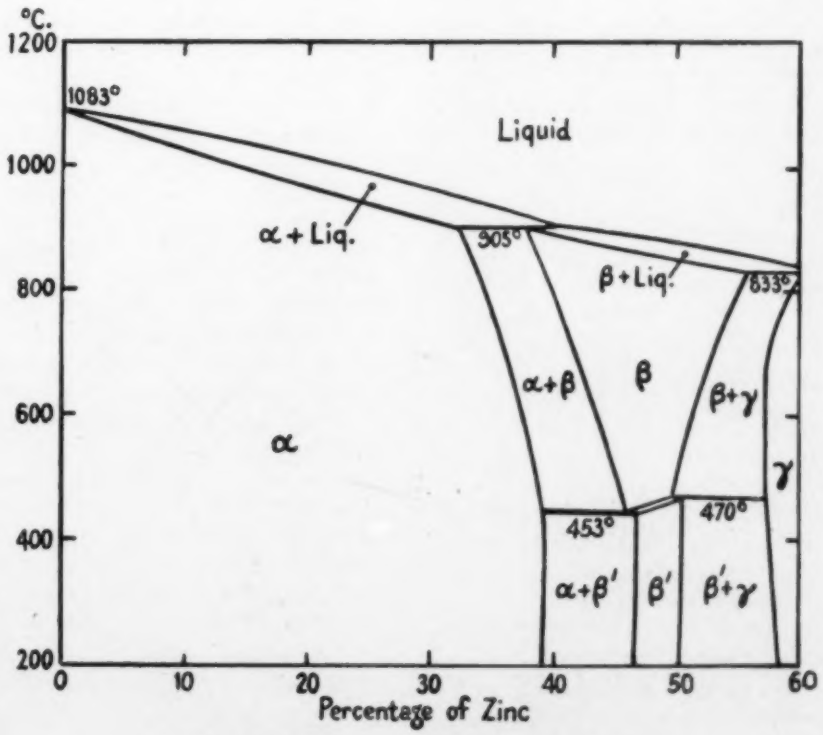


Fig. 1—Copper-Zinc Constitution Diagram.

In the "as cast" condition it has an ultimate tensile strength of approximately 47,000 psi. with 20 to 30 per cent elongation. Its properties, are however, sensitive to casting and cooling conditions.

The addition of elements such as aluminum, tin, iron and manganese to brass alters its microstructure and the zinc replacing capacity of the elements when added to brass were evaluated by Guillet in 1906, hence their name Guillet zinc equivalents.

In passing, it is well to refer to Guillet zinc equivalents because later workers have proposed different values for some of the elements and obviously different results will be obtained when using different values. This is important when the zinc equivalent of the alloy is specified. The zinc equivalents are as follows:

Element	*Equivalent Value*
Silicon	10
Aluminum	6
Tin	2
Magnesium	2
Iron	0.9
Manganese	0.5
Nickel	—1.3

Thus 1 per cent aluminum has a similar effect to 6 per cent of zinc, and tin a similar effect to 2 per cent of zinc, but the negative sign prefixed to the value for nickel, implies that 1 per cent of its functions in the same manner as 1.3 per cent of copper.

The use of zinc equivalents in determining the zinc equivalent of an alloy will be illustrated.

Composition:

Element	*Per Cent*
Copper	57.13
Tin	1.31
Aluminum	0.51
Manganese	0.46
Iron	1.02
Zinc	39.57

Tin Equivalent	1.31 × 2	= 2.62
Aluminum	0.51 × 6	= 3.06
Manganese	0.46 × 0.5	= 0.23
Iron Equivalent	1.02 × 0.9	= 0.92
Total		6.83
Zinc in alloy, per cent		39.57
Total Fictitious Zinc		46.40
Copper, per cent		57.13
		103.53

$$\text{Zinc equivalent of alloy} = \frac{46.40}{103.53} \times 100 = 44.8 \text{ per cent Zinc}$$

From the copper-zinc equilibrium diagram this alloy would have a microstructure consisting of alpha and beta.

The foregoing example has been set out in detail because some workers use the equivalents in a different manner. For example, consider the addition of 1 per cent aluminum to a brass containing 58 per cent copper, 42 per cent zinc. By the unorthodox method the matter of equivalent is treated as follows: as the zinc equivalent of aluminum is 6, the above alloy would have a structure similar to a brass containing 52 per cent copper, 48 per cent zinc. Using the orthodox treatment of equivalents it is found that the alloy has the following zinc equivalent:

Element	*Per Cent*
Aluminum	1
Zinc	41
Copper	58

$$= \frac{[41 + (1 \times 6)]}{(58 + 41 + 6)} = \frac{47}{105} \times 100 =$$

44.7 per cent zinc equivalent.

Therefore, it is well when viewing values of zinc equivalent quoted in the literature to ascertain by calculation the manner in which the published figures were obtained.

There is this to be said, however, that no doubt the workers using what has been referred to as the unorthodox method have included in their interpretation of Guillet zinc equivalents such local modifications of the principle of equivalents that they have found to apply to their practice—or in other words, the differences are not due to errors in arithmetic.

But be that as it may, it is well to retain a uniform treatment of Guillet zinc equivalents and not mix the originals with modifications as it leads to confusion.

The foregoing shows how the zinc equivalent of manganese bronze may be calculated. Then, from a copper-zinc constitutional diagram such as Fig. 1, it is possible to ascertain what phases may be present in the alloy and their approximate amounts. Reference to the copper-zinc diagram from the work of Owen and Pickup shows the following:

Phase Stable	*Zinc Content, Per Cent*
Alpha	Up to 38.1
Alpha and beta	38.1 to 46.2
Beta	44.2 to 49.5
Beta and gamma	49.5 to 60.2

Therefore, if the equivalent zinc content of an alloy exceeds 49.5 per cent, gamma may be expected in the microstructure, but this is, of course, from an equilibrium diagram for equilibrium rates of cooling, not the rates of cooling associated with sand castings. Information as to how the microstructure estimated by the use of zinc equivalents compares with that obtained by the microscope is available from the work of Logan[2].

Equivalent Zinc, per cent	*Alpha, per cent (by Zinc Equivalent)*	*Alpha, Per Cent (by microscope)*
42.7	40	41.6
42.9	37	35.6
43.3	33	31.3

The alloys he employed did not contain any added element in excess of 1.5 per cent and in general he found that with slowly cooled specimens, the alpha area as determined by the microscope came to within 3 per cent of the estimated value.

Working with a manganese bronze containing 1 per cent tin, 1 per cent iron, 0.25 per cent aluminum, 0.25 per cent manganese and varying copper and zinc to vary the microstructure. Murphy[3] illustrated the variation of mechanical properties with microstructure.

Alpha Phase, per cent	*Ultimate Tensile Strength, psi.*	*Elongation on 2 in., per cent*
100	46,500	24
78	53,400	36
70	61,000	50
45	69,200	42
25	76,600	28
0	83,400	21
0	54,600	4

The last alloy in the table contained delta in its microstructure which accounts for the low strength and poor elongation. The zinc equivalent has been included in the specifications BS./STA.7 — C.X.3. requires that the zinc equivalent shall not exceed 46 per cent (which figure is close to the upper limit of zinc content of the duplex alpha beta field).

Limitations of Zinc Equivalents. There are limitations to the usefulness of zinc equivalents.

1. When any of the added elements, e. g. aluminum, tin, iron, manganese exceed 2 per cent they are not reliable.

2. Zinc equivalence does not in any way indicate the probable effect of temperature on the structure of

the alloy. The values of zinc equivalent are given as constants, but for them to be true constants the phase boundaries must be straight lines and vertical lines. Calculated zinc equivalents apply to one temperature only.

3. Where an element does not enter the solid solutions and remains as discrete particles or forms an intermetallic compound, one would not expect the principle of zinc equivalence to apply.

4. There appears to be some doubt as to the values of zinc equivalent of some of the elements. Thus Guertler[4] and Vaders[5] consider silicon should have a value of 7 and not 10. Timms[1] considers that aluminum should have a value of 4, while Phillips[6] considers it should have a value of 2 instead of 6. Manganese should, according to Smalley[7], have a value of 0.8 rather than 0.5 Phillips[ibid] believes that 1.5 is more suitable value for tin than the usually accepted value of 2.

5. The two binary interpretation of a multicomponent alloy is neither accurate nor complete, because in a manganese bronze, there are frequently more than two solid phases present.

Three-Phase Representation

Zinc Equivalents Using a Three Phase Representation. This method is best illustrated by an example. Consider a tin-bearing manganese bronze of the type BS./STA.7—C.X.3, the zinc equivalent of the alloy less its tin content is computed, and reduced to value (100 per cent tin) in the alloy. This is then plotted on an isothermal section or sections of the copper-zinc-tin constitutional diagram.

The advantages of this method over the previous method are that one may contemplate the alloy as a three component system instead of a two-component system and this means that the relationship of phase fields containing the brittle gamma phase to the duplex alpha-beta phase is immediately clear.

Again, by using this method one can obviate the limitation of zinc equivalent, i.e. alloying element should not exceed 2 per cent. Consider an alloy containing 60 per cent copper, 4.5 per cent aluminum, 1.8 per cent manganese, 1.5 per cent iron, balance zinc. Here the aluminum exceeds 2 per cent so the zinc equivalent of the alloy less the aluminum content is computed and reduced and the composition of the alloy plotted on a section of the copper-zinc-aluminum diagram.

Isothermal and constant aluminum (or tin) sections are equally suitable for this work. There are two difficulties in the way of total applications of this method.

1. Equilibrium diagrams only are available to interpret cast structures.

2. Lack of concurrence among workers as to the position of phase boundaries in the three-component systems.

Zone Theory of Solids

Application of Hume-Rothery's Conceptions and the Zone Theory of Solids. The elements which are normally encountered in manganese bronze, and some of the impurities and some elements which have been tried in the alloy are included in the following elements:

Al
|
Sc - Te - V - Cr - Mn - Fe - Co -
Ni - Cu - Zn - Ga - Ge - As
|
Sn

The above elements are from the third row of the periodic table with aluminum from the second row and tin from the fourth row.

Some general rules governing the formation of primary (alpha) solid solutions were worked out by Hume-Rothery and although the manganese bronzes are essentially alpha beta or beta alloys, some interesting deductions which are not entirely without significance may be made concerning manganese bronze. Briefly enunciated, his rules are as follows:

1. *Size Factor.* The mutual solubility of one element in another is very restricted if the size of the atoms differ by more than 15 per cent.

2. *Electronegative valency effect.* In general, the more electronegative the solution element and the more electropositive the solvent, the greater is the tendency for the formation of stable intermetallic compounds.

3. *Relative valency effect.* Other things being equal a metal of lower valency is more likely to dissolve one of higher valency then vice-versa.

To the above may be added Hume-Rothery's later observation that there is a general tendency for the beta phases to be more restricted when the size of the atoms become larger.

Effect of Elements

Consider now the list of elements and the generalizations in relation to manganese bronze.

The first point of interest is that the two elements which confer the greatest hardening effect on manganese bronze are at the extreme end of the list. Tin and aluminum are referred to and a quantitative estimate of their strengthening effect is as follows:

Effect of 0.1 per cent of Elements on a Complex Brass Containing 57 per cent Copper:

Element, (0.1 per cent)	*Yield Point, psi.*	*Ultimate Tensile Strength, psi.*
Aluminum	+1180	+1460
Tin	+1400	+ 680
Iron	− 670	+ 50
Manganese	− 400	+ 330

As regards size factor all the elements are in the favorable range as regards copper, i.e. the atomic diameters do not differ by more than 15 per cent from that of copper. Concerning the electronegative valency effect, among the elements of group 6B is found sulphur which would tend to form a stable sulphide and in any case is regarded as an impurity.

In group 5 (b) are phosphorus, arsenic, antimony and bismuth which although still having the electronegative characteristic somewhat pronounced may form slight solid solutions. All these elements are regarded as impurities in manganese bronze.

Group 4 (b), contains silicon, tin and lead. Silicon is regarded as an impurity in manganese bronze and lead was for many years, but recently it has been incorporated in certain compositions while tin is a valuable alloying ingredient, but its amount and that of other elements notably aluminum and zinc must be carefully controlled when tin is used. It is seen therefore that passing from group 6 (b) to group 4 (b) the objectionable nature of the elements as regards their effect upon manganese bronze, decreases until a useful alloying element tin is reached.

Concerning the relative valency effect, when it is applied to manganese bronze, some rather surprising

tentative conclusions are drawn. It would appear that the elements of low valency may be incorporated in manganese bronze to a greater extent than the high valency elements before a brittle phase is encountered.

Nickel (valency 2) has a negative zinc equivalent and functions in the same direction as copper. Manganese and iron (valency 2) improve ductility. Aluminum (valency 3) has a strengthening effect but its amount has to be controlled otherwise a brittle phase is encountered.

Tin with a valency of 4 is limited to still smaller quantities otherwise brittleness is introduced. For a ternary alloy containing 60 per cent copper, gamma is encountered when the aluminum exceeds 3.0 per cent while in the copper-zinc-tin alloys for an alloy containing 60 per cent copper, gamma is encountered when the tin exceeds 1.3 per cent. Then the pentavalent elements, arsenic and antimony are impurities.

One passes now to the conception of valency electron to atom ratios or electron concentration. Briefly, Hume-Rothery has found that certain electron concentrations are associated with certain phases as follows:

Phase	*Electrons / Atoms*
Beta	$\frac{3}{2} = 1.5$
Gamma	$\frac{21}{13} = 1.615$
Eta	$\frac{7}{4} = 1.75$

and that the limit of solubility of several alpha solutions correspond to an electron concentration of 1.40.

Then Jones making certain assumptions and for a temperature of absolute zero employing the zone theory of solids arrived at the following values of electron concentration for the limits of stability of certain phases. They are as follows:

Phase	*Electrons / Atoms*
Limit of alpha phase	1.362
Limit of alpha and beta phase	1.480
Limit of beta and gamma phase	1.538
Limit of eta phase	1.70

The above limiting figures are the result of complex mathematical calculations.

In general, all the work of above type on fundamental structure of alloys has been for alloy systems wherein solute belongs to a certain group of the periodic classification. It has not been attempted to generally apply these results to complex alloys although Gregory and Bradley have found that the typical gamma structure may be obtained in ternary copper-zinc-aluminum alloys provided the composition is adjusted to keep the ratio of valency electrons to atoms at a value of 21 to 1.3.

Considering now complex alloys such as manganese bronze, the writer endeavored to ascertain whether or not these fundamental conceptions could be applied to elucidate some of the structural problems. At first sight, electron concentration would not have the limitation that an element should not exceed 2 per cent as in the case of the zinc equivalent.

Accordingly, compositions of manganese bronze, have been converted to atomic percentages and their electron concentration calculated and a comparison has been made of electron concentration of manganese bronzes containing 45 per cent alpha and 55 per cent beta with the electron concentration of ternary copper-tin-zinc; copper-zinc-aluminum; copper-manganese-zinc; obtained from the respective phase diagrams all having the same proportion of alpha and beta.

Results are as below:

Alloy	*Ratio: Valency Electrons / Atoms*
Manganese Bronze (a)	1.426
Manganese Bronze (b)	1.430
0.75% Sn, 58.15% Cu, 41.1% Zn	1.420
2.0% Al, 62.8% Cu, 35.2% Zn	1.428
2.24% Mn, 56.2% Cu, 41.56% Zn	1.430
Binary copper-zinc	1.416

It will be seen, therefore, that the ratios are of the same order.

Then an effort was made to see if any relationship could be found between electron concentration and mechanical properties, so the electron concentration of manganese bronze compositions from many sources was determined and compared with the percentage elongation as shown:

Elongation	*Number of Tests*	*Average Electron Atom Ratio*
Under 10%	7	1.505
10 to 15%	5	1.466
Over 15%	13	1.435

It would appear, therefore, that high valency electron to atom ratios are associated with low elongation. The point to be decided is, where is the line of demarcation?

It is realized that completed explanation of the constitution and mechanical properties of the alloys on the basis of these theories is militated against by the following:

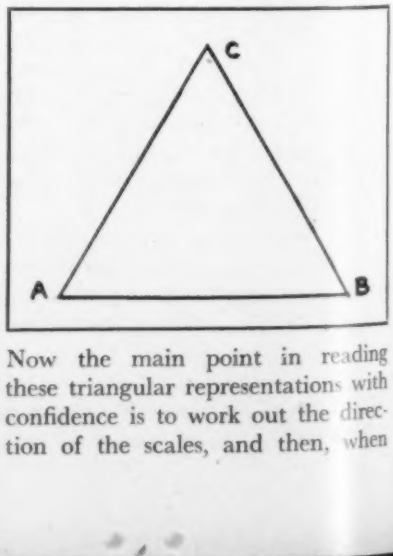
1. Effect of cooling rate on mechanical properties, i.e. non-equilibrium conditions.
2. The value of valency electrons for the metals composing manganese bronze. While a value of 1 for copper, 2 for zinc, 3 for aluminum and 4 for tin is well accepted, the writer is not prepared to use the value of zero for iron, nickel (and other elements of group 8).

Nor is he prepared to use a value of 4 for lead because it is not in solution and in any case, weakens the alloys mechanically. The value of manganese is also not finalized. Values used by the writer in foregoing calculations are those generally accepted, with the addition of nickel one, manganese two, iron two.

3. Accuracy of published analysis.
4. Accuracy of three component phase data.

Phase Rule Application

Application of Phase Rule Data to Structure of Manganese Bronze. In a paper such as this, one may not enter into a full discussion of three component systems so all that will be done is to indicate a simple method of reading these triangular representations and to show generally how the position of the boundaries between the different phases. If an alloy contains three metals, say A, B, C, then draw an equilateral triangle.



Now the main point in reading these triangular representations with confidence is to work out the direction of the scales, and then, when

Table 3

DATA OBTAINED FROM TRIANGULAR DIAGRAM

Component being discussed is.....	A	B	C
To find its percentage, draw lines parallel to	BC	CA	AB
Read off percentage of component along either of the following sides.....	CA or BA	CB or AB	BC or AC
The amount of component increases in magnitude in the direction.....	C to A or B to A	C to B or A to B	B to C or A to C
Amount of component is zero at.....	C and B	C and A	A and B
Component is 100 per cent at.....	A	B	C

using the diagram progress in the same direction. Some of the data obtained from the diagram is summarized in Table 3.

After drawing such triangles, once or twice, the use of them will become relatively easy.

Now the question arises as to what information may be obtained concerning the probable behavior of elements from their ternary constitutional diagrams with zinc and copper. Such information is conveniently obtained from a study of the isothermal section at room temperatures. The feature upon which attention is focused is three-phase field containing alpha solid solution, beta solid solution, and gamma solid solution.

The gamma solution generally imparts brittleness and it is of practical importance to know how close the boundary between the three phase field and the duplex alpha-beta field comes to the binary copper-zinc side of the triangle. If this boundary is encountered after, for example, 1 per cent of the element has been added, then 1 per cent represents the limit of this particular element.

Alloying Elements

Details of the elements used in manganese bronze will now be discussed and we will use as our arbitrary criterion of assessment a composition containing 60 per cent copper.

Nickel. As previously mentioned, nickel has a negative zinc equivalent or in other words, it functions in the same direction as copper so when added to brass one expects a considerable range of alpha and beta solutions. Reference to copper-nickel-zinc constitutional diagram for a temperature of 25° C. shows that when nickel is added to an alloy containing 60 per cent copper, a field containing gamma is not encountered at all. See Fig. 2.

Aluminum. It will be recalled this element has a high-zinc equivalent, 6, but when the isothermal section at 410° C. of the copper-zinc-aluminum system is examined, it is seen that for an alloy containing 60 per cent copper the gamma phase (or more fully) the duplex beta-gamma phase is not encountered until approximately 4.5 per cent aluminum has been added. It will be recalled that certain high strength manganese bronzes contain 4 to 5 per cent of aluminum.

Manganese. Manganese has a low-zinc equivalent, 0.5, and replacement of zinc by manganese, as it were, decreases the zinc equivalent of the alloy. In the isothermal section at 400° C. more than 7½ per cent of manganese may be added before a third phase other than alpha and beta is encountered.

However, an isothermal section at 360° C. (by the same workers, Bauer and Hansen) shows that at this lower temperature, a three-phase field alpha-beta appears, but is not encountered in an alloy containing 60 per cent copper, but is encountered at approximately 4.5 per cent manganese when the copper content is reduced to 59 per cent.

Tin. The writer regards the three-component copper-tin-zinc system as the most important diagram as it concerns the structure and properties of the tin bearing alloy C.X.3. Tin has a zinc equivalent of 2, and for an alloy containing 60 per cent copper, the three-phase region containing gamma is encountered at approximately 0.7 per cent tin (and at 56 per cent copper, at 1.3 per cent tin).

More will be said about this later, but at this stage it would not be out of place to say that because gamma is encountered at relatively low tin contents, is the reason why many writers dislike tin in manganese bronze and why foundrymen have trouble with low elongation.

Now, with the exception of lead and iron, which will be reserved un-

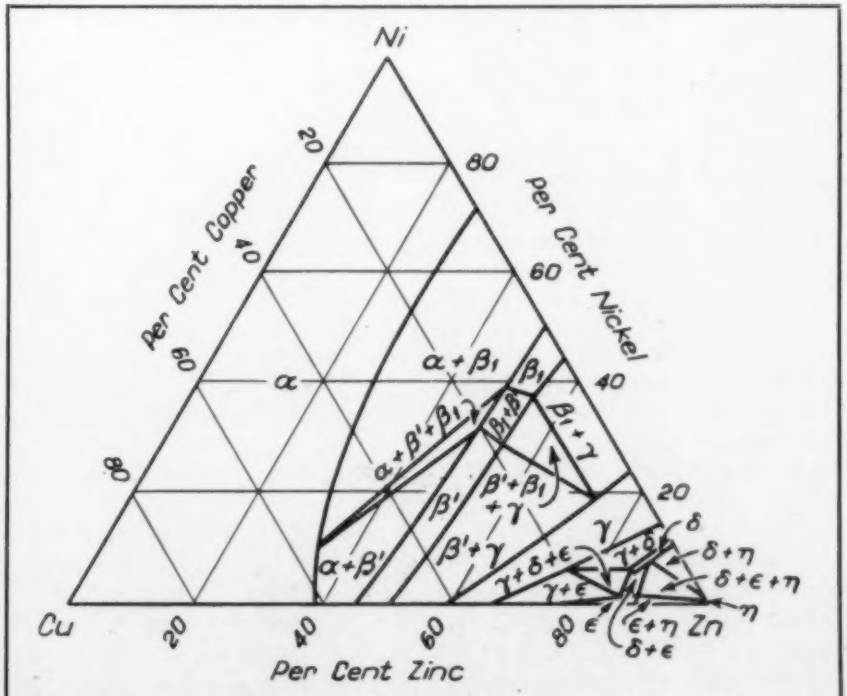


Fig. 2—Copper-Nickel-Zinc Constitution Diagram for a temperature of 25° C.

Table 4

PHYSICAL PROPERTIES OF CHILL CAST ALLOYS

No.	Cu, per cent	Zn, per cent	Al, per cent	Fe, per cent	Mn, per cent	Sn, per cent	Yield Point, psi.	Ultimate Strength, psi.	Elongation on 2 in. per cent
8.....	56.81	40.51	1.12	0.34	0.74	0.48	34,200	84,750	21.5
9.....	56.45	41.13	0.66	0.42	0.80	0.54	32,400	71,850	22.3
10.....	56.65	39.54	1.10	0.46	0.80	1.51	—	70,650	3.0
11.....	58.69	40.36	0.63	0.11	0.60	1.41	43,000	83,800	16.3
12.....	56.73	39.72	1.16	0.31	1.65	0.43	38,250	89,150	23.8
13.....	56.96	40.46	0.49	0.17	1.56	0.36	34,500	81,925	21.0
14.....	56.48	38.99	0.47	0.73	1.73	1.64	49,000	91,750	21.0
15.....	57.06	39.74	0.97	0.95	0.82	0.46	35,327	87,600	29.8
16.....	56.76	40.52	0.51	0.95	0.80	0.46	29,500	80,250	38
17.....	57.12	38.78	0.99	0.79	0.76	1.56	48,000	44,375	22.3
18.....	57.15	39.37	0.47	0.70	0.72	1.59	38,000	85,250	27.0
19.....	56.6	39.77	0.93	0.81	1.59	0.30	37,250	75,125	10.0
20.....	56.87	39.83	0.46	0.84	1.62	0.38	28,500	83,150	20.0
21.....	56.9	38.10	0.95	0.95	1.50	1.60	52,600	92,500	14.0
22.....	56.6	38.99	0.47	0.73	1.65	1.56	43,375	85,175	16.4

EFFECT OF ELEMENTS ON PHYSICAL PROPERTIES

	Cu	Zn	Al	Fe	Mn	Sn	Yield Point	Ultimate Strength	Elongation on 2 in.
8 minus 9.....	+.36	—.62	+.46	—.08	—.06	—0.06	+1,800	+6,900	—0.8
12 minus 13.....	—.23	—.74	—.67	+.11	—.09	—0.07	+3,750	+7,225	+2.8
15 minus 16.....	+.30	—.78	+.46	Nil.	+.02	Nil.	+5,827	+7,350	—8.2
17 minus 18.....	—.03	—.59	+.52	+0.09	+.04	—0.03	+10,000	+9,125	—4.7
21 minus 22.....	+.30	—.89	+.48	—.22	—.15	+0.04	+9,225	+7,325	—2.3
Sum of Differences	+.70	—3.62	+2.59	+.37	—.06	+.02	+30,602	+37,925	—13.2

til later, foregoing are elements used in manganese bronze. It is now intended to examine the phase relationship with copper and zinc, of those elements normally regarded as impurities. Unfortunately, there does not appear to be suitable isothermal sections at low temperature of copper and zinc with phosphorus, and with bismuth, and with antimony, and with arsenic, but from data available a limited discussion of silicon and cadmium may be entered into.

Silicon. Smalley and others, regard this as a most objectionable impurity. It has a high zinc equivalent of 10. The isothermal section due to Vaders (which does not go to copper contents as low as 60 per cent) indicates that a brittle phase is formed and with lower concentration of silicon as the copper content is decreased.

Cadmium. Cadmium is also regarded as an impurity and reference to the isothermal section at room temperatures shows that after approximately 0.4 per cent of cadmium has been added to an alloy containing 60 per cent copper, the stable atomic grouping Cu_2Cd appears in the field. Therefore, indications are, that as the boundary between the three phase field containing the brittle phase and the duplex field (alpha and beta) becomes closer to the copper-zinc side of the diagram, the practical consequences are that less of the third element may be added, that its amount must be more rigidly controlled and that if it comes sufficiently close, the element must be regarded as an impurity.

Lead and Iron. These elements do not enter into solid solution, but exist as discrete particles in the case of lead, and iron-rich compounds in the case of iron. Bismuth may be expected to behave in a similar manner.

To conclude the survey of the elements, titanium, vanadium, and possibly chromium may be mentioned as they appear to raise the tensile strength and reduce the elongation, yet not render the alloy brittle. It is of interest and possibly significance that these elements when alloyed with copper give rise to precipitation hardening alloys.

Before leaving the subject of constitution, it is well to discuss briefly the zinc-iron compounds, zinc-rich compounds, or blue-etching constituents as it is sometimes called. There is no finality regarding the precise composition of these compounds and they may be expected to vary in manganese bronze of different compositions.

Dix, in 1922 made an investigation into the nature of these compounds and some general conclusions have been reached. They are:

1. Melts containing less than 0.3 per cent iron were found to contain little or no "blue constituent."

2. The essential constituents of the compounds are iron and zinc. To these some would add copper, while others consider it to be still more complex and contains in addition tin and aluminum.

3. The occurence of these "iron-zinc compounds" does not impair the mechanical properties of the alloy unless they be segregated. At this stage it may be relevant to note that cases are on record where segregation of these "iron-zinc compounds" has brought about corrosion failure.

4. These constituents are not dissolved by heat treatment, e.g. soaking 7¾ hr. at 850° C. and water quenching.

An indication that these "iron-zinc compounds" may not be stable atomic groupings in the usual sense of the term is provided by modern views on the structure of metals. Iron is ferromagnetic because the group of 18 electrons is not yet completely filled and that the incompletely filled shell has a critical size compared with the distance between adjacent atoms.

That is to say, that insofar as a group of electrons can be said to have a size, it may be said that the atoms in iron are just sufficiently far

Table 5

EFFECT OF 0.1% OF AL, FE, MN, AND SN ON PHYSICAL PROPERTIES OF 51% CU ALLOY

Element	Proportional Limit	Yield Point, psi.	Tensile Strength, psi.	Elongation, per cent	Reduction in Area, per cent	Rockwell B Hardness
Al	+1,280	+1,180	+1,460	−0.51	−0.49	−0.49
Fe	+940	−670	+50	+1.06	+0.93	−1.42
Mn	+180	+400	+330	−xx	−0.58	+0.01
Sn	+690	+1,400	+680	−0.64	−0.51	+0.42

NOTE: Effect of manganese on elongation first gives rise to an increase, then a decrease.

apart to produce the ferromagnetic effect. If they were closer together there would be no ferromagnetism and if they were further apart, they would be too far apart to influence one another.

Now manganese bronze is very slightly magnetic, this has been shown by observation—and if this slight magnetism be viewed as a case of ferromagnetism resulting from the small iron content of the alloy it would indicate that the iron has formed a somewhat loose union with the zinc and other elements which may be present in the complex.

Effect of Elements on the Physical Properties. Having discussed the general effect of the added elements on the constitution, it is logical now to discuss their effects on the physical properties. Instead of having to consider the effects of each element on the property of a brass containing 58 per cent copper, 42 per cent zinc, Ellis by rather ingenious method of combination and elimination succeeded in isolating the effects of each element using manganese bronze alloys.

He published his findings as curves, by means of which it is possible to calculate the mechanical properties of an alloy from its composition and he found surprisingly good agreement. He worked on chill cast alloys. An example will be taken from his paper to illustrate the methods employed. See Table 4.

The alloys are arranged in pairs so that the effect of aluminum can be estimated. There are differences in proportion of the other elements so it is not possible to determine the effect of aluminum from one pair of alloys so the differences between pair is averaged as shown.

By dividing by 25.9 which is ten times the sum of the differences of aluminum content, he obtained the approximate effect of 0.10 per cent aluminum on the alloy containing 57 per cent copper.

Result:

Differences in percentages—	
Copper	+ 0.027%
Zinc	− 0.140%
Iron	+ 0.014%
Manganese	− 0.002%
Tin	+ 0.001%
Limit of proportionality	+ 1280 psi.
Yield Point	+ 1180 psi.
Ultimate Tensile Strength	+ 1460 psi.
Elongation	− 0.51%
Reduction in Area	− 0.49%
Rockwell B	+ 0.49

In a similar manner, the effect of 0.1 per cent of other elements on an alloy containing 51 per cent copper, may be obtained and the results are summarized in Table 5.

The method of using data from Ellis' curves to estimate the mechanical properties of an alloy from its composition is illustrated as follows:

Composition:

Element	*Per cent*
Copper	51.95
Iron	3.82
Aluminum	1.88
Manganese	0.98
Zinc	Balance

From Ellis' curve of binary copper-zinc alloy, the tensile strength of a brass containing 51.95 per cent copper is seen to be 72,700 psi. and from an appropriate curve, it is seen 3.82 per cent of iron reduces its strength by 6000 psi. Similarly 1.88 per cent aluminum is seen to raise its strength by approximately 2600 psi. and that 0.98 per cent manganese raises its tensile strength by 1400 psi. Therefore, estimated tensile strength is:

Strength of simple brass, psi	72,700
Increase due to 1.88 per cent Al, psi	26,000
Increase due to 0.98 per cent Mn, psi	1,400
	100,100
Decrease due to 3.82 per cent Fe, psi	6,000
Estimated tensile strength, psi	94,100

Strength of the foregoing alloy from a chill cast test bar was 92,200 psi. i.e. estimate was 2 per cent too high.

The percentage elongation is estimated thus:

Elongation of the simple brass, per cent	19
Increase due to Fe, per cent	13
Increase due to Mn, per cent	5
	37
Decrease due to Al, per cent	20
Estimated elongation, per cent	17

The actual elongation was found to be 18.5 per cent. The accuracy of estimates of strength and elongation is shown by the following averages of 36 tests (due to Ellis).

	Actual	*Estimated*
Ultimate Tensile Strength, psi	78,000	79,200
Elongation, per cent	31.3	29.1

The practical value of being able to estimate the mechanical properties of an alloy from its composition will be self evident, it being particularly useful when synthesizing new alloy.

It is not proposed at this stage to discuss the further detailed effects of the elements on the physical properties of manganese bronze because it is felt these will be covered when the three examples, tin-bearing, high aluminum and lead-bearing are discussed, but on the other hand there appears to be merit in endeavoring to summarize the effects of the elements.

From a qualitative point of view, aluminum may be regarded as the element which confers the strength, so does tin and in addition, confers resistance to dezincification, the iron increases the ductility without effecting a reduction in strength (conferred by the aluminum) and manganese increases the strength without reducing the ductility.

Following this discussion of constitution and effects of elements on mechanical properties, it has been thought well to mention briefly some of the more practical features of manganese bronze.

Alloying Manganese Bronze

Surveying the literature, it is seen there are quite a few recipes for alloying manganese bronze. Rather than catalogue them, an effort will be made to pick out the main points.

1. The general principle of hardeners is used extensively to facilitate alloying of high melting point ingredients and in this regard much ingenuity has been displayed in selecting combinations. Typical examples are, iron and ferromanganese, iron-tin-manganese, iron-aluminum, cupromanganese, iron - manganese-copper.

2. As ferro-alloys may be used in preparing some of the hardeners it is desirable to select alloys of low carbon content or take steps to reduce the carbon content for carbides may persist in the finished alloy as lumps of varying size.

3. Consideration is given to employing usefully any heat of formation which may be evolved, e.g. when aluminum is added to copper, any rise in temperature is employed to dissolve the hardener.

4. Prior to zinc additions, the al-

loy is cooled to avoid excessive losses.

"Deoxidation." The process of deoxidation about to be discussed appears to be somewhat singular for after the metal is alloyed, some workers add phosphor-copper prior to pouring. The reasons for this appear to be:

1. Addition of phosphorus appears to inhibit the oxidation of the surface of the metal while pouring, such beneficial action is considered by some to be due to the formation of reducing envelope of volatile phosphorus surrounding the metal stream.

2. The addition of phosphorus prior to pouring would probably assist any aluminum oxide present to rise to the surface, for such a method is described by Thews[11] for refining scrap brass contaminated with aluminum.

Concerning orthodox deoxidation, it would appear beneficial to deoxide the molten copper prior to making additions and to as far as possible make progressive deoxidations to ensure that there is the minimum of oxygen present when aluminum and particularly tin is added.

Fluxes Used

Fluxes. It is not intended to discuss fluxes at any great length. Boron compounds are frequently employed. They are borax, boric oxide and borocalcite. The relative solubility of oxides in borax and boric oxide is indicated by the following table which, however, does not give the limit of solubility in the case of zinc, cadmium, aluminum and iron because only two grams of the oxide in question were heated with twenty grams of flux.

Weight of oxide taken— 2 grams
Weight of flux —20 grams

Oxide used—	*Borax, grams*	*Boric oxide, grams*
Zinc	2.0	0.56
Cadmium	2.0	0.84
Nickel (NiO)	0.62	0.62
Aluminum	2.0	None
Tin	0.84	None
Copper (CuO)	1.45	None
Iron (Fe_2O_3)	2.0	None
Chromium (Cr_2O_3)	0.001	None

Borax has, therefore, a greater solubility for the heavy metal oxides than boric oxide, but boric oxide will dissolve silica resulting in corrosion of the crucible. Sometimes boric acid is mixed with alkali carbonate and when the latter is used in excess it will flux any acidic or amphoteric oxides which may be present, e.g. sodium carbonate will dissolve zinc oxide, but not tin oxide. Glasses (complex borates) are used as covers and certain varieties (green bottle glass) may dissolve heavy metal oxides, e.g. zinc.

Borocalcite has the advantage of crucible corrosion and is an efficient solvent for the reduced oxides.

Common salt (sodium chloride) has also been used, but its action is probably solely that of a cover. Two compositions of flux which have been used in reverberatory furnace melting of manganese bronze are:

	Per cent
1. gypsum (calcium sulphate)	50
fluorspar (calcium fluoride)	50
2. soda ash (sodium carbonate)	30
Silica	20
Fluorspar	33
Borax	17

It should be noted, however, that Clamer has pointed out that while calcium sulphate is not reduced by any of the metals in manganese bronze with possible exception of aluminum, but calcium sulphate mixed with charcoal or in contact with the graphite of a crucible is reduced to calcium sulphide and sulphur from that source enters the metal. Concerning charcoal, it is not considered by some to be good practice to use it with manganese bronze as it promotes the formation of the carbides.

Effect of Pouring Temperatures. The effect of pouring temperature on the mechanical properties obtained from test bars has been reported upon by several workers and the range is reasonably wide. The following figures are due to Mochrie[12] and refer to BSS 208 Class 1. In addition there is qualitative observation concerning zinc volatilization and fume production.

Pouring Temp. °C.	*Tensile Strength, psi.*	*Elongation on 2 in. per cent*	*Amount of Fume*
1090	71,600	40.6	High
1020	71,600	40.6	High
970	71,200	43.7	High
940	71,100	34.4	Low
920	71,300	42.2	Low
890	71,500	40.6	Slight

Guillet has found a range of 997° C. to 1163° C. satisfactory, but recommends 1066° to 1093° C. Hesse, Myskowski and Loring found 950° C. to 1025° C. satisfactory.

For C.X.3, pouring temperature of 930° C. to 1060° C. have been found satisfactory.

Control of Alloys during Melting. When making castings which require large quantities of metal, e.g. propellors of size for S.S. "Queen Mary" it is, of course, highly desirable, if not imperative, to be able to assess the quality of the metal as melting proceeds and before it is pouring. A summary of the methods which have been used is as follows:

1. *Fracture of a Chill Cast Bar.* By its use Murphy[3] claims the zinc content may be estimated to within 0.5 per cent and when using the test, it is well to prepare first a set of standard fractures of known zinc content. As a control these would appear to have wider application to the beta brasses rather than the duplex alpha-beta brasses.

2. *Torsion Test.* This rather ingenious method was used by LeThomas[15] for controlling the zinc content of a copper-zinc, 5 per cent nickel, 5 per cent manganese alloy, the angle of twist at which rupture occurred being related to the zinc equivalent. The following values no doubt apply only to the alloy above.

Zinc Equivalent Per cent	*Angle of Rupture, in deg.*
40	260
45	160
50	123
55	108
60	95

Combined Chemical Analysis and Microstructure. Davis[16] suggests that a microscopic examination be used in conjunction with the analysis for zinc and copper, thus:

$$\text{Zinc equivalent} = \frac{465 - a}{10}$$

where 'a' is the percentage of alpha present.

From the practical angle, it was the disadvantage that the accurate determination of zinc is not a rapid method which may detract from its value as a control test.

Hardness. Ensslen[17] uses the hardness of chill cast test bar as a means of control, certainly there appears to be a relationship between hardness and tensile strength, while Roberts[18] considers that the addition of 1.5 per cent zinc raises the Brinell Hardness by 5, but the writer has

not found it sufficiently linear in its relationship to strength to be used unreservedly as a control test.

No doubt these methods of control during melting have practical value, but they require considerable experience in the nature of preliminary testing before they become established new routine in any foundry.

Heat Treatment

Heat Treatment of Manganese Bronze. First, stress relieving of castings will be discussed. The practical minimum temperature appears to be 400° C. and the importance of time and temperature is well illustrated by work of Benson and Allison[19] who worked with an alloy having a yield point of 48,900 psi., ultimate strength 94,700 psi., and elongation 28 per cent.

Heat Treatment, 3 hr. at temp.	*Initial Stress, psi.*	*Stress Relieved by Heat Treatmene, per cent*
300	10,520	77.5
350	10,520	92.5
250	11,000	54.5
200	11,200	23.5
400	9,200	100

Of practical importance is the rate of cooling from the temperature, if it is too fast the stresses which were relieved by heat treatment will be put back into the work.

Stress relief of castings is important where they are to be used for precision work, e.g. optical work in order that permanency of dimensions is retained. It is important also, where manganese bronze, particularly beta alloys, are subject to action of low melting point metals and alloys as in tinning. Should there be any appreciable stress in the alloy it will fail by cracking.

The phenomenon is well illustrated by the work of Dickenson[20]—some of his results are as follows: He took bars 10 mm. sq., supported them at the ends and applied central load, provision was made for heating the bars and applying the different low melting point metals. The load to fracture them and the deflection were measured and are summarized in Table 6. The letters "UB" signify the specimen was unbroken. It will be seen that the application of a fluid low melting point metal greatly reduces the load and deflection before fracture occurs.

After burning, castings could with advantages be stress relieved, as burning may induce stresses of the same order as the elastic limit.

It is now intended to discuss heat treatment which has for its object the alteration of mechanical properties. The published work on this matter is very sketchy, consisting almost entirely of heat treating a test bar and comparing the results obtained with "as cast" bars made at the same time. If the heat treatment of manganese bronze is viewed from the fundamental angle, it will be seen that there are four separate and possible "heat treatment reactions" which may influence the final result.

1. *The effect of heat treatment on the proportion of alpha and beta.* If, by the application of Guillets equivalent, the complex alloy is viewed as a two component alloy consisting of equivalent zinc and equivalent copper and its composition referred to the copper-zinc diagram, it will be seen that certain compositions may have the amounts of alpha and beta varied by heat treatment.

Thus an alloy with equivalent zinc content of 40 per cent contains approximately 40 per cent alpha and 60 per cent beta. At temperatures above 500° C. the solubility of zinc in copper forming beta solution increases so that at 750° C. the alloy is entirely beta. On cooling alpha separates, and the amount retained in the final structure is a function of the cooling rate, the faster the cooling, the less the amount of alpha and the stronger will be the alloy. Binary alloys containing 38-46 per cent zinc behave similarly.

2. *Possible Precipitation Hardening.* Matsuda[21] has reported upon the precipitation hardening which occurs when certain copper-zinc alloys are reheated after quenching. This results from the precipitation of alpha and the phenomenon is most pronounced in alloys containing 40 per cent of zinc.

With less than this amount of zinc, quenching and tempering effects are less pronounced owing to the difficulty of completely retaining them as beta.

On the other hand the effect is not so pronounced in alloys containing more than 40 per cent zinc for it has been found, when they are quenched to retain the beta structure, the resultant structure does not differ greatly from that obtained by slow cooling and also by virtue of the fact that beta of higher zinc content is more stable.

The reheating temperature for quenched alloys containing approximately 40 per cent zinc is 250 to 350° C. which produces a large quantity of fine particles of alpha.

3. *Relationship of Phase Volumes Containing a Brittle Phase to the Binary Copper-Zinc Phase of the Diagram.* Here the complex alloys are viewed as being composed by three components, not two, and the volumes containing brittle constituents with which one is concerned are:

a. Duplex, alpha and gamma.
b. Three phase, alpha, beta and gamma.
c. Duplex, beta and gamma.

The significance is best explained by diagrams but as a general rule,

Table 6

FRACTURE LOAD AND DEFLECTION OF 10 MM. SQ. BARS SUPPORTED AT ENDS

Condition	*Temp., ° C.*	*Test*	*Beta As Cast*	*Alpha Beta As Cast*
Uncoated	16	Load, lb.	2860	1980
		Deflection, in.	0.407	0.290
Mercury coated	14	Load, lb.	440	880
		Deflection, in.	0.006	0.016
Uncoated	105	Load, lb.	2640	1925
		Deflection, in.	0.553 UB	0.512
Coated with alloy	103	Load, lb.	835	1100
		Deflection, in.	0.010	0.034
Uncoated	250	Load, lb.	1980	1650
		Deflection, in.	0.563 UB	0.496
Coated with tin	245	Load, lb.	220	1056
		Deflection, in.	0.003	0.042

the closer the boundary comes to the limit of composition, with respect to the component other than copper and zinc, e.g. tin in case of C.X.3., the more likely is this factor liable to influence the results of heat treatment.

In the case of the tin-bearing alloys C.X.3 it is the three phase field which is more likely to influence the process and in the case of the high aluminum alloys H.S.B. 44 it is the duplex beta and gamma field.

4. *Effect of Heating on the Amount of Iron-Zinc Compounds.* One cannot be dogmatic concerning this aspect of heat treatment. Although from the work of Dix[8] it would appear that their solubility is not affected by heat treatments. He found the amount was unaltered in specimens after water quenching from 850° C. after 7¾ hr. heating.

In others it was slightly rounded and less in amount after the same heat treatment. However, it is listed as a possibility for one is not yet sure that the nature and stability of the compounds is not independent of total composition of the alloys in which they occur.

Heat Treating Effect

Detailed discussion of heat treatment will be given when dealing with examples of types of alloys but the possible general effects of the four reactions may be somewhat as follows:

Duplex alpha beta alloys (tin free) e.g. BSS 208/1. Results of heat treatment would be affected by (1) (2) and possibly (4).

Duplex alpha beta alloys containing tin, e.g. C.X.3. Results are affected by (1) (2) (3) and possibly (4).

Essentially beta alloys, e.g. HSB 44. Results are affected by (3) and possibly (4); if small amount of alpha present, also by (1).

Finally, if manganese bronzes are held at temperatures over 700° C. for an undue length of time, there is grain growth resulting in very large crystals. This has been reported on by Logan[2] and confirmed by the writer.

Discussion of Three Types. In the foregoing, constitution, effects of elements on mechanical properties, and some points of technology were discussed. Now it is proposed to discuss three types of manganese bronze with which many foundrymen have a considerable amount to do and use the foregoing as a basis of discussion. They are

1. Lead-bearing alloys
2. High-aluminum alloy H.S.B. 44
3. Tin-bearing alloy C.X.3.

Lead-Bearing Alloys

Survey of the literature shows that opinion has been divided upon the effects of lead in manganese bronze, the limiting proportion and the form in which it occurs. Space does not permit a review of the work, but an attempt will be made to state the present position and highlight the reasons for conflict of opinion.

It is recognized that lead lowers the mechanical properties of the alloys but does not necessarily lower them so much as to be useless. Indeed examination of some of the examples of composition and mechanical properties purported to show the poor mechanical properties even in the absence of the lead. In other words, the manganese bronze was bad to start with.

The recent work of Hesse, Myskowski and Loring[14] and Halliwell[22] is the most comprehensive and thorough, in that the composition as a whole is studied. From his work which included 483 tests, Halliwell concluded that lead up to 0.6 per cent lowers the ultimate tensile strength by 60,000 psi. and lowers the elongation by 10 per cent and the Brinell Hardness by 10. The effect of 1 per cent on alloys containing 1 per cent aluminum, 1.3 per cent iron, 0.5 per cent manganese and copper is shown in the following table.

EFFECT OF LEAD

Property	*57% Copper*	*62% Copper*
Tensile Strength, psi...	−6600	−2800
Yield Point, psi.....	−2600	−1010
Elongation, per cent....	−6.0	−6.0
Brinell Hardness.....	−5.0	−6.5

Permissible lead content has been related to copper content by Halliwell and to tin and copper and zinc ratio by Hesse, Myskowski and Loring which indicates a much more fundamental approach particularly in view of the fact that in order to utilize lead-bearing scrap, certain specifications, ASTM 132-41 T, embody up to 1.5 per cent lead.

The conclusions of Halliwell show copper and lead range for different mechanical properties.

Tensile Strength, psi.	*Elongation on 2 in., per cent*	*Copper, per cent*	*Lead, per cent*
65,000	20	56/62	0.15
65,000	20	57/60.5	0.40
65,000	25	56/60.5	Trace
65,000	25	58.5/60	0.4
60,000	20	56/62	0 to 0.2
60,000	20	58/61	0.2 to 0.6
60,000	20	58.5/61	0.6 to 1.5

Hesse, Myskowski and Loring have related the lead content to the balance of composition as follows:

Lead, per cent	*Copper-Zinc Ratio*	*Tin, per cent*
0.7	1.20-1.25	Under 0.6
0.0	1.20-1.25	1.0
1.5	1.26-1.28	0.3 to 0.6
1.0	1.26-1.28	0.0
1.0	1.30-1.34	0.4 to 0.6
0.0	1.30-1.34	0

But in the writer's opinion, the main point which has been overlooked by practically all workers is the distribution and fineness of lead. It is believed that if due care is given to this point, there would be little difficulty in incorporating up to 1.5 per cent lead in many alloys. For example, 1.3 per cent of lead was added to the essentially beta alloy H.S.B. 44 and mechanical properties were as follows:

	Ultimate Strength, psi.	*Elongation on 2 in., Per Cent*
(a)	102,500	20
(b)	102,500	20
Required	98,500	13

Then to illustrate the influence of distribution and particle size the following test bars whose mechanical properties are shown below were in some cases cut from adjacent small bars and in all cases from the same pot of metal. The alloy without lead additions had a yield point of 42,700 psi., ultimate tensile strength of 92,600 psi. and elongation 22 per cent. With an addition of 0.8 per cent lead, results were as follows:

Yield Point, psi.	*Ultimate Strength, psi.*	*Elongation on 2 in., Per Cent*
n.e.	69,500	10
37,600	77,000	21.5
33,000	79,000	26.0
31,800	65,000	8.7

It would appear that the greatest lowering in mechanical properties

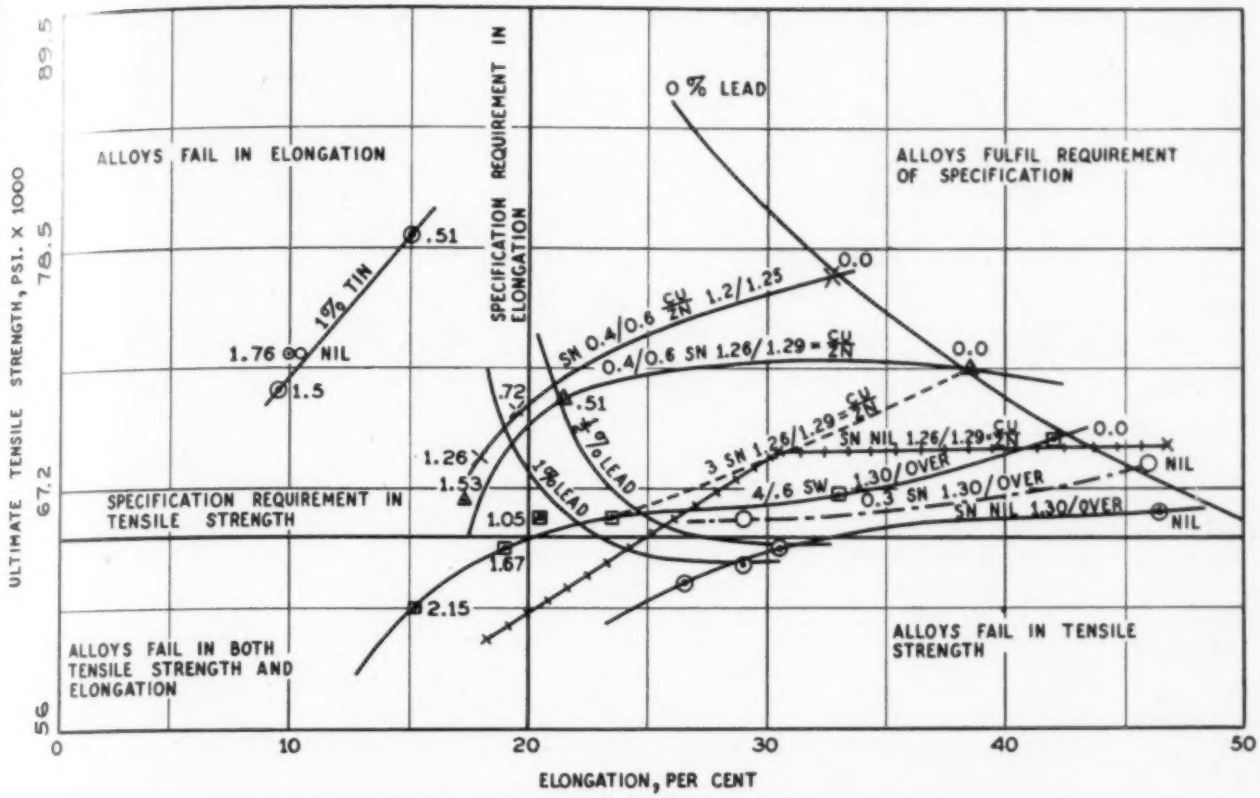


Fig. 3—Curves designed to show effect of tin and lead on tensile properties of manganese bronze.

occurs with the addition of the first half per cent of lead, this is shown in Fig. 3 which is a replot of some published results. In assessing the value of lead in manganese bronze, its beneficial effect on corrosion resistance in many cases should not be overlooked.

High Aluminum Alloys e.g. H.S.B. 44. As aluminum is the element which gives rise to great increases in strength, so accordingly it is found that the high strength manganese bronzes having tensile strength of 98,500 psi. and over are high in aluminum. For example, with appropriate composition, Duma uses an alloy containing 4.5 per cent of 5.7 per cent aluminum to produce an alloy having an ultimate tensile strength of 100,000 psi. maximum.

The following may be taken as a typical composition of H.S.B 44 alloy.

Element	*Per Cent*
Copper	62.25
Aluminum	4.5
Iron	1.50
Manganese	2.25
Zinc	Balance

The zinc equivalent of the alloy using Guillet's equivalents (but realizing their limitations) is 49.6 per cent. Using three-component representation, the alloy reduces to the following:

Aluminum	4.5 per cent
Equivalent Zinc	32.4 per cent
Equivalent Copper	63.1 per cent

From the zinc equivalent of 49.7 per cent reference to the copper-zinc phase boundaries due to Deven and Pickup shows the alloy consists of beta with some gamma which means it would be brittle. Using an older diagram which places the boundary at 50 per cent is a little more comforting, but not very helpful.

Considering the alloy as a three component, one is more informative for reference to isothermal sections of the copper-zinc-aluminum systems. It shows the narrow range of composition over which solely beta alloys are obtained. As aluminum content is increased, it is necessary to increase copper content to avoid the occurence of gamma which renders the alloy brittle.

The avoidance of brittleness, by which is meant poor elongation, is perhaps the problem which is encountered in this alloy rather than in BSS 208, Class 1 which is an alloy of much lower tensile strength. Instead of having yield point 53,800 psi., tensile strength 97,500 psi. and elongation 12 per cent, the properties become nil, 44,200 psi. and nil respectively when appreciable gamma enters the structure.

In practice this may result from too high a zinc equivalent, generally brought about by adding a percentage of zinc to balance zinc loss in melting and the zinc loss planned for, not taking place. This may appear surprising, particularly if the same procedure has been applied for duplex alloys of the BSS 208, Class 1 type.

But it must be borne in mind that the zinc loss is not as great when melting beta alloys. This is so because firstly, on account of the film of aluminum oxide, resulting from the higher aluminum content inhibiting the vaporization of zinc, and secondly there is not as much zinc present in the alloy, 20 per cent as against 10 per cent in the lower strength alloys.

Returning to the study of the phase diagram for copper-zinc-aluminum alloys, the following observations relative to high strength manganese bronzes may be made.

1. The phase boundaries on the diagram Fig. 4 are for equilibrium conditions or very slow cooling con-

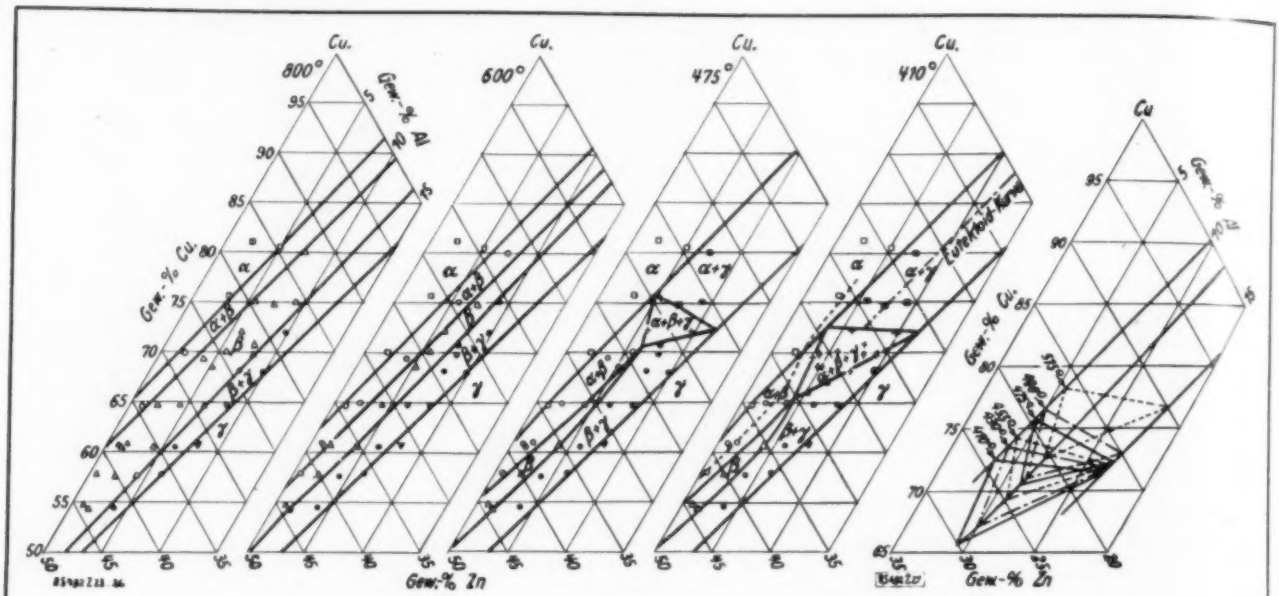


Fig. 4—Isothermal sections of copper-zinc-al uminum system. (After Baner and Hansen.)

ditions whereas, with castings the cooling rate is much greater. This has a practical consequence for the three-phase alpha-beta-gamma area which persists at temperatures up to 515° C. has disappeared at 600° C. So, if the cooling rate of the casting is sufficient to retain the structure of 600° C. then alloys which would normally contain gamma would be free of it in the cast state.

2. The maximum amount of aluminum at which could be used in the alloy without entering the duplex beta-gamma field with increase of aluminum or the three-phase alpha-beta-gamma field with increase of copper is approximately 6.5 per cent—with reference to the isothermal section at 475° C. which as an approximation could be considered to represent the cast alloys.

3. Copper may not be increased without limit without encountering a field containing gamma. For equilibrium conditions at 410° C., this appears to be 65 per cent and increases to 73 per cent at 515° C. For castings the limiting amount would be about 70 per cent copper.

4. Reference to certain sections of constant zinc content Fig. 5 show at low temperatures, there is a field either three phase, or two phase containing gamma. Attention is drawn to these in the sections of constant zinc content.

The implication of them is that prolonged heating at low temperatures may render the alloy brittle.

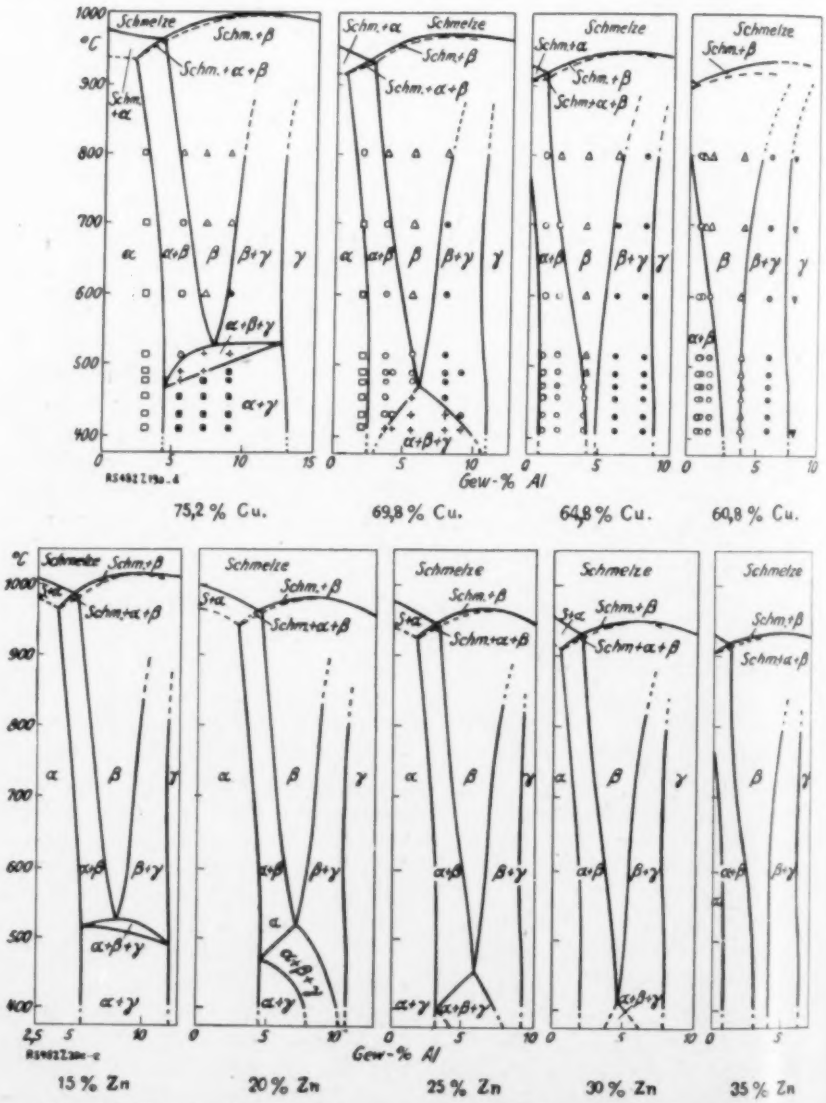


Fig. 5—Constant zinc sections, copper of zinc-aluminum system. (After Baner and Hansen.)

Below are the results of heat treatment of bars of H.S.B. 44 which illustrate this point. All bars were cast from the same pot of metal.

Condition	Yield Point, psi.	Tensile Strength, psi.	Elong. on 2 in., per cent
As Cast	66,600	108,000	21.0
1¼ hr. at 700° C. water quench	58,300	100,000	10.0
As above, plus 4 hr. at 425° C.	52,000	56,000	0.5
As above, plus 4 hr. at 525° C.	47,200	58,700	0.5

The cooling rate of the as cast bar gives rise to the best mechanical properties and the brittleness of the quenched and tempered bars is noticeable. The foregoing is not a complete explanation of the phase changes which may occur in these complex alloys but indicate that trends and tentative conclusions which have some practical implication may be drawn.

They are, that the cooling rate normally associated with casting may be sufficiently rapid to suppress undesirable phase changes which may give rise to loss of ductility. It is advisable not to evaluate metal quality on the results of a chill cast test bar if it is to be used in sand castings.

This may sound redundant but it is frequently the practice to do so and sufficient has been said to indicate that what may be regarded as beta alloys using a two phase representation may really turn out to be alloys which undergo undesirable changes in phase, at relatively low temperature.

Tin-Bearing Bronze

Tin-Bearing Manganese Bronze, e.g. BS/STA.7/C.X.3. The first point to be considered is the necessity for tin in these alloys. It may appear redundant to discuss this, but it is thought necessary because some writers quite recently considered tin in manganese bronze unnecessary and undesirable. It is undesirable because of the brittleness which results from excessive tin, and which may result when a suitable amount of tin is added to an unsuitable composition.

But tin has a definite function; it is to confer increased corrosion resistance, and while this is doubted by some, there is no doubt in the writer's mind concerning its value in this regard. It is not appropriate to consider in great detail the "tin vs. tin-free manganese bronzes" for a large number of corrosive media but some remarks will be made concerning the corrosion of manganese bronze in sea water.

Arising out of a discussion of copper sheathed vessels with manganese bronze propellors, some experiments were made by taking copper sheet and specimens of manganese bronze whose superficial areas were approximately in the ratio of the areas of these materials when used on the wooden boats.

Corrosion Resistance

The manganese bronze was placed on the copper and the lot immersed in sea water which was agitated for 4 hr. daily. Three grades of manganese were used.

1. Beta alloys—H.S.B. 44.
2. Duplex alloy alpha and beta—BSS 208 Class 1.
3. C.X.3.—which contains 0.7/1.4 per cent tin

while the other two specimens were tin free.

The behavior of the three specimens was as follows: The tin-free alloys rapidly dezincified, a white non-adherent "growth" formed around them and beneath which was the usual copper color dezincified surface. The C.X.3 did not behave in this manner, it did not dezincify, but the beta areas were corroded in preference to the alpha.

Loss in weight of the specimens under these conditions was as follows:

Alloy	Loss in weight Milligrams/ sq.cm./24 hr.
C.X.3.	0.134
BSS 208 Class 1 (no tin)	2.15
HSB 44 (no tin)	2.74

The time of immersion was of the order 300 hr. and the superior behavior of the tin-bearing alloy is evident. In order to give some idea of the expected performance, Hudson considers that if the loss in weight does not exceed 0.7 milligrams/sq. cm. per 24 hr., satisfactory performance may be expected.

Dezincification

Hesse, Myskowski and Loring found that tin was the most effective element in inhibiting dezincification. They used an accelerated test involving hydrochloric acid as a corrosive media and determined the ultimate tensile strength at intervals. A summary of their results is given below in psi.

Alloy Containing	Original	After 1 week	After 3 weeks
No tin	68,500	18,100	0
½% tin	69,700	50,900	29,000
½% tin, ⅓% lead	68,900	58,200	38,000
0.5% lead	62,300	38,900	100
0.1% phosphorus	61,000	26,500	8,950
0.24% arsenic	64,100	37,800	7,900
0.15% antimony	61,300	13,600	0

Concerning galvanic action, practically all lists of potentials (relative to calomel) of different metals in sea water do not distinguish between tin-free and tin-bearing manganese bronze and frequently do not show composition. The potentials in sea water of the three grades of manganese bronzes relative to saturated calomel half cell are:

Alloy	Volts
C.X.3.	−0.29
H.S.B. 44	−0.35
BSS 208 Class 1	−0.33

The above serves to show the relative values but a lot more detail is necessary to fit them into some of the more comprehensive published lists. Without going deeply into the matter, it may be said that such matters as polarization and relative areas must be considered when applying these tables of potentials.

Examination of several of the published ones, show the order of some metals is slightly different from others but one repeats that detailed qualifications of conditions is necessary. For those interested in this matter, the work of LaQue and Cox[23] is very comprehensive.

It is, perhaps, appropriate to close these notes on corrosion resistance with a reference to the use of copper sheathing in conjunction with manganese bronze propellors. Published opinion does not recommend it.

In discussion of Parsons and Cooks' paper in *Transactions* of Institute of Naval Architects (1919), Sir John Thornycroft drew attention to the fact that corrosion of the propellor may occur in a copper bottomed boat. Murphy[3] and Hensel[24] have expressed similar views.

The writer understands, however, that some degree of satisfactory service has resulted from copper sheathing in conjunction with tin-bearing

manganese bronze, and in fact the accelerated corrosion test tends to confirm such to be the case but one would not view this practice with great enthusiasm because of the many variables encountered in ship corrosion. A few of these are, temperature of the water, nature of the water, e.g. fresh water, water near decaying vegetable matter, nature of the port in which the vessel is plying or is tied up.

Frequently when one investigates the composition of alloys quoted by work with the object of demonstrating the ill effects of tin on the mechanical properties, it is seen that tin is incorporated in unsuitable compositions, for example the aluminum is unduly high. Following is such an example:

Element in per cent or Property	*Alloy*	
	Tin-bearing	*Tin-free*
Copper	62.25	62.5
Tin	1.0	Nil
Aluminum	3.0	3.0
Iron	2.0	2.0
Manganese	2.5	2.5
Zinc	29	30
Yield Point, psi	51,500	49,300
Ultimate Strength, psi	98,500	94,000
Elongation, per cent	14	25

At first sight it would appear that the above is evidence of the ill effects of tin on the mechanical properties of manganese bronze, but attention is drawn to the high aluminum contents.

Alloy C.X.3 Properties

Dealing now with the constitution and properties of C.X.3 its specified composition is as follows:

Element	*Per cent*
Copper	55 to 57
Tin	0.7/1.4
Lead	0.2 max.
Iron	0.5 to 1.5
Manganese	0.05 to 0.5
Nickel	0.2 max.
Aluminum	0.05 to 0.5
Other elements	0.1
Zinc	Balance
Equivalent zinc	46 max.

The limitation of zinc equivalent implies a further restriction on choice of composition for the specified range of composition has a minimum zinc equivalent of 43 per cent and a maximum equivalent of 47 per cent. Yet the specified maximum zinc equivalent of 46 per cent is not without practical significance as regards mechanical properties as the following examples of test bars show.

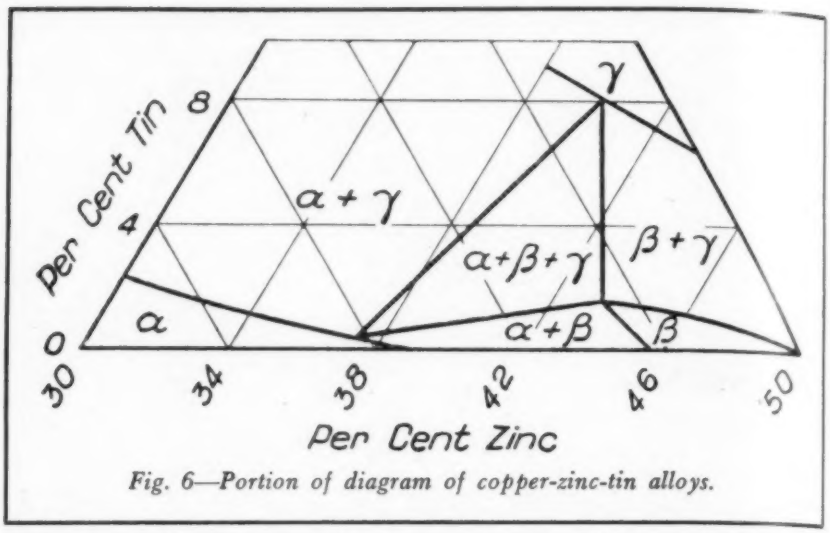


Fig. 6—Portion of diagram of copper-zinc-tin alloys.

Zinc Equivalent, per cent	*Yield Point, psi.*	*Ultimate Strength, psi.*	*Elongation, per cent*
46.23	47,400	79,800	11.0
46.30	48,400	75,200	7.0
44.94	46,600	83,000	23.5
44.9	38,300	78,500	34.5

The above may be criticized on the grounds that limits are too small but it is emphased throughout that there are only narrow ranges when working with C.X.3.

Consider now the three-component representation of this alloy. Figure 6 is an enlarged portion of the copper-tin system at a temperature of 410° C. Then the limits of zinc equivalent are from 41.2 to 46 per cent and one may draw a parallelogram on this copper-zinc diagram, the sides of which represent approximately the limits of C.X.3 as regards tin content and zinc equivalent of the remainder of the alloy.

For clarity, the data shown by the parallelogram has been replotted on rectangular co-ordinates as shown in Fig. 7 and the boundaries between the regions of composition in which different microconstituents or phases occur are shown.

It will be seen there are four regions from each of which alloys of differing microstructure and mechanical properties may be obtained. It is desired to produce alloys whose compositions lie in the field wherein the structure is duplex alpha and beta, so the choice of composition is still more restricted.

Let us make some observations from the phase diagrams which have some bearing on practical problem.

1. Gamma may be encountered in the microstructure, yet the alloy will give the desired mechanical properties. This may sound odd because in the beta alloys, the gamma present renders them brittle and of poor elongation. The reason appears to be that in these alloys gamma is associated with alpha as well as beta and the alpha is present in considerable amounts. See section of 1 per cent tin alloys in Fig. 8.

Of course, if the gamma is associated only with beta then brittleness will result. At this stage, it is interesting to recall that brasses are rendered brittle by a trace of gamma, but bronzes can tolerate an appreciable amount of delta. Now, as the delta constituent of copper-tin alloys has a crystal structure similar to the gamma phase in copper-tin alloys it may be referred to as gamma for uniformity of nomenclature.

It would appear that the difference in behavior is one of kind and amount of associated phase rather than kind of alloy. Some writers re-

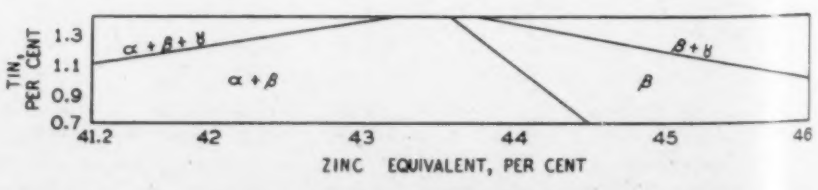


Fig. 7—Composition range of alloy C.X.3. Figure shows possible structure of alloys conforming to requirement of chemical components.

fer to the above mentioned gamma as $Cu_{31}Sn_8$ or more correctly Cu_3Sn_8.

2. Great care must be exercised in melting for uncontrolled and excessive zinc loss can give rise to two effects of opposite kind. First, lowering of the zinc content will lower the yield point, ultimate strength and increase the elongation. Secondly, in high-tin alloys the effect of lowering the zinc content is to change the structure from duplex alpha beta, to three phase alpha, beta and gamma and if enough gamma the alloy will have low elongation.

This is rather confusing, as is generally an unexpected effect of loss of zinc. One of the most practical measures of control is to melt one or more pots of metal under different conditions of temperature, and also hold it for varying times, pouring test bars at intervals, the test bars are pulled and the metal analyzed and a curve of variation of properties and composition against time is plotted simple to the example shown in Fig. 9. The object being, of course, to melt under constant conditions.

3. Viewing the isothermal section again, one sees that to maintain an essentially alpha-beta alloy, the minimum zinc content does not correspond to the maximum tin content and that a maximum zinc equivalent of 43.5 per cent permits the maximum of tin to retain an alpha-beta structure. Beyond this point relatively small increases of zinc necessitate a rapid reduction in zinc content to avoid the formation of an entirely beta structure.

Therefore, one cannot add ½ to 1½ per cent of zinc to make up zinc

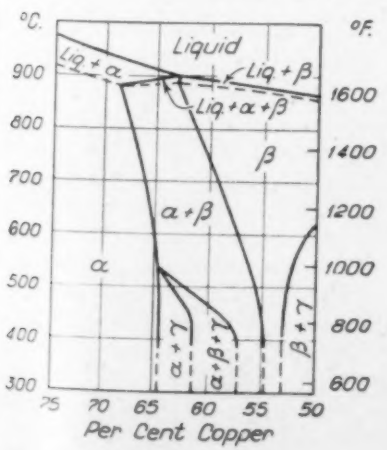


Fig. 8—Vertical section of copper-zinc-tin alloys containing 1% tin.

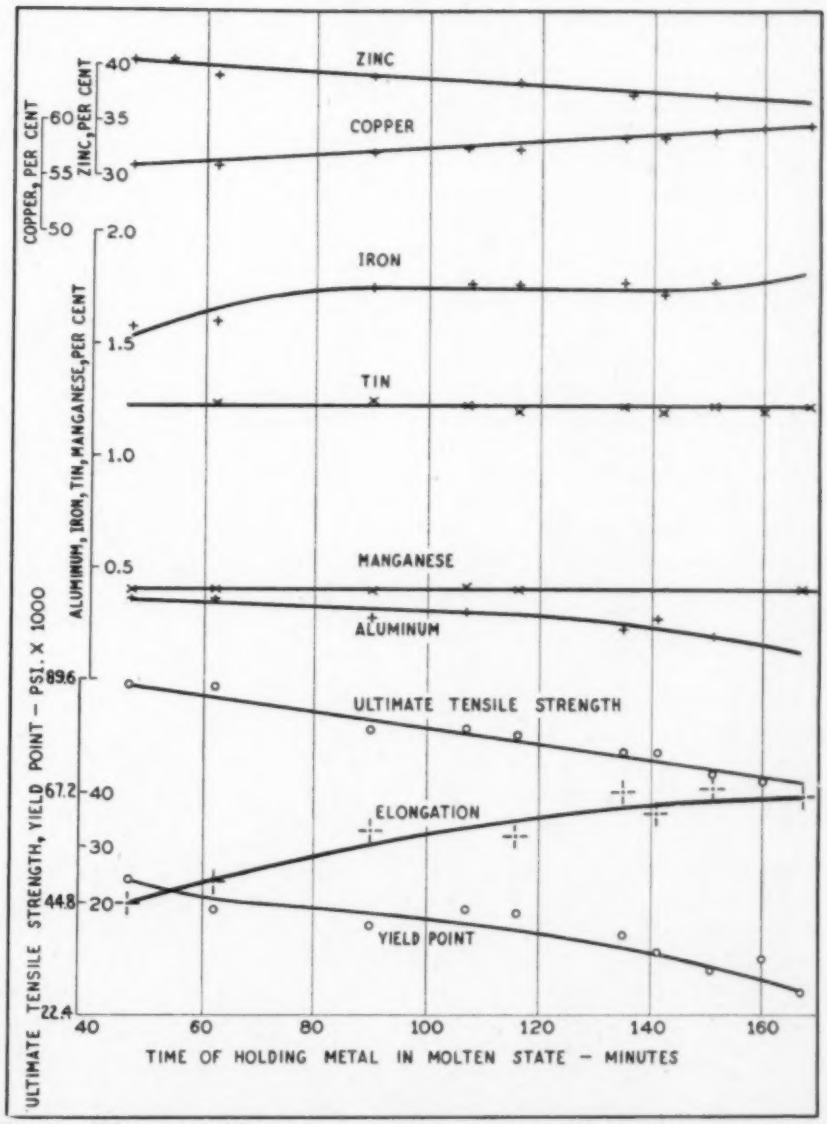


Fig. 9—Effect of holding time on physical properties and composition of manganese bronze (C.X.3.). Size of melt, 180 lb.

losses with the same delightful abandon as in BSS 208, Class 1 because the tin content has important bearing on the matter.

4. Consider the three-phase volume alpha, beta and gamma. Figure 10 is an attempt to show it in three dimensional form. The first point is that as temperature rises the boundary passes to higher tin contents or it recedes from the binary copper-zinc side of the diagram.

This is of assistance in the production of castings because the cooling rate being faster than equilibrium rates preserves a structure which is reasonably stable and free from gamma. In fact the faster the equilibrium cooling gets founders out of many difficulties with compositions which, while within specified limits are not within the most desirable range of composition.

Following are the results of heat treatment on C.X.3 bars made from adjacent pieces of suitably headed stick.

No.	Heat Treatment	Ultimate Strength, psi.	Elongation on 1½ in., per cent
1.	As cast	89,300	16.65
2.	1 hr., 700° C., F.C.*	91,000	22.0
3.	1 hr., 700° C., W.Q.	83,000	9.33
4.	1 hr., 700° C. then A.C. to 500° C. Hold 1 hr.	81,400	9.33
5.	1 hr., 700° C. then A.C. to 400° C. Hold 1 hr.	90,800	21.3
6.	1 hr., 700° C. then A.C. to 400° C. Hold 4 hr.	88,700	18.7
7.	1 hr., 700° C. W.Q., then 350° C. for 1 hr.	52,000	2.0

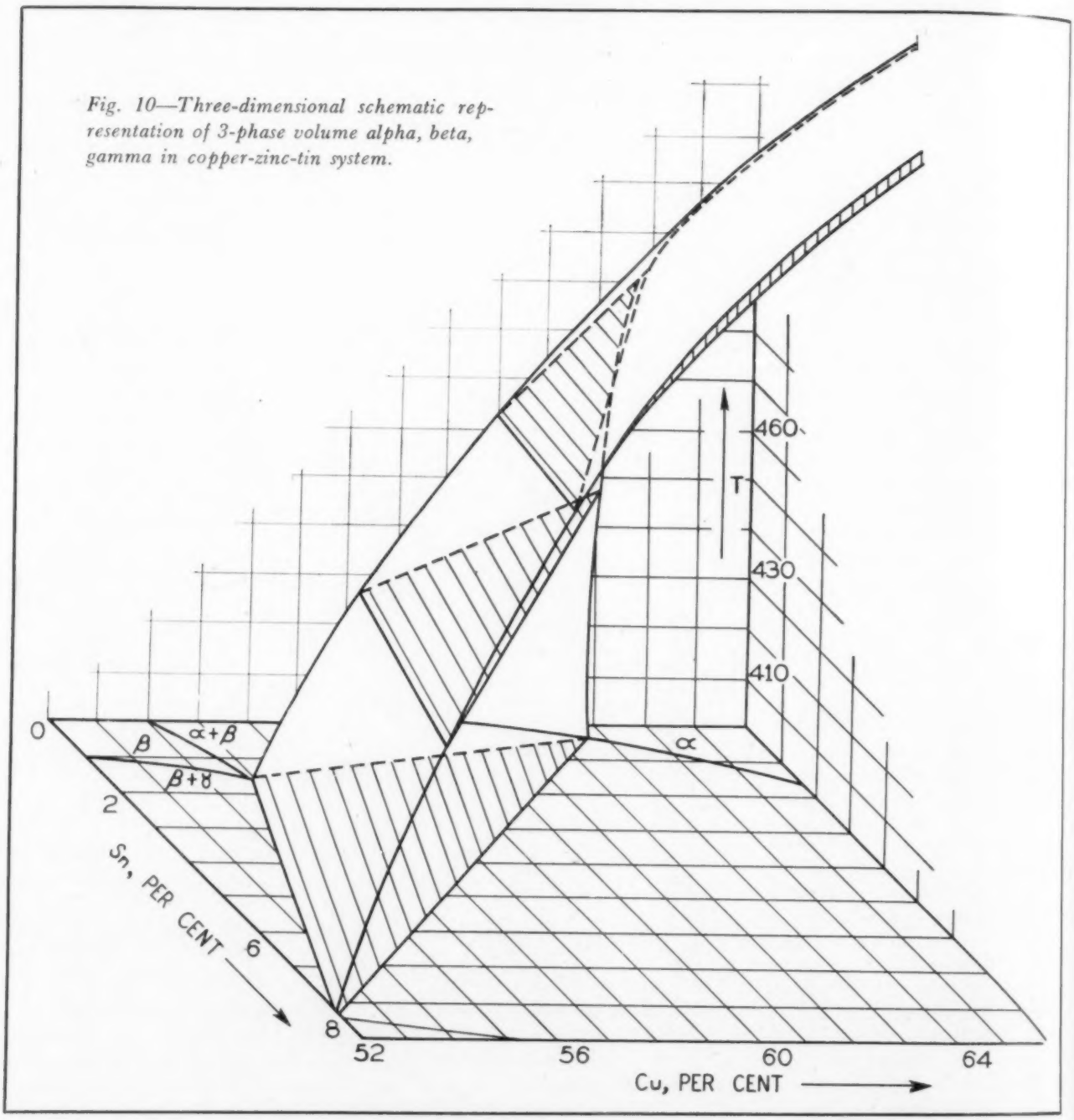


Fig. 10—Three-dimensional schematic representation of 3-phase volume alpha, beta, gamma in copper-zinc-tin system.

8.	1 hr., 700° C. W.Q., then 350° C. for 4 hr.	44,800 2.0
9.	4 hr. at 350° C.	86,000 14.7
10.	8 hr. at 350° C.	13.3

*F.C.=furnace cooled; W.Q.=water quenched; A.C.=air cooled.

Water quenching lowers ductility appreciably, but water quenching and reheating lowers strength and elongation very appreciably, elongation drops to 2 per cent.

Examination of the microstructure shows that grain boundary precipitation occurs which is probably an intergrowth of alpha and gamma. The grain boundary phenomena was not perceptible in those specimens which had been held at high temperatures then cooled to 400° C. and it will be seen that the elongation was not greatly impaired.

This brittleness may, of course, have important practical consequences particularly if a casting is rapidly cooled, e.g. a small one and then reheated to a relatively low temperature. The effect of reheating an as-cast structure is shown by comparing the result of tests 9 and 10 in the preceding table with those of test 1 in the same table.

It will be noticed that heating at the relatively low temperature of 350° C. has resulted in some decrease in elongation but not a very pronounced one as in the case of the water quenched and reheated specimens. Explanation for this behavior is sought in the phase relationships in the copper-zinc-tin alloys, attention being directed to the boundary fields containing gamma and alpha and beta. With increase of temperature, this passes to higher tin contents—for constant copper content.

Should the composition of a complex alloy be such that when the equilibrium at room temperatures its structure contains gamma, but as a result of rapid cooling a structure

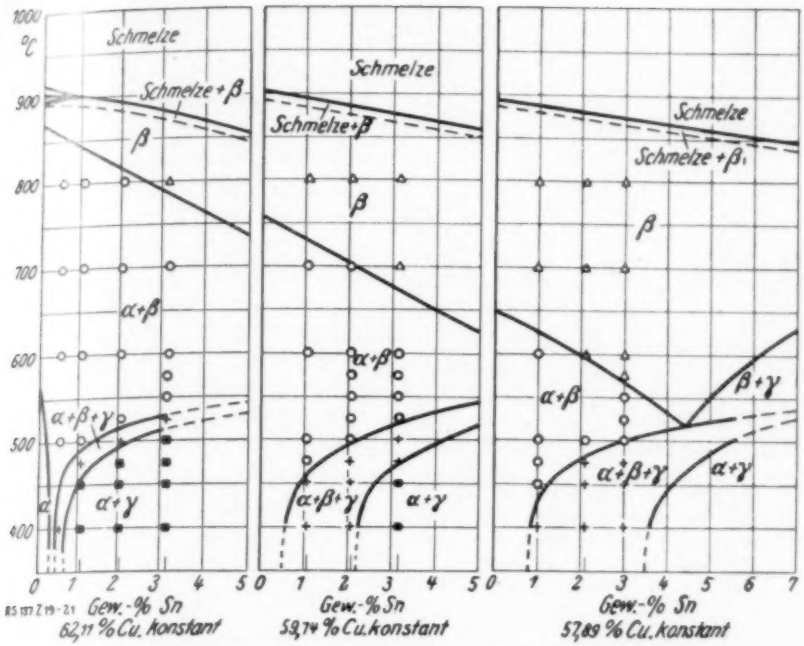


Fig. 11—Constant copper sections of copper-zinc-tin system.

typical of equilibrium at high temperatures is retained, their heating will tend towards an equilibrium and a varying amount of gamma may be expected in the structure depending on whether or not equilibrium is reached.

Study of sections of the copper-tin-zinc system in Fig. 11 and 12 shows the trend of the phase relationships with increasing copper content and an interesting and significant point arises in that increase of copper results in appear of the three-phase gamma containing field at lower tin contents. Probably practical consequences are as follows:

1. Should the equilibrium structure of the alloy show it to contain gamma, then reheating a casting, e.g. to stress relieve it at temperatures wherein the equilibrium structure shows it to contain gamma may result in a decrease in elongation. Such temperatures are of the order 350° to 400° C. Actually this has been found to occur.

2. While 400° C. is generally considered to be the minimum stress relieving temperature, probably 500° C. would be safer—thus avoiding the three-phase field with all but copper contents over 61 per cent.

3. On the other hand, control of composition so that the equilibrium constitution does not contain gamma allows one to reheat at temperatures of, e.g. 350° C. without any reduction in ductility.

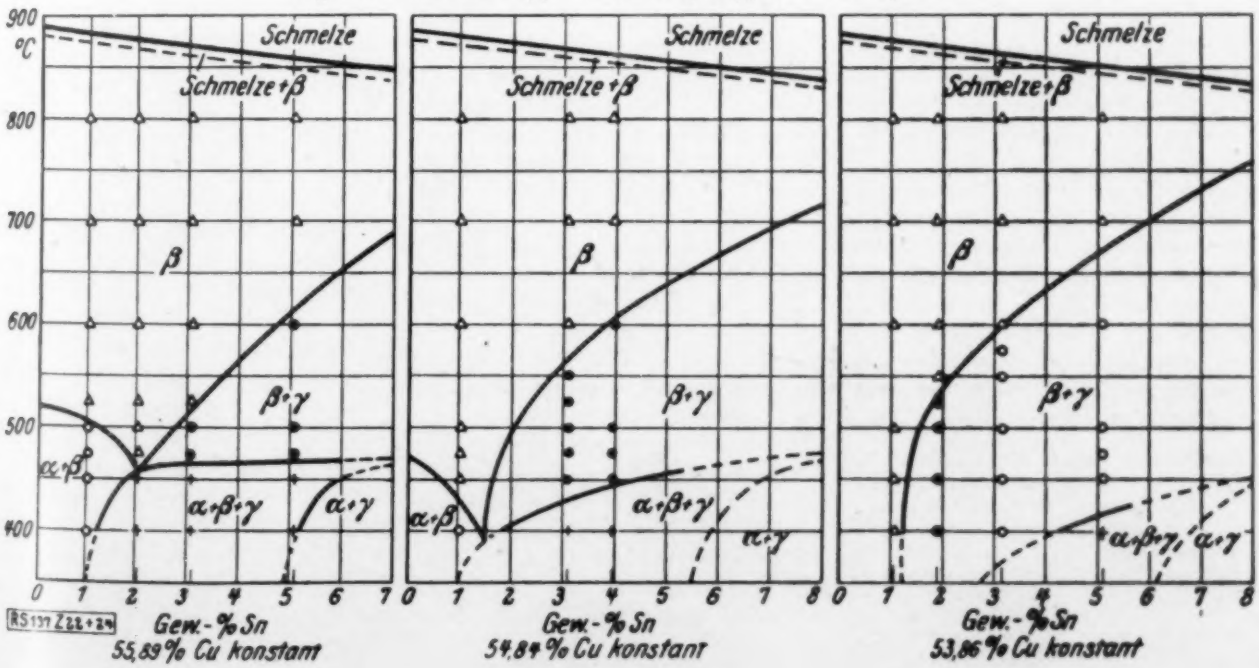
It is desired to emphasize that relatively slight differences in composition will differentiate between alloys which behave as in (1) and as in (4). One of the reasons for the difference in mechanical properties between chill cast and sand cast test bars of C.X.3 is found in the foregoing. In the case of C.X.3 chill cast test bars frequently have more desirable mechanical properties than sand cast test bars as is shown by the following examples of test bars cast from the same heat of metal.

	—Ultimate Strength— psi.		—Elongation— per cent	
	Chill Cast	Sand Cast	Chill Cast	Sand Cast
(a)	81,500	71,600	68,300	15,700
(b)	90,500	83,000	54,900	33,600
(c)	80,600	75,700	78,500	33,600

It is therefore unwise to subscribe to the somewhat careless habit of evaluating manganese bronze on test bars without inquiring to the manner in which they were cast, whether they be C.X.3., or duplex alpha alloys or beta alloys.

Finally, to conclude the section

Fig. 12—Constant copper sections of copper-zinc-tin system.



dealing with C.X.3. one has been able to effect improvement in the mechanical properties by furnace cooling from temperatures of the order 700° C. as the following results indicate.

	Yield Point, psi.	*Tensile Strength, psi.*	*Elongation, per cent*
As Cast	45,100	81,500	21
Furnace cooled 700° C.	43,400	80,000	27

General Physical Properties. These are in the nature of a supplement to the information on yield point, ultimate strength and elongation and are intended for reference purposes.

Young's Modulus of Elasticity.

Alpha beta .....12,200,000 psi.
Beta .....15,800,000 psi.

Compression and Torsion. For an alloy having an ultimate tensile strength of 70,000 psi. and 32 per cent elongation on 2 in. strength data is as follows:

Modulus of Elasticity.....14,235,000 psi.
Compression:
Proportional limit..... 15,500 psi.
Compression Strength.. 42,390 psi.
Torsion:
Modulus of Rupture.... 61,790 psi.
Modulus of Rigidity.... 4,619,000 psi.
Shear Strength..... 43,190 psi.
Endurance Limit ..... 15,000 psi.
(Specimens broke 20 million)
Poisson's Ratio 0.306
Brinell Hardness:
Alpha beta 110-140
Beta 180-220

High Temperature Properties

Properties at Elevated Temperatures. The properties of both cast and drawn manganese bronze change rapidly with increasing temperature. The tensile strength decreases rapidly with temperature being at 60 per cent of its original value at 300° C.; elongation increases rapidly. It has been found that elongation reaches a maximum at about 300-350° C. As its strength is not very great at elevated temperatures, due care must be given to packing propeller blades in heat treatment, to prevent warping.

Properties at Subzero Temperatures.

Temperature °C.	*Ultimate Strength, psi.*	*Elongation on 2 in., per cent*
Room Temp.	72,500	44
−10	70,500	45
−80	75,500	43
−120	81,000	45
−180	94,800	—

Thermal Conductivity.

Mean Temp., °C.	*Calories/sq.cm./sec./°C./cm. thickness*
81.5	0.171
120.5	0.175
181.5	0.181
230.5	0.184
266	0.195
311.5	0.200
392	—

Electrical Conductivity.

Alpha beta .....24% of copper standard
Beta .....13% of copper standard

Coefficient of Thermal Expansion.

	Temp. Range, 70-200° F.	*Temp. Range, 70-500° F.*
Alpha beta	0.00001077	.00001177
Beta	0.0000099	.00001160

Coefficient of Friction. Following are values of coefficient of friction of alloys against mild steel using static pressure of 10 to 120 psi. Lubricant was rape oil.

Gunmetal 0.155
Whitemetal 0.109
Manganese Bronze 0.124
Aluminum 0.111
Axle Steel 0.137

Applications of Manganese Bronze. Rather than present a list of known applications, it is proposed to discuss the metallurgical features governing the selection of the alloy for a given application.

Resistance to Cavitation. Cavitational damage of metals is caused by the impact which results when bubbles in the fluid, e.g. sea water collapse upon the surface of the metal. Such damage is found in propellers, pump impellers and similar metallic parts.

Of recent years, there has been considerable research into the cavitation resistance of metals and alloys with the result that manganese bronze has been found to have high resistance. Those qualities which confer resistance to cavitation upon a metal are high strength and hardness, ability to take a high polish and retain it, good corrosion resistance, fine grain size and freedom from dross and inclusions.

Thus manganese bronze is used for pump runners, turbine parts, and ship propellers. In connection with the latter, attention is drawn to the remarks in the historical note on the length of time manganese propellers have been used, and their magnitude is indicated by the cast weight of the propellers used from the "Queen Mary," the propellers weighed 55 tons.

Corrosion Resistance. By virtue of its good corrosion resistance the tin-bearing manganese bronzes are used extensively for underwater fittings of ships. It finds application in pickling equipment.

Strength. The high strength of the alloys make it useful for many applications, e.g. gun mountings, spin and gear wheels. In regard to propellers, the high strength allows thinner blades to be made which has a net result of increasing the efficiency by about 7 per cent. It also finds application for making non-sparking tools for oil and explosive industries.

Non-magnetic Quality. As the alloys are essentially non-magnetic and strong, they are used in the construction of electrical equipment.

Substitute for Forgings. Owing to their high strength and aquiaxed grain structure which does not give rise to planes of weakness in castings in manganese bronze are frequently used to replace forgings.

Substitute for Malleable Iron. Dornblatt and Woldmann in *Engineering Alloys* list alloy No. 711 and describe it as a substitute for malleable iron. Locally, of course, the millions of boot hobs which have been produced may be regarded as a substitute for malleable iron, although, of course, the manganese bronze is a better product. These hobs were used on boots for jungle wear.

Ferrous materials were reported to be unsatisfactory owing to corrosion effects, zinc-base die castings wore out in a matter of hours, so did brass, but manganese bronze (minimum hardness 170 V.P.N.) had a much longer life. Some millions were supplied for this rather interesting application of manganese bronze.

Antifriction Properties. Manganese bronze finds application for bearing where heavy loads are encountered, e.g. trunnion bearings. In this regard, it is well to employ compositions which are of the higher copper, higher aluminum variety, as certain compositions are unsatisfactory in their behavior.

Specific Tenacity. This empirical factor relates the strength of the material to its specific gravity and is useful for forming an estimate of the

weight of a component in relation to its strength.

Manganese bronze has about the highest specific tenacity of copper-base alloys as the table shows, and this no doubt influences its selection for many applications where weight as well as strength is an important consideration.

	Specific Tenacity
Manganese Bronze (alpha beta)	10,976
(beta)	13,888
Aluminum bronze	13,440
Monel	7,616
Y alloy	15,904
Electron AZ8	19,936
Gunmetal	4,600

Summary and Conclusion. The foregoing has dealt with the cast alloys only, but the wrought manganese bronzes constitute a very important group of alloys.

It will have been inferred from the remarks made, that these complex alloys involve complex problems of constitution and methods of approach have been indicated, while when dealing with specific types the importance of phase relationships has been stressed.

From the engineering point of view, the alloys are very important in that their high strength and corrosion resistance allows them to be widely used, and there appears to be very good reason to adopt their use more widely in engineering practice.

Acknowledgments

First of all the writer wishes to make general acknowledgement to the authors of papers which he has read from time to time and has possibly inadvertently omitted to detail some of them in the references.

To Mr. Russell N. Stokes, Managing Director of Stokes & Sons Pty. Ltd. Brunswick and to Mr. C. Homer Fraser, Joint Managing Director of Stokes Foundries Pty. Ltd. Brunswick, Victoria, acknowledgement is made for placing facilities at the disposal of the writer to carry out the work embodied in, and to prepare the paper and finally to colleagues for assisting with preparation of drawings.

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DISCUSSION

Chairman: G. P. HALLIWELL, H. Kramer & Co., Chicago.

Co-Chairman: A. H. HESSE, R. Lavin & Sons, Inc., Chicago.

H. J. ROAST[1]: This is rather an exceptional paper. I want to congratulate Mr. Halliwell for the very masterly manner in which he summarized it. The general trend is that of the increasing importance of physical metallurgy, bringing in, as it does, advanced mathematics and the new atomic theories, as we have recently had more reason to look into them, since the atomic age.

The practical foundryman may not get a great deal from this paper, but the theoretical metallurgist should pay a great deal of attention to it. I feel we should record very definite appreciation of this paper.

MR. HALLIWELL: In reading over this paper I was impressed with the amount of tin permitted in Australian specifications. I am not familiar with the British or Australian specifications, but from the amount of aluminum herein indicated to be present, I would expect a high strength alloy. It has been our experience that with the exception of the 60-70,000-psi. alloy, tin reduces the elongation in the high-strength (beta) alloys.

[1] Canadian Bronze, Montreal, Quebec, Canada.

PRODUCING BOMB CASTINGS by Use of the CORE BARREL

Basil Gray

English Steel Corp. Ltd.
Sheffield, England

DURING THE WAR the output of the steel foundries of Great Britain was more than doubled by the additional output from new, government sponsored factories managed by existing firms, and by converted iron foundries. These projects, the castings produced as a result of them and the methods employed could well supply appropriate subjects not merely for one but for several exchange papers. Some of the mass-production foundries designed to produce bomb cases and tank track links, for instance, were so fully mechanized that castings up to 500 lb. in weight were produced with a labor force comprising, in some cases, 80 per cent women.

Good Cooperation

In the many developments in tank armor castings the leading producers co-operated to a degree previously unknown, with excellent results, as shown by the improvement in quality and technique that followed. The co-operation, started in this way, has engendered a new spirit in the steel castings industry of Great Britain, and today the British Steel Foundry Technical Association, which in itself would form the subject of an interesting talk, is a live organization in which members of some 80 per cent of the foundries take an active part with the fullest interchange of ideas and information.

Production of the "grand slam" and "tallboy" bombs was chosen as a subject, not because it supplied front-page news in the public press a year ago, but because the method to be described included a revised version of an old and little-known technique which, at least, in a limited field, has proved of great value.

Bomb Production

Production of these bombs (Fig. 1) has been illustrated in American technical journals with a creditable description, largely deduced from photographs and brief technical data, but the reasons for adopting such an unusual method for constructing the core must, nevertheless, have been rather puzzling to many readers, and a detailed description may be of interest.

Molding with a Strickle. The striking up of a mold with a strickle (sweep) is almost as old as the art of founding itself, but the striking up of cores on a core barrel, although also an old practice, is less well known. It is used in Great Britain, to some extent, by a few heavy steel foundries as well as by iron foundries for jobbing work, but not often for mass production work.

Twenty-fifth in an unbroken series of annual exchange papers—1922-1946—from the Institute of British Foundrymen to the American Foundrymen's Association . . . the present paper is an example of the value of this broad exchange of knowledge and practical developments for the common advancement of the foundry industry and international understanding.

From Bells to Bombs

Nearly 100 years ago cast steel church bells were made in some quantities by the author's firm, using a similar process. The mold was formed on a plate with the strickle (sweep) post fixed to it, and a bell-shaped cast iron piece, with many perforations to allow the escape of gases and moisture, was threaded over it. Loam was applied, struck up to shape with a strickle board and dried. A clay mixture came next, which was strickled to the shape of the outside of the bell and also dried. The cope was then rammed up by ordinary methods, using the clay as the pattern or "thickness," as it used to be called. When the cope had been lifted the clay "thickness" was broken away and the mold was completed.

Earlier samples of the 6-ton "tallboy" were made by jobbing methods, using a barrel core. They had a number of successes, such as the penetration of the "U" boat pens at L'Orient and the collapse of the entrances to an underground garage for the V-1 rocket planes.

Foundry Conditions. The demand for maximum output which followed came at a time when all mechanized foundry and the heavy molding machines were fully occupied on tank castings and similar important work, but when the heavy molding pits and their crews were comparatively slack owing to the cessation of the

Presented at a Steel Session of the Fiftieth Annual Meeting, American Foundrymen's Association, at Cleveland, May 10, 1946.

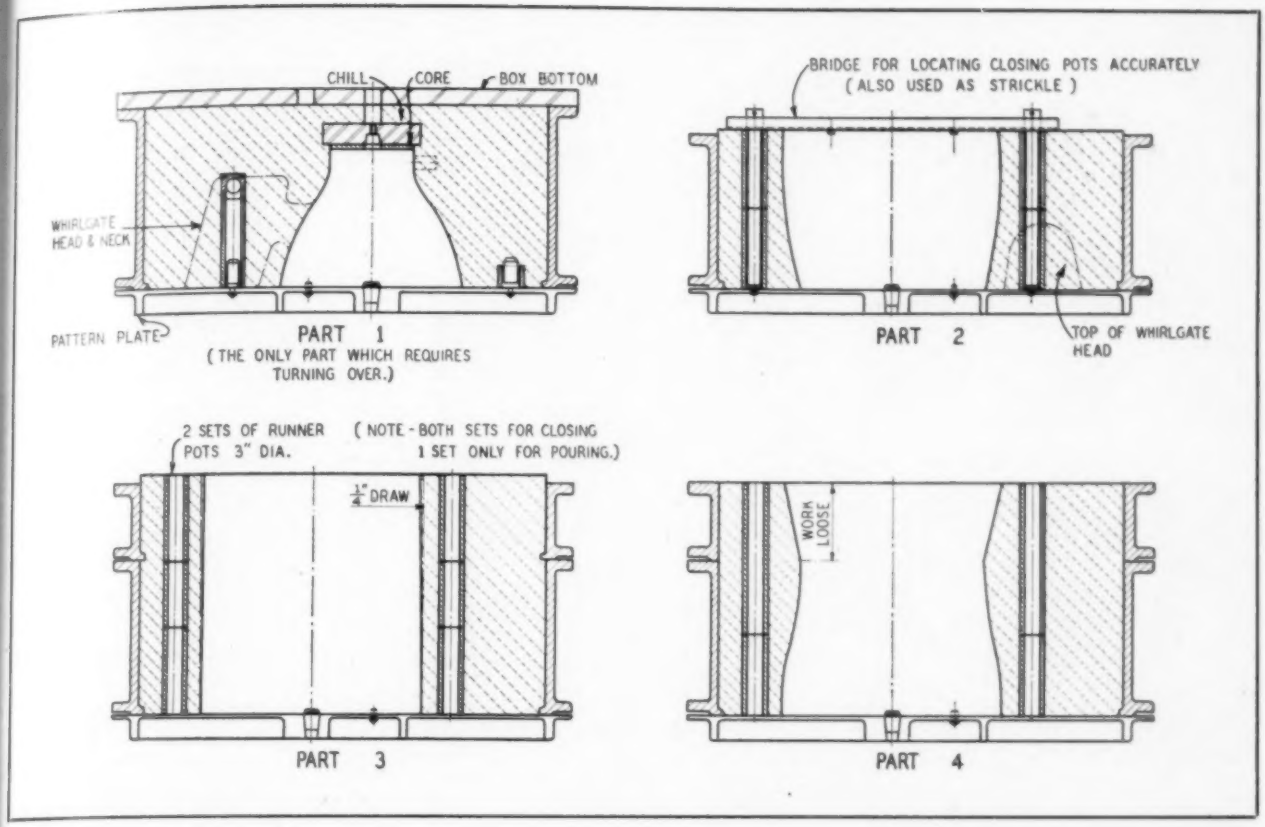


Fig. 2—Drawing showing design of molding boxes.

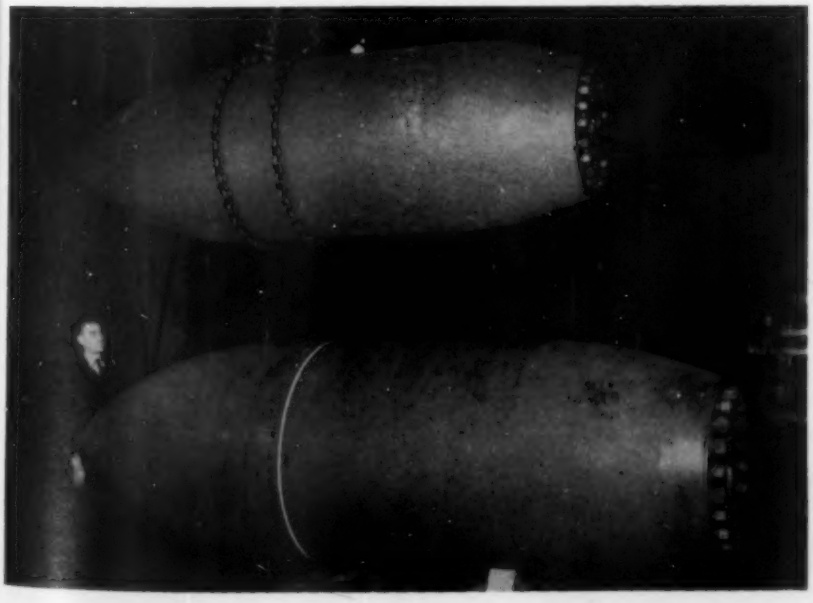
demand for heavy war production plant. In the same way, steel was not available in the convenient amounts obtained from the electric furnace, and only the 40-ton or 60-ton acid open-hearth furnaces could supply the quantity necessary.

The grand slam bomb required 10 tons of steel and the tallboy 6 tons, so that the foundry had to assemble eight tallboy molds or four grand slams at a time to make up a heat. One pit, enlarged to hold 12 boxes of tallboys, and a smaller one to hold four were available, but these pits could be served only by one crane.

Cycle for melting was 14 hr. from tap to tap, and it was essential for the foundry to be able to produce molds sufficiently fast to take successive heats if the rate of production was to be maintained. The molds could be removed from the pit after 8 hr. and knocked out after 24 hr., so that some 24 sets of boxes and tackle were required.

It was evident that the time taken to assemble the molds with the one crane was going to be the bottleneck in production, and that everything possible must be done to speed up the process.

Fig. 1—Photograph showing (bottom) the "grand slam" and (top) "tall boy" bombs.



Special Plant Out

In the 5th year of war the installation of a special plant was out of the question, owing to the tremendous demands of D-day on the engineering shops, and the methods of production now described were an improvisation designed to meet the special circumstances.

A high degree of accuracy was required, both in thickness of wall and in streamline form, to ensure true flight. The poor quality of timber available for patterns was likely to distort, and the problem of obtaining the large numbers of machined and pinned boxes required

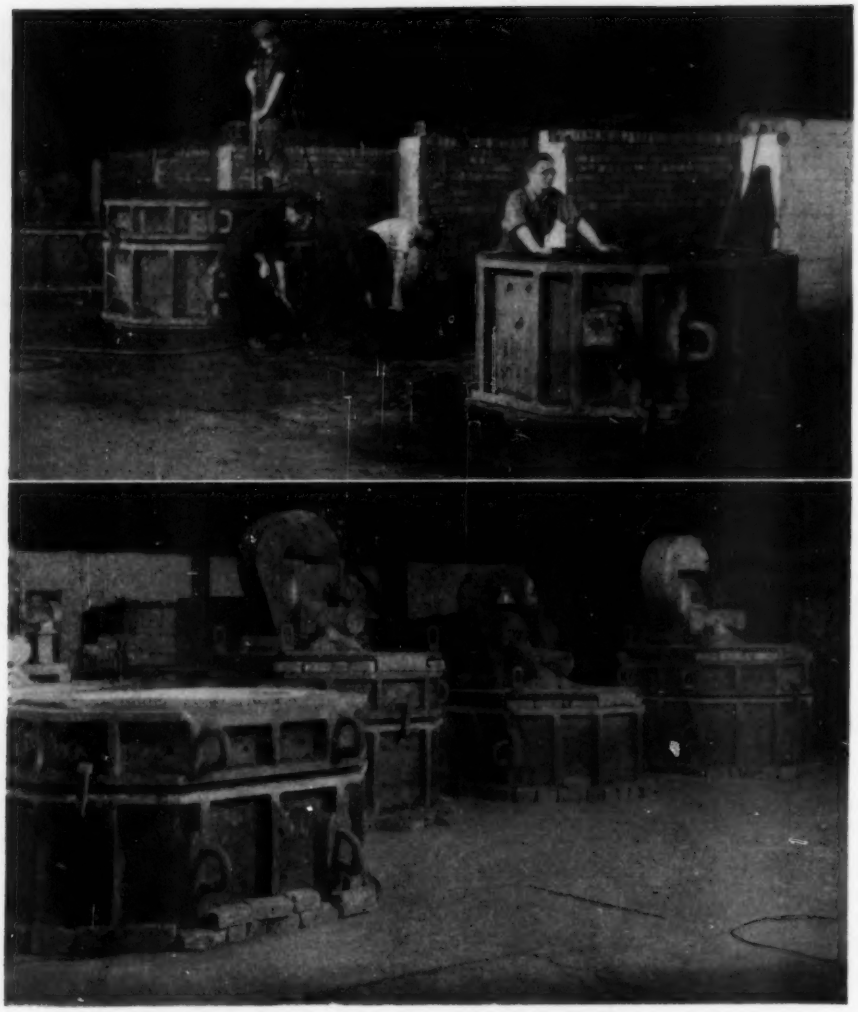


Fig. 3 (top)—Boxes at molding stations. Note method of clamping. Fig. 4 (bottom)—Method of drying boxes with portable coke-fired stoves.

for rapid and accurate closing seemed insuperable under the conditions prevailing.

On the other hand, sufficient lathes were available to machine the outsides, which were therefore given a machining allowance. One or two were cast to size and proved acceptable, but the molds took too long to close and more time was required for finishing in the dressing shop where the labor force was short.

Molds. Molding equipment followed ordinary practice, with patterns in sections on plates, and ramming by pneumatic hammer. Five boxes of special design, as shown in Fig. 2, were used for the tallboy, while the heavier bomb, of which fewer were needed, was made in a similar way, but in the foundry's standard shallow 6 × 4-ft. boxes clamped together in suitable numbers to bring the joints to the right level (Fig. 3).

Figure 2 shows the bottom part as molded, with a heavy chill, separated from the top of the pattern by a thin core. The chill serves to locate the core and chills the heavy nose end of the bomb. When ramming up is completed the sand is strickled off and the bottom plate bolted on. The other boxes shown call for little explanation, but a point worth noting is the fire clay pipe down-runner to the side head feeding the bottom.

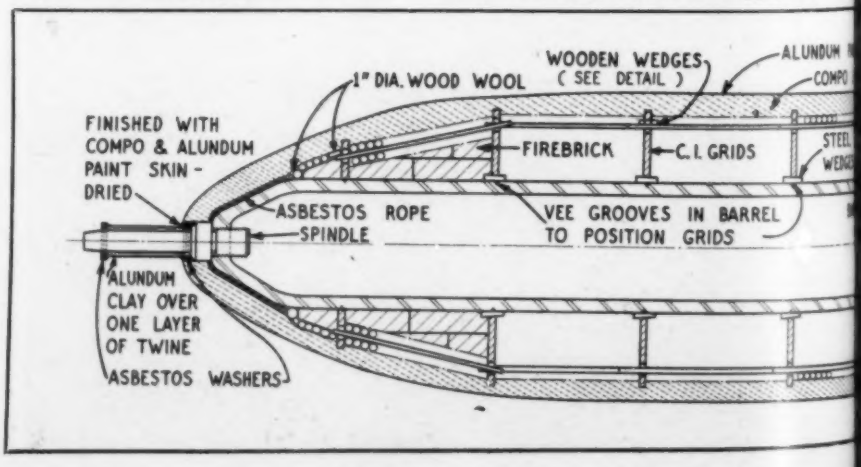
This type of head with a tangential runner is known in England as a "whirlgate," and has been much used in recent years. The down-runner and a similar dummy runner on the other side of the pattern were used with the help of a close fitting rod to locate the boxes in closing. The method proved successful and presents advantages over the more usual ball-and-socket device often employed in such circumstances. The boxes were dried with portable coke-fired stoves (Fig. 4).

Structure of the Main Body Core. Normal methods of core construction were considered but were not adopted because of the difficulty of obtaining satisfactory wood and because the slight adjustments that might be required to the dimensions would be more difficult. A still stronger argument was that the cores used in the earlier bombs had proved successful, and few foundrymen are in a hurry to abandon a success in search of problematical economies.

Core Barrel

As can be seen in Figs. 5 and 6, the main body core consists of refractory material built around a wire cage, or skeleton, mounted on a core barrel.

The more common practice in striking up similarly shaped cores is to build up the rough shape of the core by means of fire bricks, loam or compo* bricks on a steel barrel



*"Compo" is little used in America. The base is alumina ground in an ordinary pan mill with a clay bond. It was probably the only molding material used for steel castings in Europe until after the discovery that some of the Belgian sand deposits contained an almost perfect natural sand for the purpose. Compo originally consisted of a mixture of the discarded pots used in the coke crucible melting furnaces and old fire bricks. In recent years the supply of the

wound with wood rope; this rough, brick-built shape being covered with refractory material and strickled off to shape. The time required for building up a core in this manner is considerable, and the result is of rather solid construction, allowing less chance of contraction and consequent risk of "pulls" even though the barrel is pulled out as soon as the steel has set sufficiently.

It was considered that with a hollow cage the core would be quicker to make and the risk of "hot pulls" would be eliminated. Its construction could be broken down into simple operations requiring little skill, thus releasing the expert molders for the finishing work.

Inside Skin Good

In practice it has been found that of the many hundreds of these bombs which have now been made not one has been torn by contraction. Furthermore, the inside skin of the castings has been exceptionally good, and one can only conclude that this has been brought about by the absence of severe pressure from contraction between the skin of the casting and the molding material.

Core Barrel. Both the tallboy and grand slam bombs are manufactured in a similar way, the only difference being the size of the parts used. Therefore, in giving a description of the tackle the sizes will be given for the tallboy only.

The backbone of the core is a cast steel core barrel 12 ft. 6 in. long, 15 in. outside diameter, with walls 1½ in. thick. As can be seen from the diagram (Fig. 5), the nose end is closed and the tail end has a bridging star fitted. Both the nose end and the star are drilled and tapped to take screwed trunnions.

Vents Formed

At intervals of approximately 9-in. pitch, 1½-in. diameter, holes are bored or burnt in the barrel to form vents. This is done in order to allow any gases which form during casting to pass through and up the barrel to the atmosphere. At the tail end of the core barrel are two 4½-in. diameter holes.



Fig. 6—Method of constructing main body core.

These holes are used for suspending the cores in a special rack while waiting to be put in the molds, and also for pulling out the barrel after the bomb has been cast. The core rack can be seen in Fig. 6.

At intervals along the barrel, six V-grooves are machined in order to indicate the location of the grids, which are of "open cast" iron about ¾ in. thick.

Assembly of the Cage. The core barrel described is supported on two stands by the trunnions, and the six cast iron core grids are placed over the barrel and wedged into position over the marking grooves by two steel wedges. Through the holes in the grids mild steel rods of ½-in. diameter are passed and secured into position by soft wood wedges.

Nose End Solid

At the nose end of the core the cage is made solid by packing between the grid and the barrel with fire brick. This is done in order to strengthen the structure to withstand the upward pressure of the steel during pouring, but is not really necessary, as will be explained later.

The grid at the nose end, as can be seen in Fig. 6, has ½-in. diameter rods cast in. These are bent over the fire brick and wired down to give the desired shape.

Over this steel and iron cage 1-in. diameter wood rope is wound. This is done by one operator turning the whole barrel by means of a handle while another operator feeds the rope on. The rope has to be fed on

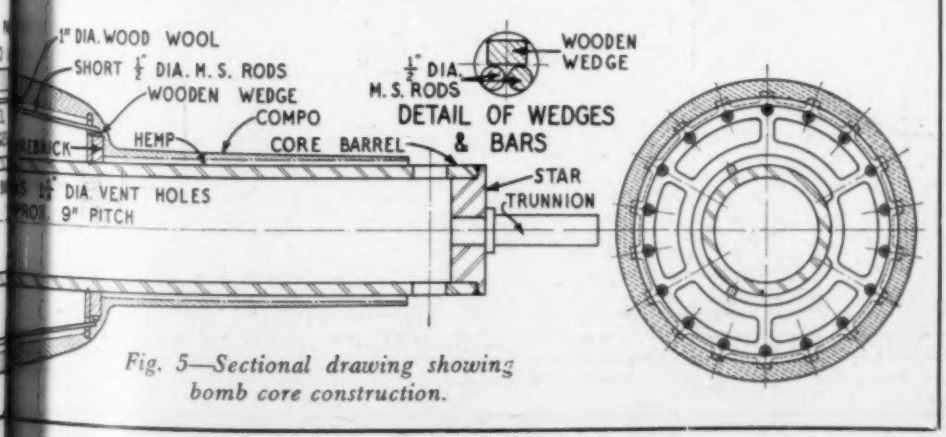


Fig. 5—Sectional drawing showing bomb core construction.

former has been limited owing to the introduction of high-frequency furnaces, and calcined high-alumina grog, as used in the manufacture of fire bricks, has been substituted. Where an especially high quality of face is required, as in the interior of the bombs, the mixture is made of the grog only, ground to a specified mesh. In Germany a similar mixture still is used, almost universally, except in light production castings. It is highly refractory when of good quality, and is mainly used for heavy castings. In this case it is used because it can be applied in a pasty condition, and is thus suitable for loam molding.





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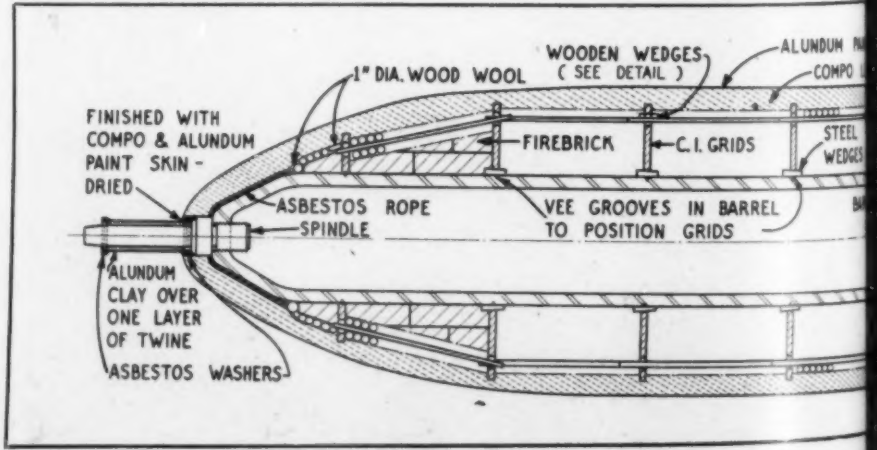
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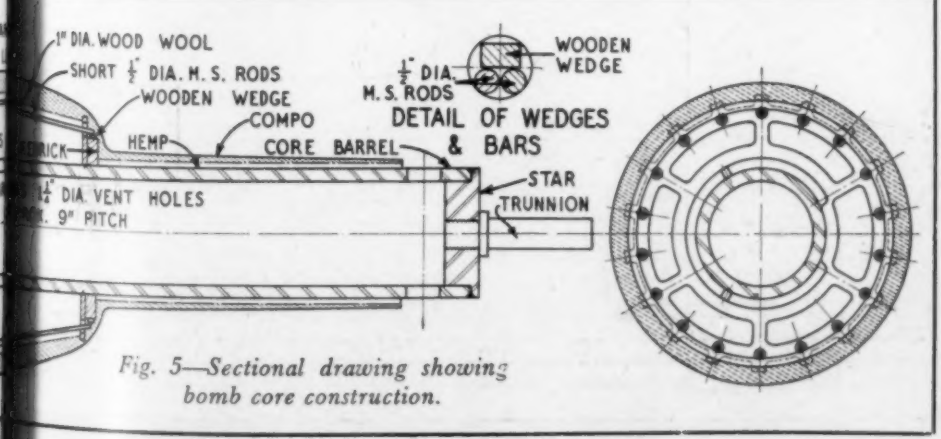


Fig. 5—Sectional drawing showing bomb core construction.

These holes are used for suspending the cores in a special rack while waiting to be put in the molds, and also for pulling out the barrel after the bomb has been cast. The core rack can be seen in Fig. 6.

At intervals along the barrel, six V-grooves are machined in order to indicate the location of the grids, which are of "open cast" iron about ¾ in. thick.

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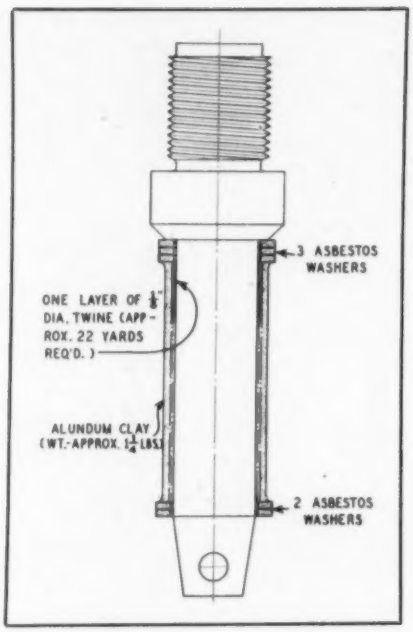


Fig. 7—Sketch showing core locating pin.

carefully so that it forms a tight continuous skin.

Coremaking. The coremakers now take over the assembly, which is still on the pedestals or stands. Their first operation is to cover the whole of the structure with semi-stiff "compo" loam. When the structure has been thus covered, long lengths of reinforcing ¼-in. diameter rod are laid along the length of the barrel and pressed down into the loam. Wire, 1/16-in. diameter, is then wound tightly round the barrel at 2-in. pitch, holding the rods in position and making further reinforcement.

After drying for 4 hr. at 400° F. in special stoves, the core is taken out and, while still warm, a further layer of wet "compo" loam ¾-in. thick is added and roughly strickled to shape. Due to the heat already in the core this layer of loam dries fairly rapidly, and the coremaker immediately applies a coating of chamotte (ground calcined fireclay) strickling to shape all the time.

This is followed by a coat of alundum paint. The whole core is then placed back in the stove and dried for an additional 2 hr. Between operations the strickle board is adjusted to give the thickness desired in each layer. The core is checked for size before and after drying, a profile template as well as calipers for diameter being used.

Making Nose End Core. The nose end core is built up on a specially screwed pin which replaces the trunnion in the main barrel. The other end of the pin is tapered to fit into the chill plate of cast iron in the bottom of the mold (Fig. 7). Both the trunnion and the chill plate are carefully machined as they form the locating point of the core.

The pin shown in Fig. 7 is mounted on a special strickling machine. One layer of ⅛-in. diameter twine is wound on and a coating of alundum loam ⅜-in. thick is strickled over the twine. An asbestos washer is fittted, as shown in Fig. 7. The whole of the trunnion is then placed vertically in a stove and dried at 660° F. for 2 hr.

The production department's idea of a cage to replace the old laborious method was accepted with approval by the experienced coremakers, but they were careful to anchor the longitudinal rods firmly to the bottom by bending the ends into a hook. Sufficient attention had certainly not been given to this point in the original design.

As production increased, inexperienced molders were brought into the work and this part was not so well done, with the result that the steel broke through a number of times owing to the core slipping up the barrel. To overcome this the core was filled in (at the nose end), although no trouble was experienced when the experts were on the job These were the only occasions of core failure throughout the work.

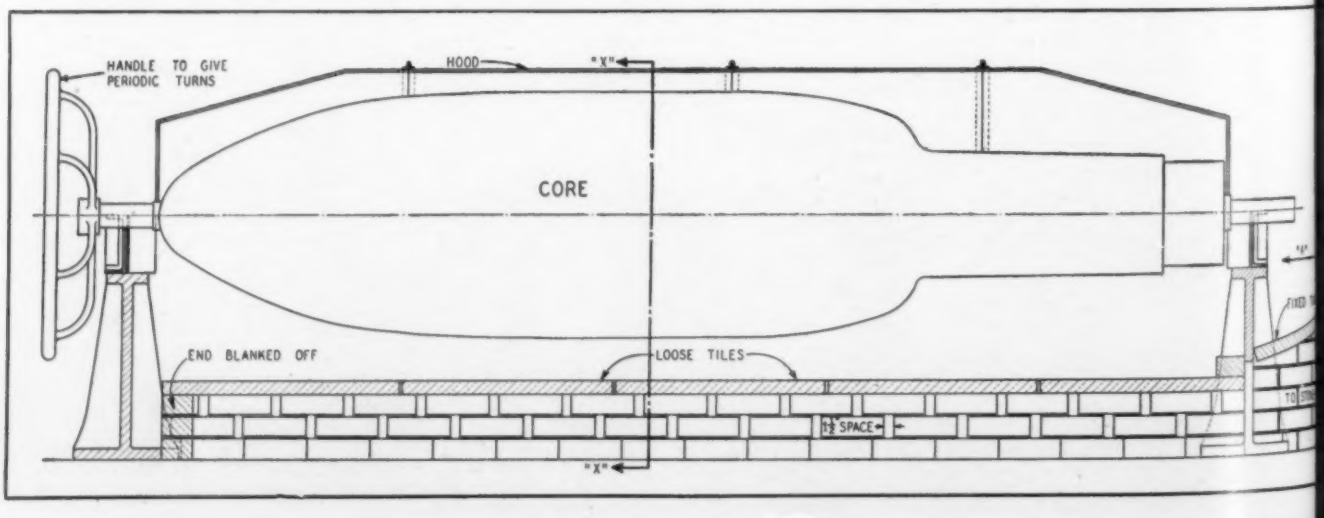
Special Coremaking Tackle

Core Drying Stoves. Owing to the output of bombs required it was necessary to design and erect a battery of special core-drying stoves, capable of quick loading and unloading. The design finally adopted, as shown in Fig. 8, consists of two trestles around which a brick wall is built.

Over the top of the brick wall light asbestos lined steel hoods are fitted, while running the length of the chamber is a brick-built flue with outlets for hot air at intervals The hot air is introduced by means of fans coupled to portable coke-burning stoves. A plate on either side of the air inlets induces a circulation around the bomb and an exit through vents at the bottom of the outer wall.

Temperature is controlled by thermocouples introduced at intervals along the stove. The cores are moved about during manufacture on

Fig. 8—Diagrammatic sketch of drying stoves for special bomb cores.



the special lifting beam shown in Fig. 6.

Core Transport Bogies. Owing to the geography of the shops it was necessary to transport the cores across two bays. In order to do this without damaging the core a special bogie was made of stock materials.

The core is rested on the bogie trestles and the trestle at the nose is pivoted, allowing the core to be up-ended by means of an overhead crane while still on the bogie, thus avoiding damage to the body of the core. The bogie is hauled by a locomotive on a carefully timed schedule, and is also used for the transfer of molding boxes.

Core Storage Rack. Figs. 9 and 10 show cores hanging on storage racks. This serves not only to store the cores but also as a means of holding them while the temporary trunnion is replaced by the locating pin with its refractory covering. Otherwise, this would have to be done while the core was hanging in the crane, which would take too long as the joint has to be made up and dried with a blow lamp. The core can be taken from the rack and set in the mold in a few minutes.

Assembling the Mold. The three bottom boxes often are assembled ready outside the pit to save time, and Fig. 10 shows two sets side by side, into one of which the core is being lowered. The woman standing on the edge of the mold is holding the wooden wedges with which the mold is temporarily stayed.

A hole is left in the bottom mold through which one molder can guide the core centering pin into its socket in the chill. A core is used

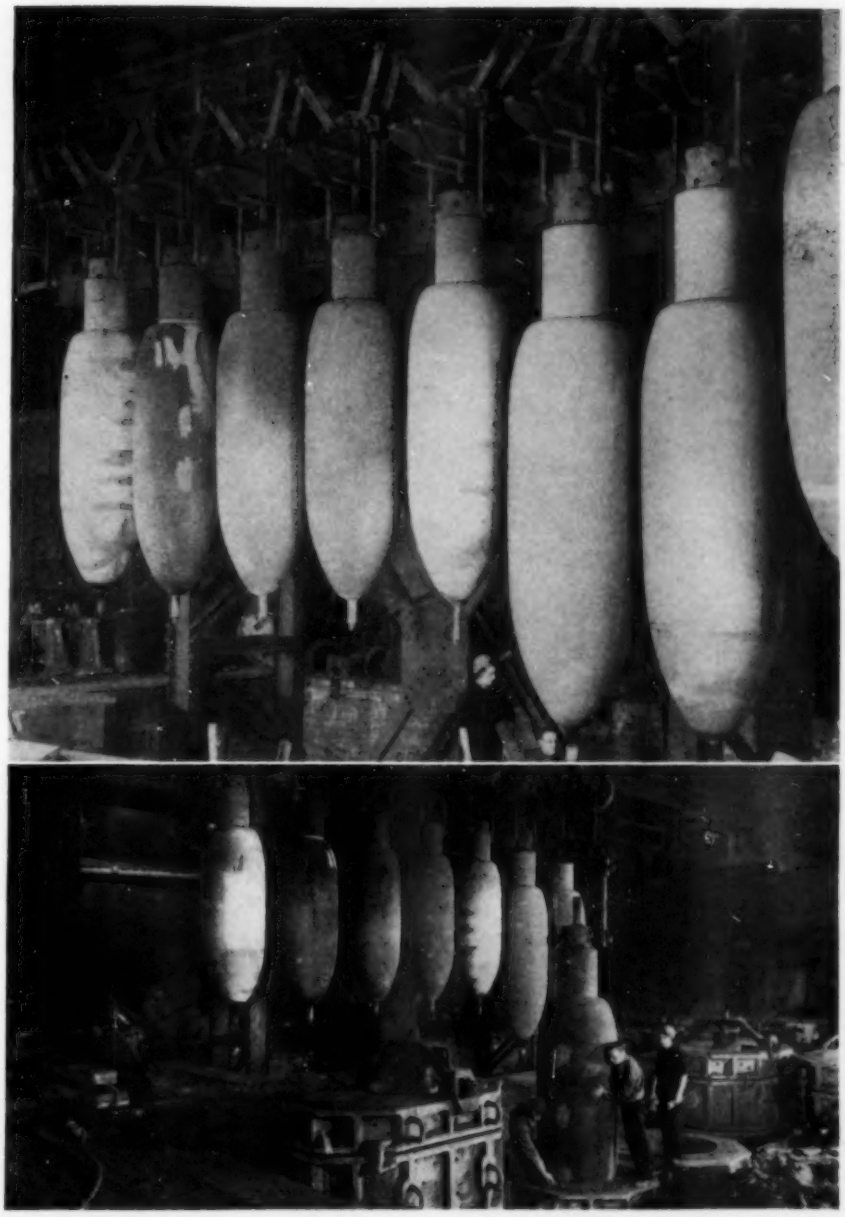
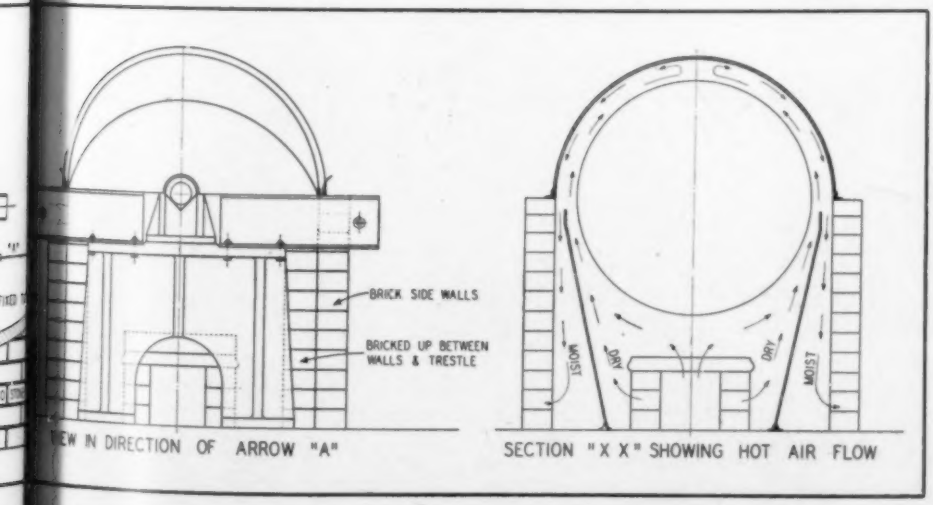


Fig. 9 (top)—Bomb cores hanging on storage rack. Fig. 10 (bottom)—Two sets of bottom boxes (3) assembled outside the pit to save time, with core being lowered into one set.



to fill the hole when core setting is completed.

The remaining boxes are lowered into position and the whole clamped up in the usual way. The core is held down by the special bar, which in turn is fastened to the box by the rings and wedges. The trunnion projecting through the holding down bar can be adjusted sideways by wedges to center the core in the mold, and the adjustment of the holding down pressure is also obtained by a wedge lightly driven between underside of bar and barrel.

Care must be taken in this operation as the wedges sometimes come

loose, if left for an hour or so, owing to the contraction of the barrel when cooling from the drying operation.

The arrangements described achieved their object in minimizing the time taken to assemble a mold to the extent that on several occasions the men closed a full cast of eight molds, from start to finish, in less than 3 hr., using only the one crane.

Casting. The steel in the smaller bomb is run with a 1¾-in. diameter nozzle in about 1 min. 40 sec. when the ladle is full, and care is taken to ignite the gases from the core as soon as they start to come out from the barrel. These gases were the principal cause of anxiety, as a hollow core had not been tried before and it was not known whether there would be danger of explosions.

In practice, no explosions occurred unless the steel broke in, as in the few cases previously mentioned. Even then there was no "blow" if pouring was stopped at the first indication of trouble, and at no time was the explosion dangerous or even alarming to those used to dealing with heavy steel castings.

Table 1

BOMB STEEL ANALYSIS

Element	*Content, per cent*
Carbon	0.30 — 0.34
Chromium	3.2 — 3.5
Molybdenum	0.50 — 0.60
Manganese	0.50 — 0.70
Nickel	0.50 max.
Silicon	0.35 "
Sulphur	0.030 "
Phosphorus	0.030 "

Core Removal

One-half hr. after casting the core barrel is pulled out with the overhead crane, the cast iron grids fastened to it are broken in the process and the core is thus most thoroughly eased. The full boxes seen on the right (Fig. 10) are stripped after 24 hr. The heads are burnt off hot and the castings charged into a furnace where they are annealed at a temperature of 1830° F.

Dressing. The roughing-off operation which follows is laborious with a solid core, but with the core construction used it is much easier. The remains of the grids are hooked with a chain and pulled out, which breaks up the core still further and makes it an easy matter to clean off the greater part of the remaining sand with pneumatic chisels and long bars.

Steel Used. A chrome-molybdenum steel with analysis range shown in Table 1 was used in the bomb.



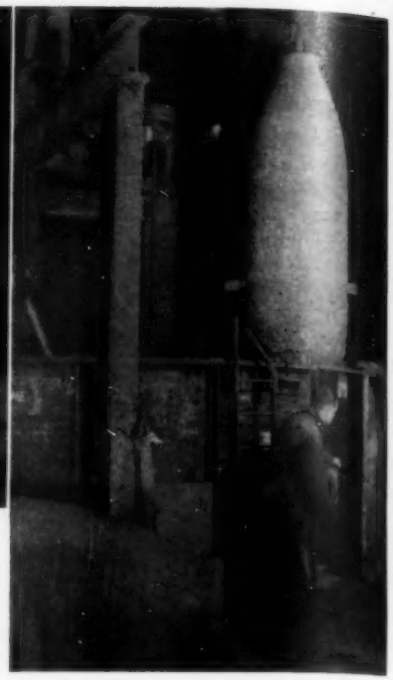


Fig. 11 (above)—Special fixture for oil quenching the bomb casting. Fig. 12 (right)—Shot blasting the bomb casting.

Heat Treatment. Final heat treatment of the bomb casting consisted in air or oil hardening from 1688° F. and drawing at 1112-1256° F., according to specifications that were amended from time to time.

A tensile and Izod impact test piece cut radially from a test ring removed at the base of the top head was prepared from representative castings, but each casting was Brinelled in six positions taken spirally from the top to bottom ends. Typical results obtained in these mechanical tests are shown in Table 2.

The special fixture for oil quenching is shown in Fig. 11. It consists of a frame with a fixed ring at one end which is threaded over the nose while the casting is on the furnace car. Another ring sliding on the frame is then pushed over the large end with bars and the casting can then be lifted by the overhead crane to a vertical position and quenched immediately in the oil.

Finishing. The inside of the bomb casting is shot blasted in what was a rotating table wheelabrator machine, altered to suit the work. The wheelabrator roof was removed and the bomb slung in its place from a beam. The bomb is rotated by fibre-covered friction wheels bearing on its outside and driven through V-belts by the motor that previously rotated the table.

A long blast pipe with nozzle at right angles is fed up the bomb by

Table 2

MECHANICAL PROPERTIES DEVELOPED IN BOMB STEEL BY HEAT TREATMENT

Property	*Heat Treatment*		
	A[1]	*B*[2]	*C*[3]
Tensile Strength, psi	105,280	106,624	133,952
Yield Strength, psi	75,264	81,536	103,040
Elongation, per cent	20.5	20.0	19.0
Reduction in Area, per cent	49	46	44
	30	55	19
Izod Impact Strength, ft.-lb.	25	52	18
	28	36	20

[1]Normalized at 1688° F. (920° C.), drawn at 1256° F. (680° C.)
[2]Oil hardened at 1688° F. (920° C.), drawn at 1256° F. (680° C.)
[3]Oil hardened at 1688° F. (920° C.), drawn at 1112° F. (600° C.)

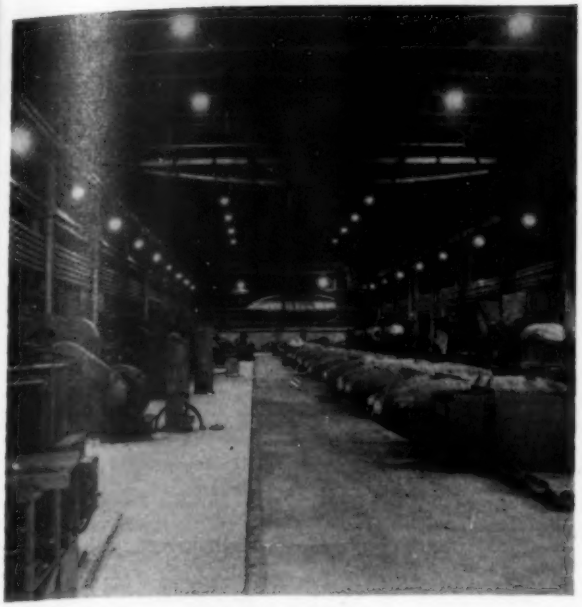


Fig. 13—View of machine shop where bomb castings were finished.

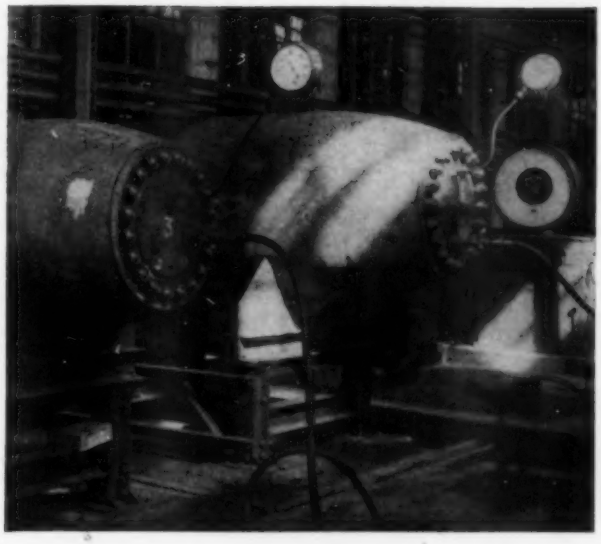


Fig. 14—Pressure testing of bomb castings.

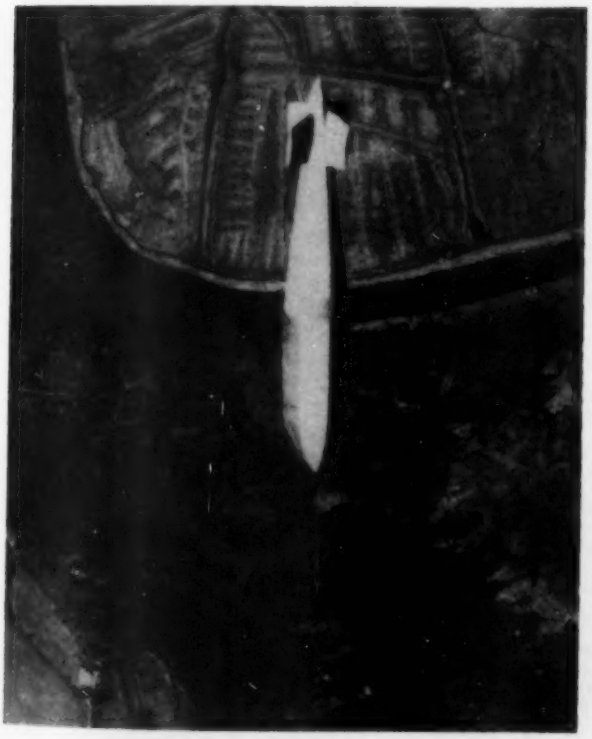


Fig. 15—Finished bomb with fin.



Fig. 16—Inside of U-boat pen at l'Orient after bombing.

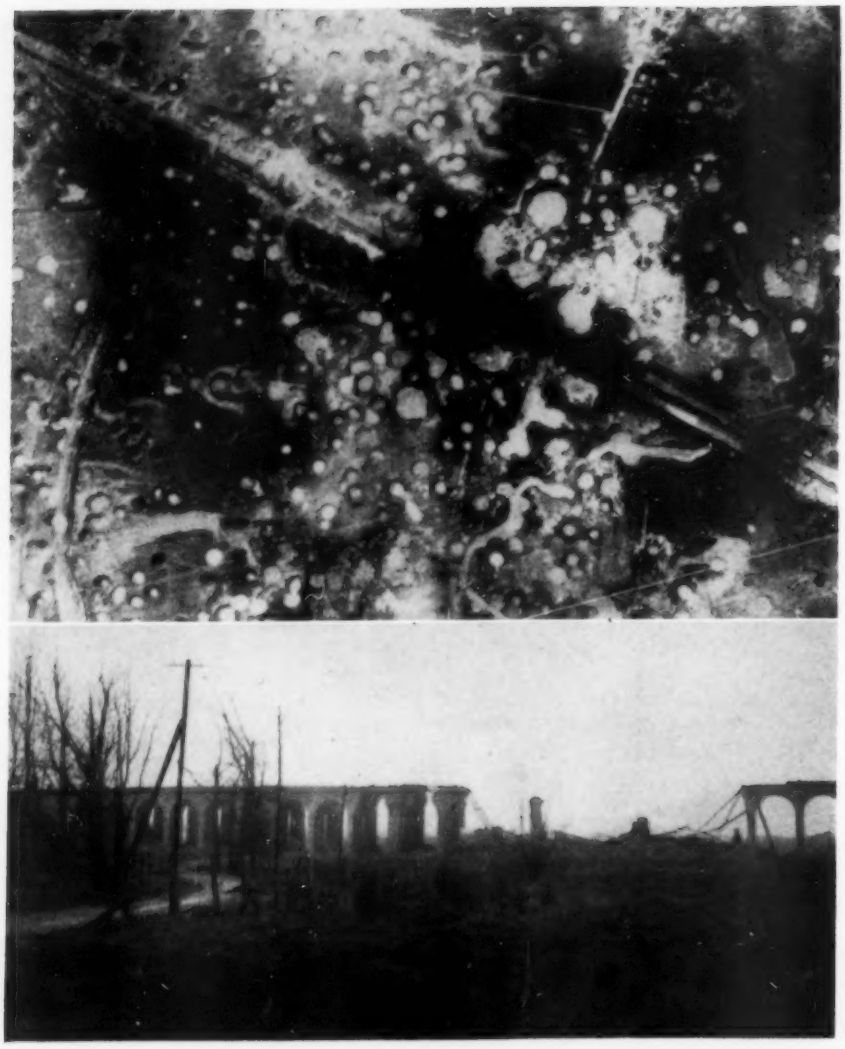


Fig. 17 (top)—Bielefeld railway viaduct from the air. Fig. 18 (bottom)—Same viaduct (Fig. 17) after near miss by "grand slam" bomb.

means of the winch and wire rope shown in Fig. 12. The success of the core in giving a clean skin results in little further work being required on the inside after shot blasting, except at the nose end which cannot be reached by the shot. A smooth surface is required to avoid danger while filling with explosive, but, in spite of that, welding was seldom necessary.

This is not the place to describe the machining operations, but a few views of the machine shops (Figs. 13 and 14) may be of interest. Figure 15 shows the bomb as carried in the aircraft, its tail being about 22 ft. long.

Targets. A short account of the results obtained with the bomb may be excused, although the destructive power of the atomic bomb tends to make all earlier bomb types seem obsolete. Nevertheless, the 'Grand Slam' bombs of the R.A.F. had proved themselves the most destructive and at the same time the most accurate bombs of their time.

Mr. Wallace, already famous as the inventor of the "geodetic" construction used in the Wellington bomber, and of the special bombs of another type that were used to blow up the Mohne dam on the River Ruhr, conceived the idea that the detonation of a bomb with a heavy bursting charge that would penetrate deep into the earth's surface would produce earth tremors amounting to a small earthquake.

Penetration was to be obtained by the weight of the bomb and the height from which it was dropped: accuracy, by careful streamlining and a weight that was little affected by air currents; the earthquake by the explosion of a bomb with a light container of high tensile alloy steel and the heaviest possible bursting charge.

The U-boat pens at Hamburg, a relatively small target were hit by 12 out of 18 bombs, a large proportion perforating the 20-ft. roof of concrete with which they are provided. One of the near misses is thought to have sunk two ships at anchor in the harbor nearby. Figure 16 shows the inside of a U-boat pen at l'Orient after a bombing.

Another target, shown in Fig. 17, is the Bielefeld railway viaduct which carried the last uncut main line railway to Belgium and the Ardennes, during the fighting there in the spring of 1945. Ordinary bombs had been showered like hailstones at this bridge but had only chipped off bits of it.

The first grand slam bomb that was dropped just shook the masonry to pieces. It was a near miss, from an altitude of 15,000 ft., about 160 ft. to the north (Fig. 18). The subsoil was so badly shattered that it was thought hopeless to bridge the gap and the railway was diverted.

This paper described one method of making this somewhat unusual casting, but it should be mentioned that it has also been made successfully by more normal methods in another foundry in Scotland, and also in a foundry in the U.S.A.

It has also been made in the U.S.A. from a welded tube to which castings were welded to form the nose and base. Nevertheless, there is no doubt that if the job had to be started again little alteration in method would be made in the foundry whose work is described.

A sand slinger with a pattern draw machine would improve the molding, and some improvement might be made in the construction of the cage for the core. The cost of the core may be a little high, but that is more than offset by savings elsewhere, particularly in the dressing shop.

Acknowledgment

A special word of acknowledgment is due to Mr. C. A. Chaytor, Foundry Manager, Messrs. Whitehouse and Shepherd, Foundry Engineers, who were directly responsible for actual production and design of tools, for their assistance in preparing this paper.

The author is indebted to the R.A.F. for the action photographs, and to both the R.A.F. and the Directors of the English Steel Corp. Ltd. for permission to publish the paper.

DISCUSSION

Chairman: J. W. JUPPENLATZ, Lebanon Steel Foundry, Lebanon, Pa.

Co-Chairman: J. B. CAINE, Sawbrook Steel Castings Co., Lockland, Ohio.

CO-CHAIRMAN CAINE: Can you make a comparison between chamotte and sand, and give the advantages and disadvantages of each?

MR. GRAY: In England, chamotte is usually called compo and used to be made, mainly, from old coke-fired crucible pots, used fire-bricks and ganister. It was very cheap then, as most of the steel in Sheffield was made by the crucible process.

Presently, it is made with the crucible pots omitted and 20 per cent of calcined grog, such as is used for the best quality fire-bricks, substituted. The ganister, also, is omitted, the remainder of the mixture being old fire-brick and clay with 10 per cent of water. The grading is 50 per cent minus 6 B.I.S. sieve plus 22 B.I.S. barrel is pulled out of the mold, which

The author forgot to mention in the paper that within ½ hr. of casting, the sieve with 27 per cent clay and silt grade.

This mixture stands up to 1600° C. firing quite well. A typical sample of German grog from Bochum Verein, Weitmar, gave a similar grading, but a test piece fired at the same temperature squatted and was full of pin holes.

Good quality chamotte is more refractory than ordinary silica sand, and is excellent for really heavy castings, but gives a decidely rough skin when used for small castings. Molding with chamotte requires a special technique of its own. As men that are really expert in its use are becoming scarce, we should prefer to use a silica sand if any was available in England of the type that is used in America for heavy castings.

Chamotte requires to be dried at a minimum of 400° C., and preferably at 600° C., to drive off some of the combined moisture. It is, therefore, more expensive than silica sand in drying costs.

The loam used for the cores of the bombs was made of backing sand in which there was a considerable portion of old chamotte faced with a mixture of pure grog and fireclay of which the grading was minus 30 B.I.S. sieve 88 per cent. The paint is made up of chamotte fines, ball clay and dextrine.

breaks the grids and leaves only a thin shell of sand an in. or so thick. Certainly, the skins left by these cores were some of the finest we have ever produced.

MEMBER: Why was oil quenching preferred to water quenching on this particular job?

MR. GRAY: We did not have a water quench tank that was deep enough for the casting.

H. R. MCCOY[1]: Could you tell us a little more about the type of wash used in the connection with chamotte? I understand that is was sulphonate wash, rather than a silica wash.

MR. GRAY: It was the same material, only ground finer. It was applied with the paint brush or spray. We use both. It was then slicked over by hand until smooth.

J. A. RASSENFOSS[2]: How long did it take to make one of these cores?

MR. GRAY: It took two men six hr. I do not think it is as economical as making the core in core boxes, but it saves an enormous amount of time in the finishing work. Also it gives greater collapsibility. We made thousands of these pieces without a hot crack in a single casting.

[1]Ohio Steel Foundry Co., Lima, Ohio.
[2]American Steel Foundries, East Chicago, Ind.

TECHNICAL ORGANIZATION of the FRENCH FOUNDRY INDUSTRY

Official exchange paper presented on behalf of the French Foundry Technical Association at a Gray Iron Session at the Fiftieth Annual Convention of the American Foundrymen's Association, at Cleveland, May 9, 1946.

A. Le Thomas
President
French Technical Committee
of Foundry Industries
Paris, France

TWENTY YEARS AGO, the author was requested to present, in the name of his country, an exchange-paper to the International Foundry Congress in Detroit, 1926. He then chose a thoroughly technical subject. Owing to the fact that he always was strongly attracted to pure technic, he devoted that paper to laboratory researches, workshop adjustments, adaptation of metallurgical and foundry possibilities to structural engineering requirements. Now, although his great desire for such knowledge and improvements has not lessened, he has had to devote himself to the task of taking up a broader scope which is very important and urgent for the foundry industry of his country.

Present Day Problems

At the present time the up-to-date methods of work and machinery present problems of great extent and extreme complexity in many of the French foundries. In what follows the writer shows why these problems are put forth and discusses the methods which have been adopted to solve them.

It is necessary to follow the evolution of the French foundry industry, to recall the leading rank it held before 1914 and the considerable technical effort it furnished in the course of the first great war.

Although one-third of its territory, including the regions which were best provided with mines and metallurgical plants, was then occupied by the invader, France had nevertheless succeeded in performing a decisive part in the industrial production adapted to the making of armaments (artillery, aviation, etc.), which allowed it to provide military equipment to a certain number of our allies.

One can easily understand in what state the French industry found itself after World War II. Therefore, we must build again,[1] modernize our machinery and revise our methods.

Industrial Framework

Framework of the French Foundry Industries. The prevailing feature in the structure of the French foundry industry is the fact that it is scattered over the country with a great number of small and medium sized firms. For instance, in 1938, the last normal year before the war, it was estimated that of a total of 1,800 foundries more than half employed less than 20 workers each, with a combined production amounting to less than one tenth of the total trade.

A little over one-third numbered 20 to 100 workmen each, and produced together about 40 per cent of the total output. The remainder, being only 12 per cent of the concerns, had an effective force greater than 100 workmen, producing over half the total foundry output.

There were hardly 200 firms with more than 100 workmen each, among which were less than 15 exceeding 1,000 workmen. But these are relatively old statistics which do not show the present condition in the state of under-production and general ruin where the whole of the French economy is still struggling.

Before discussing the consequences which result from these conditions, just another word on the geographical positions of these firms, which are also natural data of the trade, and must be taken into consideration. Most of these foundries are grouped in the regions of industrial concentration of the country (especially where coal or iron ore is to be found, or in the neighborhood of dams)[2] such as: the Paris districts, the North, the East, the Center (Lyons, Ste.-Etienne).

Besides this, iron foundries especially can be found in other districts sometimes very crowded, such as the Meuse and above all the Ardennes, a curious survival which testifies to ancient spots likely to be found in a very old country whose industrial past goes back many centuries.

Small Foundries

This predominance of small concerns (nearly nine-tenths including less than a 100 workmen and representing one-half of the total production) has without a doubt weakened the industries, chiefly considering the seriousness of the world economic crisis that raged from 1931 and

[1] It is estimated that the reconstruction of destroyed buildings in France amounts to nearly the number of houses built in one century. For example, nothing is left in the author's native town (Brest, on the Atlantic) but one house fit to live in.

[2] The North, the East, the Paris districts, the Ardennes, etc., in conjunction or very near the frontier, are very vulnerable to attacks. This fact, which might not be well known abroad, explains a lot of the behavior of France.

from which France had scarcely begun to recover in 1938.

Consequently the result was, on a national market already confined by nature, an implacable competition bringing with it a progressive lowering of prices and quality, so that outside a few powerful concerns which had understood the prime value of technical research and could assume the financial cost, the industry counted few real laboratories.

This state of affairs had not failed to worry clear-sighted people who, before the war, visioned in the voluntary cooperation of firms the only solution to the weakness of their individual means. This policy manifested itself in agreements, purchasing groupings, sales combinations, and training of staff chiefs and handicraft specialists. But we shall specially mention here the technical cooperation, from which was developed the French Foundry Technical Association which had almost a thousand members in 1938.

All these initiatives were registered in a large trade-union body which gathered the whole trade into an organization, The General Syndicate of French Foundries, directed by M. Maurice Olivier, whose clear thinking and social sense is admired by all French manufacturers. With few exceptions all foundrymen voluntarily belong to this council, and join afterwards with councils of specialists and local councils closely connected with the general syndicate.

Organizations Necessary

We had already anticipated that war has rendered these essential organizations more necessary. Without even mentioning war damages (destructions, bombardments, and lootings, which went on from May, 1940 to Spring, 1945), estimated to about $400,000,000 (at the 1939 rate of exchange) for the entire trade, it is evident that in the course of the last six years, all the industrial concerns have been greatly affected in their inner works. Skilled workers and staff chiefs were deported, dispersed, and underfed. Machinery and equipment could not be cared for or renewed. Stocks, treasury and stores of all kinds little by little vanished in a penury never heard of before.

At the same time as these common calamities pointed out to our industrialists that strength was in the future, the problem of professional reconstruction took such a hold that it has gone beyond individual initiative. Only the government could provide the indispensable material means necessary for the rebuilding sought by the French foundrymen. They did so by securing in a legislative measure means for financing of technical research.

Research Grants

A tax established a year ago on the price of foundry products is especially granted to provide for the expenses of studies and technical research of the industry. One half keeps the Technical Center of the Foundry Industry, an autonomous establishment which is entirely devoted to the study of all the trade technical problems.

The other half is in principle taken over by firms, provided it is used for their own technical studies. This is, therefore, a double operation, on one side individual, each foundry on its own account; on the other side collective, in the form of a professional organization managed by the trade, open to all its members, who discuss the vital questions of general interest and at the same time coordinate the individual efforts. We shall now see how this works in practice.

The French Foundrymen's Union for Technical Progress Research. The improvement of technical work of the French foundry industry is the result of the following cooperating organizations: Association technique de Fonderie (The French Foundry Technical Association), l'Ecole superieure de Fonderie (Superior Foundry School), the finishing courses and those concerning training, and le Centre technique des Industries de la Fonderie. (Foundry Industry Technical Center).

Economic Interests

These organizations are themselves closely connected with those in charge of economical and social interests, as it is a fact that technical development has no other aim but the prosperity of concerns, of the men who work in them and of the whole country. It cannot give the expected results without the help of a lasting social rule and well-seated economy.

(The French Foundry Technical Association). It is hardly necessary to extol the French Foundry Technical Association, so old, solid and cordial are the bonds that unite it to the American Foundrymen's Association. As it has an approximately equal number of manufacturers and engineers belonging to the French foundry industry, and also an important number of foreign members, it takes an active part in lectures and foundry conferences.

Its organization goes back to 1911, and we know that many American fellow-members remember one of its organizers, Eugene Ronceray. With the faith of an apostle, he devoted the best part of his activity to promote the ideas of technical progress, which is supported by our Associations.

The present chairman of the Association is M. Paul Bastien, a young but already great scientist in metallurgy and physics, a professor in one of our foremost engineering colleges (Ecole Centrale des Arts et Manufactures, Paris) and scientific manager of Schneider and Co., whose Creusot factories are famous all over the world for the extent and quality of their products.

Monthly Magazine

Right up to the last war, the works of The French Foundry Technical Association were regularly published in a monthly magazine. During the war, the situation was difficult. Without getting down-hearted, the acting chairman, M. Lobstein, an extremely courageous man, to whom the French industrialists always refer and give a foremost place in their councils, and next M. Chevenard, one of the cleverest metallurgical scientists in the world, have succeeded in carrying on their predecessor's marvelous work. The monthly meetings have been held but individual lectures replaced the editorial publication.

Actually, the lectures of The French Foundry Technical Association are published in the review, *Fonderie,* a common publication of the Technical Association and the Foundry Industry Technical Center, which is unique in France.

(The Superior Foundry School). Of course, there are in France a certain number of engineering colleges, of different grades. Some aim to develop practical knowledge, others are

based upon a scientific formation extremely far reached, while others are of all possible grades between these two extremes.

The Superior Foundry School admits young men already possessing a serious scientific education, generally verified by a university degree, having on the other hand a practical knowledge of the foundry work. This means that they have previously worked in a foundry, and they therefore, finish their training with one year's study, in all details of the industry, from metallurgy to molding.

Foundry School

The Superior Foundry School was formed in 1923 by the General Syndicate of French Foundries, under the inspiration of Eugene Ronceray, who acquired on this occasion, a new title to the gratitude of the foundrymen. It is administered by a council board in which professional representatives and delegates of the Ministers of National Education participate.

It is now directed by M. Albert Portevin, one of the scientific and metallurgical persons whose true fame, long known in our country, has reached far beyond the boundaries of France.

The Superior Foundry School each year, trains about 20 engineers, who play a first-class part in our factories.

Training and Finishing Courses. We have shown the diversified structure of the French Foundry Industry that includes a great number of small concerns, to which mechanization has not yet been extended as much as one would wish. The need of excellent workmen, possessing high manual skill is greatly felt in our country, and we can say, that we have at our disposal many excellent, trained workmen in molding and patternmaking.

Staff Vacancies

However, we have to renew constantly the vacancies that occur in our staff, occasioned by age, death, changing of situation. Our country has undergone an awful loss during the course of successive wars.

Training, in France, is the work of legal dispositions, moreover, very elastic, as each industrialist can train his own apprentices through their working on the products he manufactures. This is carried out with a great deal of discrimination under the supervision of state control.

In fact, in most cases, the manufacturers of the same districts group together in their syndicates to sponsor a common course of apprenticeship, that brings together children of fourteen leaving grade school with a knowledge of the French language and history, geography, and arithmetic.

The training lasts three years, during which time the young men receive a complement of general instruction (geometry, drawing and technology) and learn the trade of a molder, coremaker or patternmaker in a workshop, generally in a specific section of the plant, according to a rational program, with special care and attention from a committee efficiently directed by M. D. Waeles.

Besides, the best organized schools develop a sporting competitive spirit in the young men through physical education. The reward is a Certificate of Professional Aptitude for molding, coremaking or patternmaking, which assures to the young men who possess it to a good beginning wage as a skilled worker in foundries.

Finishing Courses

Then the better workers take the finishing courses for three years during which time the young workman completes his training in the same workshop where he is employed, receiving at the same time, out of his working hours, a further training in algebra, trigonometry, physics, chemistry, electricity, drawing, and molding. At the end of the course, the proficient pupils receive a written professional statement certifying they are first class workmen.

On the other hand, the abilities to perform supervisory work corresponding to the duties of foremen, are determined by a technical test which show the students' reaction to certain problems involved in definite situations. Such a test has been shown to be practically infallible.

Moreover, the best pupils are sent to The Superior Foundry School where they meet young men generally from universities, undoubtedly less skillful in handicraft, but having a higher scientific training.

We must note here that the Directeur de l'Enseignement technique en France (Director of Technical Education in France), is a high official on whom depends the training and finishing courses in The Superior Foundry School (and Institutions d'Enseignement technique corresponding to all crafts). He is a great scientist renowned for his studies of physics of materials, M. le Rolland, who has developed an apparatus to measure Young's modulus of elasticity based on an incredibly simple principle, which may not be known enough outside of France.

M. le Rolland belongs by personal ability to the Conseil d'Administration (Administration Board) and to the Scientific Committee of the Foundry Industry Technical Center. So, the technical formation of foundry workers and engineers is entrusted to a highly competent person.

Technical Center

The Technical Center of the Foundry Industry. Birth of the Technical Center. There are in France a great number of small and middle-sized foundries. We have also shown that in consequence of various circumstances met in the period between the two wars, and yet accentuated in the course of these last years, it was difficult for these foundries to perfect their technic and to modernize their equipment as much as they would have liked.

It was necessary for these foundries to unite in showing what each wanted and what equipment each had to sell. In this movement the government had to support their effort. This was accomplished by the creation of the Technical Center of the Foundry Industry which is endowed with powerful means of action and a very supple administration. This organization belongs to the industry and is, at the same time, under State Control which, however, leaves it with freedom of action, especially concerning the details of its daily activity.

Results Obtained

Since 1931, the necessity of this effort, in common with the plan of the technical factories, was felt. In order to complete the excellent results obtained by The Superior Foundry School, The Foundry Technical Association and the training

and finishing courses, one of the leading producers of foundry pig-iron, la Societe des Hauts-Fourneaux de Saulnes, whose mines and factories are situated on the phosphorus ore fields of Longwy-Brisy so well-known in Europe, had taken the initiative in creating a Research Center.

The latter, placed under the direction of the author and staffed with specialists and having an important laboratory, was put at the disposal of iron foundries wishing to obtain information of a technical nature. It seems that by this means great services have been rendered to the French foundries as well as to the Belgian, Swiss, Italian and Luxemburg foundries which had great confidence in it.

La Fonte Published

These services were of an essentially practical nature, and no fee was demanded. Moreover, the research center of the foundries of Hauts-Fourneaux of Saulines published a review *La Fonte* to which many foundrymen in all countries had subscribed in consideration of its outstanding scientific and technical standing.

Further, in 1938, the author had been charged by the Syndicate of Brass Foundrymen to extend his services to its benefit. But the first results of this new activity had hardly been obtained when war broke out. The French Minister of armaments then placed the author in charge of developing the manufacture of projectiles in very high strength cast iron.

In this work he was aided by one of his assistants, M. Guy Henon, who had acquired part of his training in the United States with the firm of Otis-Pifre, and a few other engineers of whom M. Jean Laine was actually sent on an official mission for buying machinery in the States. A satisfactory result was obtained in a few months.

Interests Upheld

After the invasion of France, the situation was evidently changed. However, the industrial foundry companies, supported by the officials of the Minister of Industrial Production who had the extremely difficult task of upholding the interests of France before the occupation authorities, did not despair of preparing our industry for the future welfare of which there was no doubt.

A great effort, in the specialization of the works and mutual appreciation of the fundamental economical interests which prevail in private concerns, was then undertaken. It is fitting to list here the names of those who were the promoters of this revival of the industry.

Among them are Engineer General Norguet, the builder of the splendid French war fleet, who was then committed to direct the French mechanical, electrical and foundry industries, M. Olivier, president of The General Syndicate of French Foundries, and M. Ricard, whose name is connected to the reorganization of our industries and now awaiting still more important duties, to which he seems to be destined, because of his extreme intelligence, remarkable conciliating mind and vision as to possibilities.

Technical Center Activities

It was under these circumstances, even during the occupation of Paris and France, in 1941 and 1942, that there was created The French Foundry Technical Center with extremely reduced means, but a minutely prepared program with the future in view.

It was first necessary to keep the foundries from dying out. The Technical Center endeavored to find solutions for the problems occurring each day caused by the constantly growing scarcity of raw materials of normal quality. An especially strenuous effort was made to prepare, for the future, keeping in mind the coming liberty which none of these men ever doubted would be achieved.

Chief Activities of the Technical Center. And so came the birth of technical services bringing together specialists charged with the solving of technical problems laid down by the foundrymen. By correspondence, inspections and sometimes lengthy stays in the factories, these engineers contributed to the re-adjustment of any disturbance in the manufacturing processes, working in complete harmony with the local leaders.

Successively, more and more requests were made to the Technical Center as soon as the success of these first surveys were confirmed.

Facing a great number of requests, among them many relating to subjects of a higher order than mere and simple assistance in case of sudden difficulties, the Technical Center has little by little, since the liberation of the country, extended its technical services by adding to its staff highly qualified engineers from the branches of the foundry industry chosen from the best technicians in their special fields.

Engineer Specialists

These engineer specialists are now each directing a technical service office covering fields such as cast iron metallurgy, steel metallurgy, copper alloys metallurgy, light alloys metallurgy, molding, raw materials, casting surface study, plant and equipment and factory organization. They play the part of real "physicians" to the foundries and answer all requests for advice by foundrymen.

They have, at their disposal, the laboratory of the Technical Center, and can thus base their findings on the results of analyses, tests and researches carried out in this laboratory. Their consultations have always been given in a precise and conscientious manner and in an extremely practical spirit which is greatly appreciated by the foundrymen.

The offices of the Foundry Technical Center are located in Paris. To meet the particular needs of certain regions, where foundries of the same type are concentrated, local representatives have been appointed. These local representatives do not have the same expert knowledge in all the details of their special field as do their Paris fellow members but are able to solve the more current technical problems. These regional representatives have, at their disposal, laboratories maintained by companies and syndicates of the regions, where they operate.

Future Considered

It is not sufficient to think only about the present, although in France at this time, such study of immediate problems is really of the greatest importance, the future has to be considered. To be prosperous, an industry must progress, vision and create new things. The creators, rich in ideas and capable of applying

them practically, are not lacking in France, a country, where technical ability was of the highest order in the world before the series of wars it has gone through.

But, as we have said, it is the quantity of material which is lacking. In consequence, we have wished to coordinate the use of those remaining and, to this task, the Foundry Technical Center answers with two series of complementary activities.

Research Director

Its services include first, a research director whose duty is to vision, organize and coordinate scientific and technical studies that show the need of change, modernization and restoration of working methods in foundries. The director of research, M. Jean Laine, will be able to devote his time and service to these projects benefiting by the very valuable information he collected in the course of his recent journey to the United States.

Actually, the Technical Center does not have at its disposal a research laboratory possessing all the necessary apparatus to continue industrial tests (melting furnaces, machinery and all sorts of apparatus) but this station will be erected in the coming years. Its present laboratory contains the strictly necessary apparatus for ordinary work. A complete research laboratory will also be built, meanwhile the better equipped among the laboratories of factories and universities are used.

On the other hand, the whole trade joined together to participate in the research program. This is the work of technical commissions, which group all the best of the French foundry technicians under the direction of the Technical Committee which is composed, under the author's chairmanship, of the president of the Scientific Commission, M. Portevin, the president of the Technical Association, M. Bastien, and of the technical commissions.

Technical Committee

The Technical Committee plays the part of determining, and then solving the problems of general interest which make up the technical program, in a general policy suited to the needs and possibilities of our country. It is striking to witness the wonderful team-spirit and the professional consciousness which animate the few hundred and twenty scientists and technical members of our commissions.

It is thus they communicate, without holding anything back, all they know relating to problems submitted to their examination. They put freely at the disposal of their respective commissions the material resources of their laboratories and factories, to assure a good ending to the technical researches proposed.

The schedule of projects laid out successfully have become extensive. The practical results are most substantial and seem to promise, for the future, extremely encouraging results. About forty research projects are now in progress. For each one, the aim to be reached is determined by the Technical Committee. A budget corresponding to an approximate cost is fixed and placed under a responsible engineer who has charge of the study. The results are published in the review *Fonderie*.

Technical Papers

However, a certain number of these problems require, for their solution, use of methods of investigation and experimental apparatus, whose scientific character is beyond normal ways of our foundries. A new knowledge may be reached capable of profitable application in foundries but might pass unnoticed by us, if no one has thought about establishing a "connection" between the scientific experts, rich in ideas and technical papers, and trained in experimental methods, and the technicians of our factories.

This is the part played by our scientific commission which brings us the assistance of scientists and metallurgists, under the chairmanship of M. Portevin. These persons assure the connection of French foundries with the laboratories of pure or applied science which they direct. The surveys are generally entrusted to young searchers, who work under the management and in the laboratory of one of the members of the commission, and who are financed by The Technical Center in the same way as they would be compensated in a factory.

Technical Correlation of the Foundry with Other Industries. We have just seen the astonishing consolidation that exists in the foundry industry, where numerous problems are discussed together. It was quite natural that the foundrymen have wished to introduce this same spirit of understanding in their relations with other industries, especially their customers.

Commissions Formed

To this end, mixed commissions with the customers' industries (motor car, machinery, and ship builders) were first formed where ideas and opinions, discussed with information based on the present methods, as adaptable to the foundry industry, were freely offered as possibilities to the technical needs expressed by these customer industries.

Later on, other mixed commissions, bringing together foundrymen and representatives of users of foundry products (steel makers, miners, manufacturers and foundry equipment men) were charged with the task of making specifications for the quality of raw materials and equipment made available to the foundrymen.

In all, seventeen technical commissions[1] are at this time in full activity and will continue to function as long as the problems they are charged to solve will not have been properly answered.

The activity of these commissions is lengthened by that of the Bureau of Standardization, associated on the other hand with the French Standardization Association, the only French organization authorized by the state to approve specifications.

Education Courses

Moreover, every time the commissions realize the need of strengthening, by lectures or education courses, the knowledge of the factory engineers, foremen, and workmen, the Technical Center takes charge. Its specialists frequently visit districts where certain number of foundries are grouped so that the local leaders

[1]Under the authority of the Technical Committee there are: (a) scientific commission, (b) technical commissions grouping foundry technicians, machinery and methods, metallurgy of gray iron, malleable iron, steel, copper alloys, aluminum and magnesium alloys, die-casting, pipes, domestic heat apparatus, laboratories, and (c) mixed commissions with other industries, such as heavy and medium-sized machinery, agricultural machinery, motor cars, ship-building, steel plants, collieries.

and foremen can benefit from their practical knowledge.

Diffusion of Technical Progress. To further the spread of all new ideas referring to the technique of foundry practice, orderly and striking pamphlets and technical publications are assembled at the Technical Center, together with information likely to improve the working methods applied in foundries, whether these data are included in the technical press, French or foreign, or collected at meetings.

All these documents are filed at the Library Service of the Technical Center where foundrymen can consult documents which are of any interest to them.

But to cover more thoroughly the whole trade, down to its most humble representatives, it was necessary to create an editorial office.

This important office of the Technical Center is mainly concerned in publishing the review, *Fonderie,* a joint magazine of the Foundry Technical Association and the Technical Center which has restored with increased resource the previous French foundry publication dealing with the foundry in France. The first number of *Fonderie* was issued in January, 1946.

It also publishes numerous pamphlets which, either give the results of studies carried out by the engineers of the Technical Center and by research workers financed by it after their work has been approved by the commissions or, published lectures presented to the Foundry Technical Association by authorities of the industry. It also issues practical handbooks for the use of technicians and foremen.

Finally, when the proposal for specifications submitted for the approval of commissions are ratified, it is the editorial office that undertakes to have them published.

Beginning this year, the Technical Center has taken charge of the direction of The Superior Foundry School, and, to assist financially The Technical Association. It is open to all who show initiative in developments, beginners or newcomers being quite sure of the moral support of the whole trade, and also of official organizations of the state.

Its success, which is certain, results without doubt from the need for progress displayed by the French foundry industry. It is certainly also due to the eagerness of its workers, those who belong to the staff, and those who devote most of their time to the work of commissions.

However, an extremely important part of this success must be attributed to its form of organization. It depends on the industry itself, supported by a control of the State which has taken the necessary steps to have it financed by industry (in addition to the budget and without expecting too much from voluntary contributions).

The board of administrators under the chairmanship of M. Ricard, in whom French foundrymen have great hopes, includes industrialists and technicians of all social levels, and scientists, who have the same faith in the future of the institution they administer.

Conclusion. In conclusion we wish to say this is how the Technique is organized in the French foundry industry. By a powerful concerted effort, we mean to raise it from its actual inferior state, in which war events have placed it.

The French Government has realized the importance of this question, and has brought us the inestimable help of an organization managed by industry itself, and allowed very important grants, which the State merely controls.

The French foundry industry, on its part, is quite determined to work with all its might to maintain the reputation, justly acquired during a past, full of fine technical achievments as testified by the awarding of an A.F.A. Whiting gold medal lately to one of its most courageous and sympathetic representatives, E. Ronceray, the gentleman who has created The Supreme Foundry Technical Association.

A new effort is undertaken, on ruins, but with faith that ennobles it and which, with the help of the confidence of French industry, will give us the patience to attain results at which we aim, and the reaching of which we have no doubts.

MEASUREMENT AND INVESTIGATION OF GASES ESCAPING FROM OIL CORES

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• **Official annual exchange paper from the Czechoslovak Foundrymen's Association presented at the Fiftieth Anniversary Convention of the American Foundrymen's Association, Sand Research Session, May 8, 1946, in Cleveland.**

PRODUCTION OF COMPLICATED aluminum alloy castings is closely connected with the use of oil cores. The hot metal at elevated temperatures causes decomposition of organic core substance and there is an evolution of gases, which under unfavorable conditions can impair quality of aluminum castings.

It is therefore advisable to develop a reliable method to measure gases evolved from cores and eventually to determine their composition. Methods suggested up to the present time were unsatisfactory for two reasons: (1) either they required the preparation of a special core, (by ramming the sand into a steel tube), or (2) very small samples weighing only some grams were used so that precision of the measurements was rather small.

As it will be shown later, the dead space of the apparatus plays an important part. For practical purposes it is expedient to use cores of dimensions usual in foundry work. When testing larger samples experimental conditions are more advantageous for securing precise results.

Apparatus and Procedure

The method suggested by the author (Fig. 1 and 2) used simple cores which can be prepared and baked under routine workshop conditions using cylindrical cores 32 mm. in diam. and 100-120 mm. in length, weighing 100-135 grams. Naturally also differently shaped cores can be used so far as they can be inserted in the apparatus.

The core is exposed to the high temperature without being in direct contact with the heated metal. The core as a whole is put into a steel tube *A*, Fig. 1, to be heated subsequently by a molten aluminum-silicon-alloy of 800-850° C. to a temperature near the pouring temperature of aluminum.

The aluminum alloy is cast around the steel tube into a block of 90 x 120 x 220 mm. which encloses the greatest part of the steel tube. To facilitate withdrawal of the block from the mold, starting from the parting line, the casting is tapered on both sides from 90 to 72 mm.

The casting has no gate and its purpose is only a rapid and uniform heating of the tested core under unvariable conditions. When the test is finished the casting is melted down in the furnace and the steel tube is ready for the next experiment (the casting weighs 5 kilograms).

The steel tube has a flange. Cap *B* is fastened by screws and provided with a thin tube to let gases escape and also less volatile products of decomposition. The cap of the reaction chamber is heated by heat conduction to such a degree that no condensation of water takes place upon it. For tightening the joint a packing of compressed cardboard was found very useful.

The greatest part of water and tar is condensed in a glass bulb *C*, 55 mm. in diameter and further in two glass traps *D*. The gas in a relatively pure state reaches therefore a gasometric bottle *E* of a volume approximately 10 *L*. The gas is kept over water. It is saturated with moisture and meets the requirements usual in gas analysis.

When heated, organic substances in the core decompose, the most rapid gas evolution taking place between the 2 and 5 min. whereupon the decomposition rate diminishes. Readings of gas volumes were always taken for first 30 min. The small gas quantities which later develop need not be considered.

The author describes apparatus and method suggested for measurement and sampling of gases driven off from oil cores at pouring temperature of aluminum. Accuracy of this method is verified on some core binders.

Precise Method

During the experiments a very simple and precise method for volume measurements of the escaping gases by means of the mentioned gasometric bottle was worked out. Parallel to the axis of the bottle is fixed and with the bottle content connected a glass tube of 10 mm. inside diameter, a real water pressure gauge of high precision.

The bottle is filled up to the stopper with water and the water level in the side-tube is adjusted to the same height. When collecting the gas in the bottle, water is run off until the water levels in the bottle and the side-tube equalize. Under these conditions the water volume measured in a calibrated cylinder corresponds to the gas volume under the atmospheric pressure.

By this simple and rapid procedure it is possible to take readings of the gas evolution; every volume determination requires only one reading and simultaneous volume

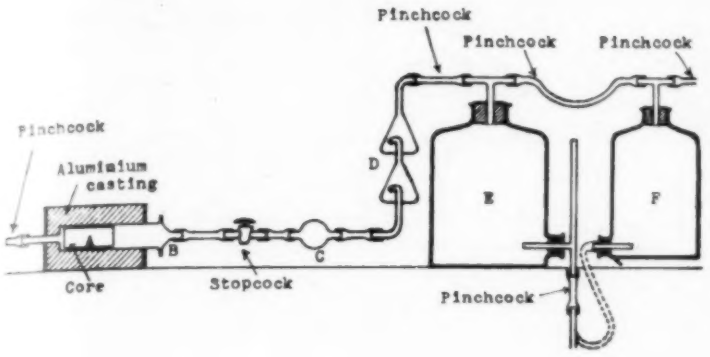


Fig. 1—Diagram shows equipment used in measuring and investigating gases escaping from oil cores; A—steel tube receptical for test core, B—cap fastened by screws, C—glass bulb into which water and tar is condensed, D—two glass traps, E—gasometric bottle, F—collecting bottle.

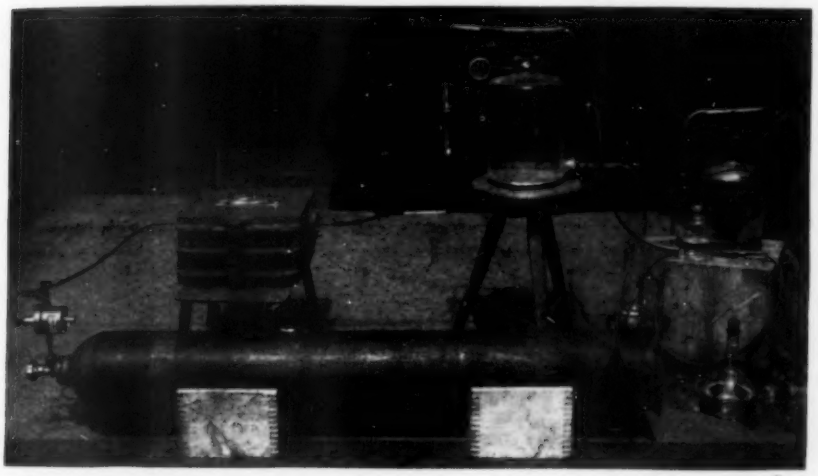


Fig. 2—Photograph shows arrangement of equipment used in measuring and investigating gases escaping from oil cores.

and pressure readings as usual in gasometric work, are not needed. The volume measurement does not depend on the shape of the bottle and the wall of the bottle need not be calibrated (internal diameter of the bottle being about 155 mm.).

The chemical composition of the escaping gases is subject to changes during the decomposition of the core. It is therefore necessary to mix the gas thoroughly, if an average sample is to be taken.

After 30 min. the gasometric bottle *E* is disconnected from the steel tube containing the core. Gas mixing is achieved by pumping the gas from the gasometric bottle *E* to a further collecting bottle *F* (volume 5 *L*) and vice versa (this procedure being repeated three times)

The necks of both bottles are connected by a flexible rubber tube, just as the tubular necks at the bottom, by lifting or lowering the bottle *F* (at the beginning full of water) the gas is pumped from one bottle to another.

From the smaller bottle *F* a gas sample can be taken into a gas pipette provided with two stopcocks. The technical gas analysis is carried out by the Orsat apparatus—when possible on the same day—using analytical routine methods. The ethylene amount was checked also by Haber's method (addition of bromine water).

When heated, the core is surrounded by air. The developing gases therefore mix with the air and then escape from the steel tube, but it is inevitable that they remain to

Table 1

HEATING OF CORES IN AIR AND NITROGEN

Ref. No.	Core Oil	Weight of Core, Grams	Volume, cc. of Escaping Gases	Volume, cc. of Gases from Core	Chemical Composition of Gases from the Core, Per Cent CO_2	CO	H_2	CH_4	C_2H_4	Volume of Gases from 100 Grams of Core, cc. Total Volume	CO_2	CO	H_2	CH_4	C_2H_4
1	Linseed oil in air	98.9	644	459	25.8	30.4	21.0	13.0	9.8	464	120	141	98	60	45
2		98.5	627	391	30.1	27.6	6.1	27.2	9.0	397	120	109	24	108	36
	Average				28.0	29.0	13.5	20.1	9.4	430	120	125	61	84	40
3	Linseed oil in nitrogen	96.4	727	438	22.9	30.2	16.3	15.0	15.6	454	104	137	74	68	71
4		91.8	609	360	22.5	30.5	21.5	16.4	9.1	392	88	119	85	64	36
	Average				22.7	30.4	18.9	15.7	12.3	423	96	128	79	66	54
5	Commercial core oil in air	96.9	753	543	28.1	27.5	22.5	12.3	9.6	560	157	155	126	69	53
6		95.3	652	464	31.2	26.3	15.6	19.5	7.4	487	152	128	76	95	36
	Average				29.7	26.9	19.0	15.9	8.5	524	155	142	101	82	44
7	Commercial core oil in nitrogen	98.1	704	475	27.8	26.1	10.2	26.8	9.1	484	135	126	49	130	44
8		97.9	695	434	28.2	26.9	8.5	25.3	11.1	443	125	119	38	112	49
	Average				28.0	26.5	9.4	26.0	10.1	464	130	123	43	121	47

Table 2

CHEMICAL CHARACTERISTICS OF OILS USED

Oil Type		Saponification No.	Acid No.	Iodine No. (Wijs)
Linseed-Oil				
	Series I.	190	5.5	179
	Series II.	185	2.4	148
Core Oil K-Group 1a				
	Series I.	184	13.9	94
	Series II.	183	16.9	64
Castor Oil		178.5	4.1	85

some extent in the tube. The author's experiments to avoid dead space were unsuccessful.

To remove the gases remaining in the steel tube after finishing the test, the same are carried away by pure nitrogen. It was found that for practical purposes a nitrogen volume equal to the gas volume is sufficient. The measurement of the nitrogen volume is accomplished also by allowing water to run from the gasometric bottle *E*.

The exchange of core-gases and nitrogen takes place in such a short time that no secondary reactions can be established to a greater extent. The above mentioned procedure of mixing the gas and taking the sample is naturally carried out just after admission of nitrogen.

According to the author's method of calculation it is supposed, that oxygen and nitrogen in these experiments originate from the air only and not from the decomposed organic substance. All other components are taken in consideration and in this way the influence of the air is eliminated.

Analytically determined percentages of carbon dioxide, carbon monoxide, methane, hydrogen and unsaturated hydrocarbons (ethylene) are expressed in cc. per 100 grams of dried core substance.

Decomposition of Cores in Nitrogen. A core ready for experiment can be slipped lightly in the steel tube (inside diameter 35 mm., length 235 mm.). Decomposition is then carried out in presence of air. It is possible to study decomposition of cores in an atmosphere of nitrogen in the same apparatus.

This being the case the steel tube is filled with nitrogen at the very beginning of the experiment. Table I gives the results of comparative tests obtained on cores bound by linseed-oil in air and nitrogen respectively. The cores were baked together with other cores, all at the same time in a workshop drying oven.

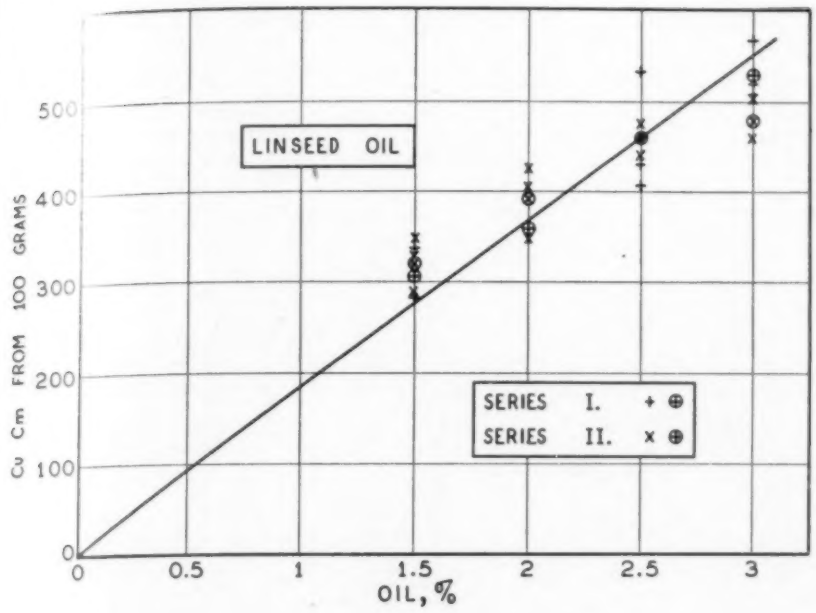
Volumes in Table I were reduced to 0° C., and the pressure of 760 mm. of mercury. The total gas volume (expressed in cc. for 100 grams of core, Column 11, Table I) does not change in nitrogen. In the chemical composition a small shifting takes place, principally in disfavor of carbon dioxide.

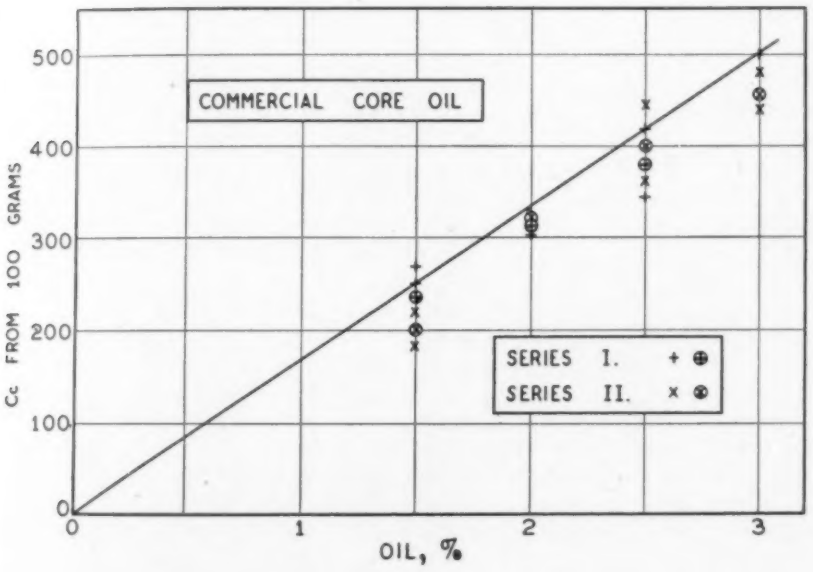
During the tests of some other core-oils the fact was verified, that the total quantity of gases very little changes when the core decomposition takes place in nitrogen instead of air. It is to be stated that in

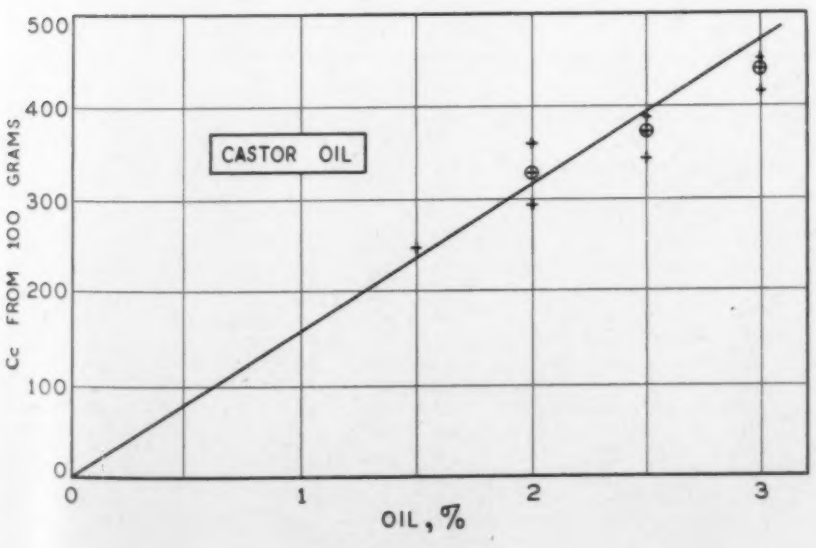
Table 3

LINSEED OIL DATA

Ref. No.	Oil, Per Cent	Weight of Core, Grams	Volume, cc. of Escaping Gases	Volume, cc. of Gases from Core	Chemical Composition of Gases from the Core, Per Cent CO_2	CO	H_2	CH_4	C_2H_4	Volume of Gases from 100 Grams of Core, cc. Total Volume	CO_2	CO	H_2	CH_4	C_2H_4
					FIRST SERIES										
9	3	131.2	849	666	31.4	24.5	9.2	24.0	10.9	508	160	124	47	122	55
10		132.1	826	689	35.0	21.1	8.6	25.7	9.6	521	182	110	45	134	50
11		136.8	910	776	30.0	25.4	4.7	28.2	11.7	566	170	143	27	160	66
	Average				32.1	23.7	7.5	26.0	10.7	532	171	126	40	139	57
12	2.5	130.2	740	531	32.4	28.5	10.0	19.6	9.5	408	132	116	41	80	39
13		133.8	870	719	32.4	28.1	10.2	19.4	9.9	537	174	151	55	104	53
14		132.2	700	570	35.8	24.6	10.3	20.2	9.1	430	154	106	44	87	39
	Average				33.5	27.1	10.2	19.7	9.5	458	153	124	47	90	44
15	2	136.7	646	488	34.9	26.8	7.9	22.5	7.9	357	125	96	28	80	28
16		137.5	632	505	37.6	23.1	10.0	21.8	7.5	367	138	85	37	80	27
	Average				36.2	25.0	9.0	22.1	7.7	362	131	91	32	80	28
17	1.5	135.4	479	346	38.2	23.3	15.0	18.0	5.5	255	98	59	38	46	14
18		134.4	538	417	37.7	23.2	12.7	20.4	6.0	310	117	72	40	63	18
	Average				38.0	23.2	13.9	20.1	5.8	283	108	65	39	55	16
					SECOND SERIES										
19	3	136.3	842	628	41.2	22.5	2.9	27.4	5.9	461	191	104	13	126	27
20		138.3	790	697	39.8	20.1	4.5	28.3	7.2	503	200	101	23	143	36
	Average				40.5	21.3	3.7	27.9	6.6	482	195	103	18	135	31
21	2.5	135.1	751	646	33.9	22.3	7.0	27.7	9.1	477	162	107	33	132	43
22		139.0	730	614	31.8	24.5	9.3	25.4	9.0	441	140	108	41	112	40
	Average				32.8	23.4	8.2	26.5	9.1	459	151	108	37	122	41
23	2	136.7	673	484	38.6	22.2	6.7	25.8	6.7	354	137	78	24	91	24
24		137.4	684	554	37.3	23.2	6.9	25.7	6.9	403	150	94	28	103	28
25		137.1	717	584	31.0	24.3	7.6	27.5	9.6	425	132	103	32	117	41
	Average				35.6	23.2	7.1	26.4	7.7	394	140	92	28	103	31
26	1.5	136.2	561	360	40.8	19.0	1.6	32.1	6.5	264	108	50	4	85	17
27		136.4	532	407	39.7	19.8	5.5	28.0	7.0	298	118	59	16	84	21
28		136.8	578	442	37.2	19.1	5.0	30.1	8.6	323	120	62	16	97	28
	Average				39.2	19.3	4.0	30.1	7.4	295	115	57	12	89	22







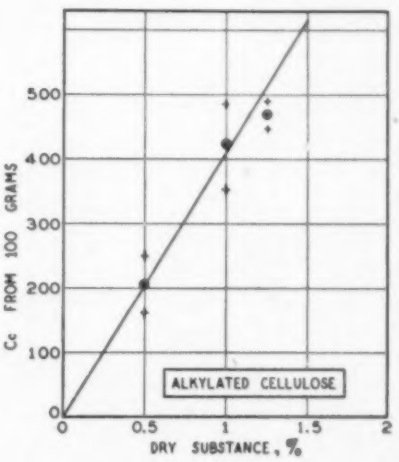


Fig. 3 (Left, above)—Curve shows cc. of gases escaping from 100 grams of test core using linseed oil binder. Fig. 4 (Left, center)—Curve shows cc. of gases escaping from 100 grams of test core using commercial core oil binder. Fig. 5 (Left, bottom)—Curve shows cc. of gases escaping from 100 grams of test core using castor oil binder. Fig. 6 (Above)—Curve shows cc. of gases escaping from 100 grams of test core using alkylated cellulose binder.

presence of air only small amounts of oxygen are taken in consideration. The reaction space has a volume of 225 cc., the volume of a core is 80-100 cc., thus for the dead space remain 125-145 cc., that is about 1 cc. per 1 gram of the core substance. The greatest differences were found for a core-oil of commercial quality (tests Ref. No. 5-8, Table I). Generally the total gas quantity in nitrogen is rather lower (for above-mentioned oil by 12 per cent).

Verifying the Method

During more than two years' testing practice on cores the author came to conviction that the accuracy of the method can be checked only on cores specially prepared for this purpose which contain known amounts of core-oil.

As core sand, pure silica sand was used; this was washed on a sieve with water to remove clayey matter and a sand fraction under 0.088 mm. The remainder on the sieve contained the following grain sizes:

Grain Size	*Per cent*
0.088 to 0.2 mm	31.9
0.02 to 0.3 mm	26.1
Over 0.3 mm.	42.0

Table 4

COMMERCIAL CORE OIL (K-GROUP 1A)

Ref. No.	Oil, Per Cent	Weight of Core, Grams	Volume, cc. of Escaping Gases	Volume, cc. of Gases from Core	Chemical Composition of Gases from the Core, Per Cent CO_2	CO	H_2	CH_4	C_2H_4	Volume of Gases from 100 Grams of Core, cc. Total Volume	CO_2	CO	H_2	CH_4	C_2H_4
					FIRST SERIES										
29	3	130.3	862	654	25.9	26.4	9.4	24.9	13.4	501	130	132	47	125	67
30		136.3	1076	918	24.3	27.1	9.6	24.5	14.5	673	163	182	65	165	98
	Average				25.1	26.7	9.5	24.7	14.0	587	147	157	56	145	82
31	2.5	131.0	589	448	42.6	23.7	11.6	16.9	5.2	342	145	81	40	58	18
32		133.4	676	557	38.8	23.8	9.7	19.9	7.8	417	162	99	40	83	33
	Average				40.7	23.8	10.6	18.4	6.5	380	154	90	40	71	25
33	2	136.5	538	442	37.5	27.7	10.5	19.0	5.3	324	122	89	34	62	17
34		133.7	538	408	42.2	22.2	10.0	19.5	6.1	305	128	68	31	60	18
	Average				39.8	25.0	10.3	19.2	5.7	315	125	79	32	61	18
35	1.5	133.3	422	255	40.4	23.2	11.2	19.2	6.0	191	77	44	21	37	12
36		128.6	479	324	37.2	24.8	15.3	16.5	6.2	252	94	62	38	42	16
37		131.5	492	353	35.6	22.2	17.2	19.4	5.6	268	95	60	46	52	15
	Average				37.7	23.4	14.6	18.4	5.9	237	89	55	35	44	14
					SECOND SERIES										
38	3	135.5	871	652	40.9	24.4	11.0	17.9	5.8	481	197	117	53	86	28
39															
40		134.5	738	595	39.3	24.2	12.1	18.2	6.2	442	174	107	54	80	27
	Average				40.1	24.3	11.5	18.1	6.0	462	186	112	53	83	28
41	2.5	135.7	565	493	34.4	26.2	10.1	21.7	7.6	363	125	95	37	79	27
42		135.3	744	606	34.4	23.6	11.5	23.6	6.9	448	154	106	51	106	31
	Average				34.4	24.9	10.8	22.7	7.2	406	140	100	44	93	29
43	2	135.8	573	445	38.9	20.4	12.1	22.2	6.4	327	127	67	39	73	21
44		135.3	573	443	37.2	21.7	14.2	19.9	7.0	327	122	71	46	65	23
	Average				38.0	21.1	13.2	21.0	6.7	327	125	69	42	69	22
45	1.5	131.4	425	242	46.2	26.7	11.1	13.9	2.1	184	85	49	20	26	4
46		132.6	438	282	41.7	24.3	12.1	15.2	6.7	217	91	53	26	33	14
	Average				44.0	25.5	11.6	14.5	4.4	201	88	51	23	30	9

Table 5

CASTOR OIL DATA

Ref. No.	Oil, Per Cent	Weight of Core, Grams	Volume, cc. of Escaping Gases	Volume, cc. of Gases from Core	Chemical Composition of Gases from the Core, Per Cent CO_2	CO	H_2	CH_4	C_2H_4	Volume of Gases from 100 Grams of Core, cc. Total Volume	CO_2	CO	H_2	CH_4	C_2H_4
47	3	137.9	705	575	43.1	18.3	11.4	20.1	7.1	417	180	76	48	84	29
48		134.1	713	601	39.5	21.8	11.4	20.4	6.9	447	177	97	51	91	31
49		131.7	743	596	36.9	21.5	12.7	20.4	8.5	452	167	97	57	92	39
	Average				39.9	20.5	11.8	20.3	7.5	439	175	90	52	89	33
50	2.5	137.9	634	475	42.7	20.5	12.8	17.1	6.9	345	147	71	44	59	24
51		137.0	651	498	36.7	20.3	11.6	22.5	8.9	363	133	74	42	82	32
52		137.5	678	570	37.2	20.9	12.4	21.2	8.3	415	155	87	51	88	34
	Average				38.9	20.6	12.3	20.2	8.0	374	145	77	46	76	30
53	2	137.2	578	406	44.8	20.7	13.5	15.5	5.5	296	133	61	40	46	16
54		137.8	618	498	37.2	24.9	13.4	18.1	6.4	361	134	90	49	65	23
	Average				41.0	22.8	13.5	16.7	6.0	329	134	75	45	55	20
55	1.5	133.9	482	332	44.5	18.0	15.1	16.9	5.5	248	110	45	37	42	14

Table 6

ALKYLATED CELLULOSE DATA

Ref. No.	Percentage of Jelly	Percentage of Dry Binder	Weight of Core, Grams	Volume, cc. of Escaping Gases	Volume, cc. of Gases from Core	Chemical Composition of Gases from the Core, Per Cent CO_2	CO	H_2	CH_4	C_2H_4	Volume of Gases from 100 Grams of Core, cc. Total Volume	CO_2	CO	H_2	CH_4	C_2H_4
56	10	1.25	132.0	849	583	19.8	40.5	14.0	21.9	3.8	440	87	178	62	96	17
57			131.6	858	614	17.5	41.3	14.9	24.0	2.3	467	82	193	69	112	11
58			134.3	922	655	18.7	41.8	12.4	23.6	3.5	487	91	204	60	115	17
	Average					18.7	41.2	13.8	23.2	3.1	465	87	192	64	108	15
59	8	1	130.6	668	460	18.8	41.4	15.5	23.4	0.9	352	66	146	55	82	3
60			132.1	741	623	13.8	39.7	14.8	28.1	3.6	487	67	194	72	136	18
	Average					16.3	40.6	15.1	25.8	2.2	420	67	170	64	109	10
61	4	0.5	132.4	464	327	21.8	32.4	17.8	24.6	3.4	247	54	80	44	61	8
62			132.8	339	213	19.6	34.0	17.0	26.8	2.6	160	31	55	27	43	4
	Average					20.7	33.2	17.4	25.7	3.0	203	43	67	35	52	6

Table 7

VOLUME OF GAS IN *CC.* FROM 1 GRAM OF CORE BINDER

	At 0° C., 760 mm. of Mercury	*Under casting Conditons 750° C., 730 mm. of Mercury*
Linseed-Oil	180	700
Core Oil of commercial quality	165	645
Castor Oil	160	625
Alkylated cellulose	400	1560

As core-oil, we used linseed-oil which was thoroughly mixed with the silica sand; no further admixtures were given in the cores. The baking was carried out in a drying-oven for 3 hr. at 200-210° C. Baked cores were uniform on the fracture. When heated to temperature of molten aluminum they darkened and very easily distintegrated.

Experiments on linseed-oil were undertaken in two separate series with oil from two different shipments. Further a core oil which was at disposal on commercial scale in the war time, was tested. During the war core binders were graded in some groups (Regulation No. 43, November 30, 1942, Reichsstelle Chemie, Berlin) and in occupied Europe could be sold only in brands specified in this Regulation.

Core Binders Used

The tested core oil complied to K-Group 1a. Also in this case two series were carried out with oil from different shipments. At last, for comparison of linseed oil with a not-drying oil, for our experiments castor-oil was chosen in the same quantity as linseed-oil. See Table 2.

In addition to oils also alkylated cellulose, suggested as a core binder was examined. In this case the handling was rather different. Alkylated cellulose is used on commercial scale in form of fine flakes which, stirred in water, produce a stiff jelly (1 part of alkylated cellulose, 7 parts of water, by weight).

This jelly can be mixed easily with core sand. The test cores were prepared from the silica sand and 4, 8, 10 per cent respectively of the jelly. They require to be baked at substantially lower temperatures (160° C., 3 hr.) and are not as darkened as oil cores.

Results of all comparative tests that were carried out, were summarized in Tables 3 to 6 and plotted in Fig. 3 to 6. All volumes are reduced to standard conditions (0° C., 760 mm. of mercury).

It is to be noted that the chemical decomposition of cores is very complicated nature and depends on many factors which cannot be completely controlled. Some tests were carried out always under the same conditions and the average values show a close agreement. For oil, where two series of cores were under test, there are only small differences in various shipments of the same core oil.

Gas Quantities

According to Fig. 3 to 6 there is linear relation connecting the quantity of the core binder with the volume of gases driven out of the core. Under usual casting conditions (750° C. and 730 mm. of mercury) the gas volume is about 3.9 times greater than at room temperature.

Unexpected is the fact that all tested oils show only small differences in gas quantities (the whole range of mean values: 15 per cent). From 1 gram of alkylated cellulose more than the double amount of gases is driven off. It should be kept in mind that owing to its great bond alkylated cellulose is used in smaller percentage than core oils.

In this way the disadvantage of a great volume of developed gases is compensated. Of some interest is also the average composition of gases emanating from individual core binders.

Gases not condensing at laboratory temperature are considered in the chemical analysis. By analysis water and tar formed in decomposition of cores are not determined.

Simplified Testing

If it is not necessary to know the chemical composition of gases, the method can be simplified for routine testing in such a way that the apparatus is calibrated in precise tests and its correction is determined. This procedure means a substantial speeding up of the work. It is used only to measure the volume of escaping gases without admission of nitrogen to the volume reduced to standard conditions. The result is then expressed in cc. per 100 grams of core substance. The calibration of the apparatus was carried out for different oils, considering the tests summarized in Table 3 and 4 and a good agreement was reached. In following Table 9, in the last column, mean deviation of a single observation concerning this simplified method is given.

Comparison with Results of Other Investigators. A. Riebold and Th. Prinz, *Die Giesserei,* Sept. 6, 1929, p. 820, also H. Ries, "Some Foreign Methods of Testing Foundry Sands," *Transactions,* American Foundrymen's Association, Vol. XL, p. 355, (1932) discuss a method similar to that suggested by the author. The core was rammed into a steel tube closed at one end. The gases were measured by means of a rather complicated apparatus. The decomposition of the cores took place at 1300° C. at temperatures which are not considered in aluminum foundry

Table 8

MEAN CHEMICAL COMPOSITION OF GASES DRIVEN OFF FROM CORES IN PER CENT

	Carbon Dioxide	*Carbon Monoxide*	*Hydrogen*	*Methane*	*Ethylene*
Linseed-Oil					
I. Series	34.9	24.7	10.0	22.0	8.4
II. Series	37.0	21.8	5.8	27.7	7.7
Mean Values	36.0	23.2	7.9	24.9	8.0
Core oil of commercial quality					
I. Series	35.8	24.7	11.3	20.2	8.0
II. Series	39.1	24.0	11.8	19.1	6.1
Mean Values	37.5	24.3	11.5	19.7	7.0
Castor Oil	41.1	20.5	13.2	18.5	6.8
Alkylated Cellulose	18.6	38.3	15.4	24.9	2.8

work. The author is not aware if this method was later used.

The question of gases escaping from cores when pouring was treated by Dr. Klaus Grassmann of Hildesheim (Germany). His method was introduced in some German aluminum foundries as a routine method of testing.

Dr. Grassmann describes (*Giesserei*, vol. 29, p. 181-185, 1942) his own method as follows: "In a tube furnace kept without interruption at 700° C. a porcelain boat with a certain quantity of the core material is introduced (1-5 grams according to the expected gas amount). In some seconds a quantity of gas developed can be collected over water or mercury and measured."

This description of the apparatus gives only little information. Further particulars are stated in German specifications for testing of core sand, core binders and cores (*Giesserei*, vol. 30, p. 276-278, 1943, and vol. 31, p. 13-16, 1944).

In this case Grassmann's method is described as a standard method for testing of material. Five grams of core substance are taken, the sample being burned at 700° C.

The apparatus consists of a resistance furnace of Messrs. Heracus, Hanau (length 600 mm.), with ineter 30 mm. and length 1000 mm.). The temperature is controlled by a Hoskins thermopile operating a current regulating device. Dimensions serted porcelain tube (inside diamof the porcelain boat: 100 x 12 x 10 mm.

Dr. Grassmann investigated also cores bound by 2 per cent linseed-oil baked at different temperatures. For a sample dried at 210° C. 3 hr. he found the gas quantity 1400 cc. per 100 grams of core substance (loc. cit. p. 184, Fig. 2), that is a value four times greater than the one found by the author's apparatus.

Grassmann's method was checked on cores of Series II, containing 2 per cent of linseed-oil. A shorter tube was used (length 350 mm., inside diameter 24 mm.). The results depend to a great deal on the weight of the specimens, the other conditions being unchanged.

The values in Table 10 agree quite well with the measurements mentioned by Dr. Grassmann. The difference of both methods is based on their principles. In Grassmann's method a mixture of gases and water vapor is formed in the tube. The corresponding volume of air is driven off and measured.

In the author's method the effect of air is eliminated. Water is condensed so that only components which remain in gaseous state at the temperature of the laboratory are measured.

From these points of view the method of Dr. Grassmann needs a further elucidation of principles it is based upon.

Table 9

CORRECTION OF THE APPARATUS

Oil	*Correction of the Apparatus in cc.*	*Mean Error of a Single Measurement, in cc.*
Linseed I. Series	148	±28
II. Series	144	±42
Core oil of commercial quality		
I. Series	146	±32
II. Series	149	±43
Most probable value of the correction	147	

Table 10

DETERMINATION OF GAS QUANTITIES BY GRASSMAN'S METHOD

Weight of the Sample, in grams	*Weight of Gas at end of 5 min. cc. per 100 grams of core substance*
2	16.4
3	15.2
4	13.6
5	12.8

Temperature 25° C., pressure 730 mm. of mercury.
Gas volumes were not reduced to standard conditions.

Conclusions

The author describes a dependable method for determining the quantity and chemical composition of gases escaping from oil cores at the aluminum pouring temperature. The results of tests can be summarized as follows:

1. The gas quantities developed from different oils are in a very close range; for this reason for commercial core oils no great differences are to be expected.

2. The gas amounts set free from alkylated cellulose are much higher.

3. The gas development from cores of various chemical composition cannot therefore be estimated with a reasonable accuracy from their loss of weight on ignition.

Acknowledgment

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MAGNESIUM ALLOY CASTINGS

Relative Significance of Design and Metallurgical Factors to Serviceability

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MODERN DEMANDS for efficiently designed, light-weight parts for mobile or reciprocating structures has prompted the extensive use of magnesium alloys for such purposes. The uses to which these structural parts are put almost invariably involve cyclically alternating stresses which are impressed on these parts throughout their normal service life. For this reason, the service performance of light metal castings is frequently governed by the fatigue properties of the castings.

Stress Reduction

In parts where incidental parasitic or sympathetic vibrations exist which lead to fatgue failure, it often is possible to improve service life by reducing the magnitude of cyclic loading. Frequently, this is accomplished by stiffening various structural parts with brackets and stiffeners or by the use of generally more rigid structures.

In cases where the intended functions of the parts involve working stresses of the cyclic type, such as in aircraft wheels, fans, propellers, and certain engine castings, it often is impractical to consider reducing the magnitude of these cyclic loadings.

For this reason, an approach to the problem of improving service performance from the aspect of increasing the resistance of the parts to fatigue failure, rather than from the aspect of reducing the amount of such stressing, greatly assists the application engineer.

A study of the criteria and practices used for selecting and using materials for structural applications, as well as a study of the parts themselves and their service performance, shows that several factors contribute to precipitating failure by fatigue in light metal parts.

Structures of the type made of magnesium or aluminum alloys usually are intended to operate at high structural efficiency. This means that all of the material in the part would be contributing efficiently to the load-carrying ability of the structure.

When such is the case, the working stresses usually are considerable throughout much of the structure, especially with respect to the fatigue strength of the structural material.

Major attention in selecting materials and in design usually is based upon the static strength properties of materials and not as often upon their fatigue properties. Reliable information for the latter has not been as generally available for the the static properties.

This difficulty is being relieved by the recent additions of information from laboratory fatigue tests on magnesium alloys*. These additions include advice regarding the most effective use of these data.

Design Aspects

Since the nominal strength of parts usually is dependent upon their shape and dimensions, and since alternating stresses are less amenable to prediction, stress calculations for parts before construction often are impossible or inaccurate. This has seriously hampered the designing of parts for satisfactory service life.

However, recent advances in experimental stress measurement tech-

Presented at an Aluminum and Magnesium Session of the Fiftieth Annual Meeting, American Foundrymen's Association, at Cleveland, May 7, 1946.

*G. H. Found, "The Notch Sensitivity in Fatigue Loading of Some Magnesium Base and Aluminum Base Alloys." Submitted for publication in ASTM.

The first in a series of papers sponsored by a new subcommittee of the Aluminum and Magnesium Division of A.F.A. This committee is entitled "Design and Stress Measurement in Magnesium and Aluminum Castings" with Charles E. Nelson, Dow Chemical Co., Midland, Mich., as chairman. It is expected that later papers will deal specifically with such direct subjects as the effects of certain mechanical finishes on castings and how they may be obtained; on methods of predicting and measuring stresses on experimental and production castings and how to design for most efficient use of metal in such parts. The general aim is to present actual worthwhile information on what factors are really important to serviceability of castings in such a manner that engineers and foundrymen can meet on a common ground of understanding.

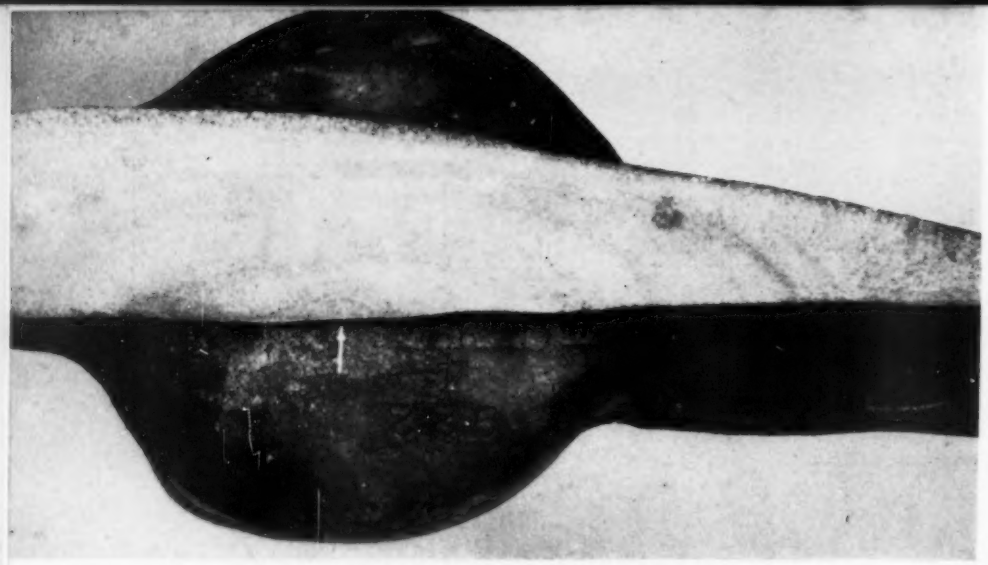


Fig. 1—Interface of a fatigue failure at a critical location in a magnesium air-conditioning fan of obsolete design. Note typical point origin of failure at surface (at arrow) and evidence of progressive encroachment of fissure into metal. Bending stresses predominated.



Fig. 2—Interface of a fatigue failure in a critical region of a magnesium landing wheel. Note same characteristics of failure as in Fig. 1. Bending stress predominated.

niques with SR-4 electric strain gauges and brittle lacquers have provided the designer with a tool that will enable him to achieve uniquely efficient casting design and, at the same time, satisfactory service life.

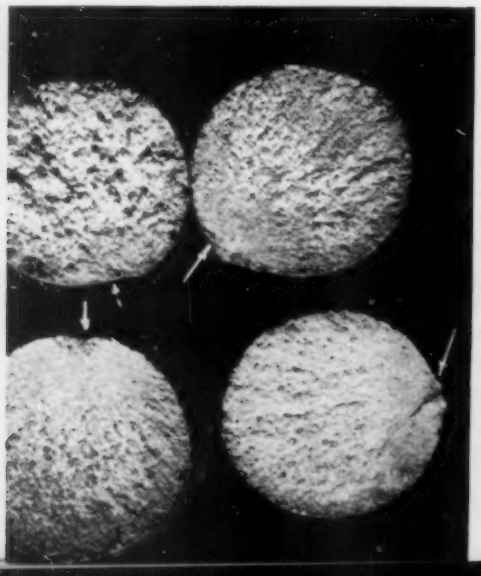
Studies of service failures and service parts have demonstrated that a better appreciation for the factors that are of fundamental importance in fatigue phenomena is necessary. A clearer understanding of the relative importance of internal metallurgy, mechanical surface condition and design on the resistance to fatigue failures must be obtained.

It is evident from this conclusion that an effort must be made to better acquaint design engineers, foundrymen, quality control men and other metallurgists alike with the factors affecting fatigue properties and serviceability which have been learned from service and service-simulating experiences.

It is the purpose of this paper to assist in remedying this last factor by arriving at some working ideas about fatigue behavior from fatigue experiences gained to date. Several interesting relations found to exist between service failures and defects

Fig. 3 (below, left)—Interface of fatigue failures induced in laboratory test bars in axial loading fatigue tests. Note that points of origin of failures are at surface. Uniform stresses across cross section existed. 2X.

Fig. 4 (below)—Interfaces of fatigue failures induced in aircraft motor castings show inception of failure at surface at foundry defects extending to the surface. Fatigue portion of failure is localized around these defects (at arrows). Uniform stresses across cross section existed.





in metallurgy and design are brought to attention in this discussion.

Evidence for the Nature of Fatigue Failures: Importance of Points of Weakness and Surfaces. The static strength of cross sections upon which loads are imposed may be evaluated in terms of the aggregate strength of the entire cross section. Thus, if any point of weakness is observed within a cross section, the static strength of the cross section is automatically thought of in terms of the net load-carrying ability of the remaining non-defective cross section.

Fatigue Resistance

Unfortunately, this and other simple conceptions of static strength do not apply as the basis for thought about fatigue problems. In other words, the strength to resist failure of a metal area under alternating stress application should not be rationalized on the basis of static strength criteria.

There has been a tendency to focus attention upon factors of relatively minor importance such as grain size, internal porosity, and other metallurgical factors rather than to consider some of the factors such as designed-in stress concentrations, surface quality, and the absolute location of porosity and defects which have a much more important bearing upon the fatigue properties or serviceability of the casting.

Shown in Figs. 1 and 2 are sections through fatigue failures in two large magnesium castings in locations where bending stresses pre-

Fig. 5—Fatigue failures in cast magnesium test panels showing failures originating at defects (at arrows) which extend to the surface. 2X.

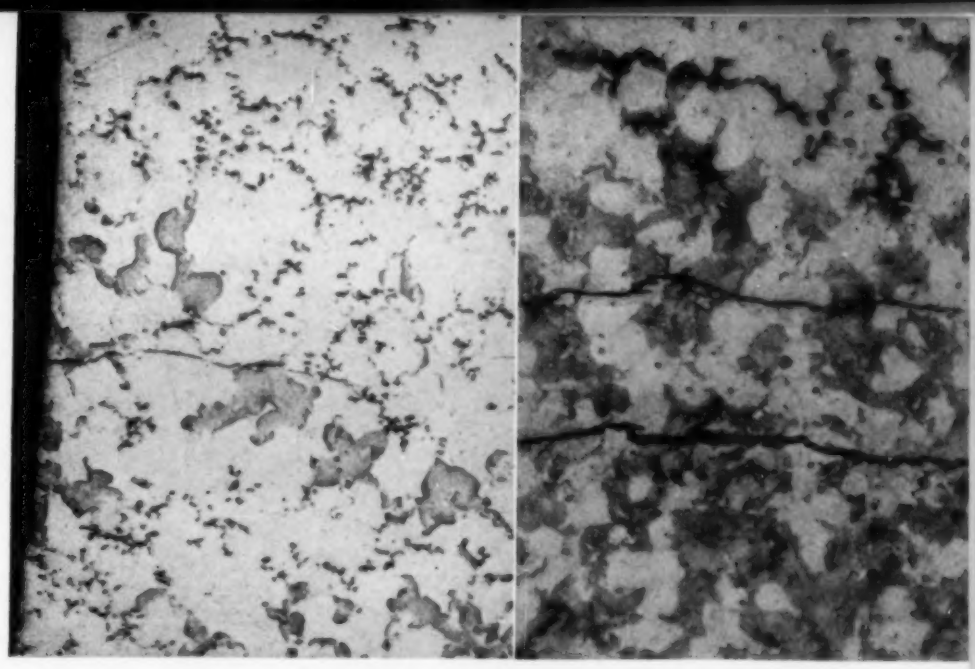


Fig. 7 (left)—Microsection of a fatigue failure induced in a part showing the relative independence between the course of the failure and the location of concentrated porosity. 50X. (Right)—Fatigue failures through a part showing independence between course of failure and proximity of nearby porosity. 75X.

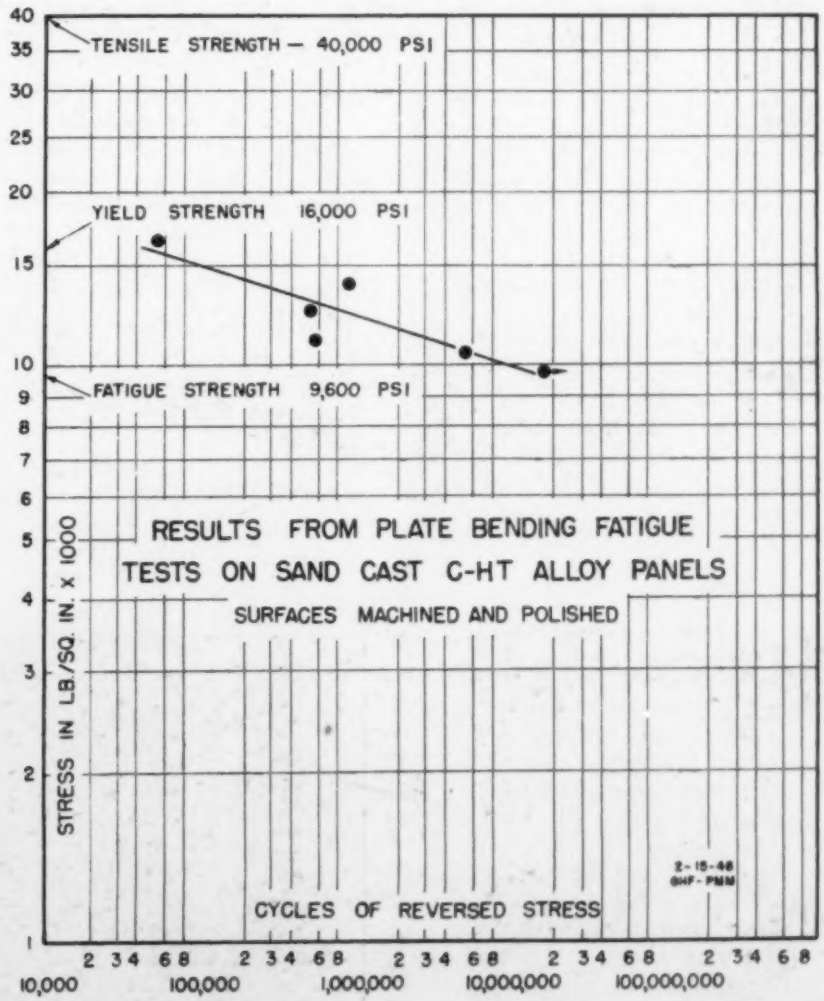


Fig. 6—Typical fatigue curve showing fatigue properties of cast panels and the static properties of the same material.

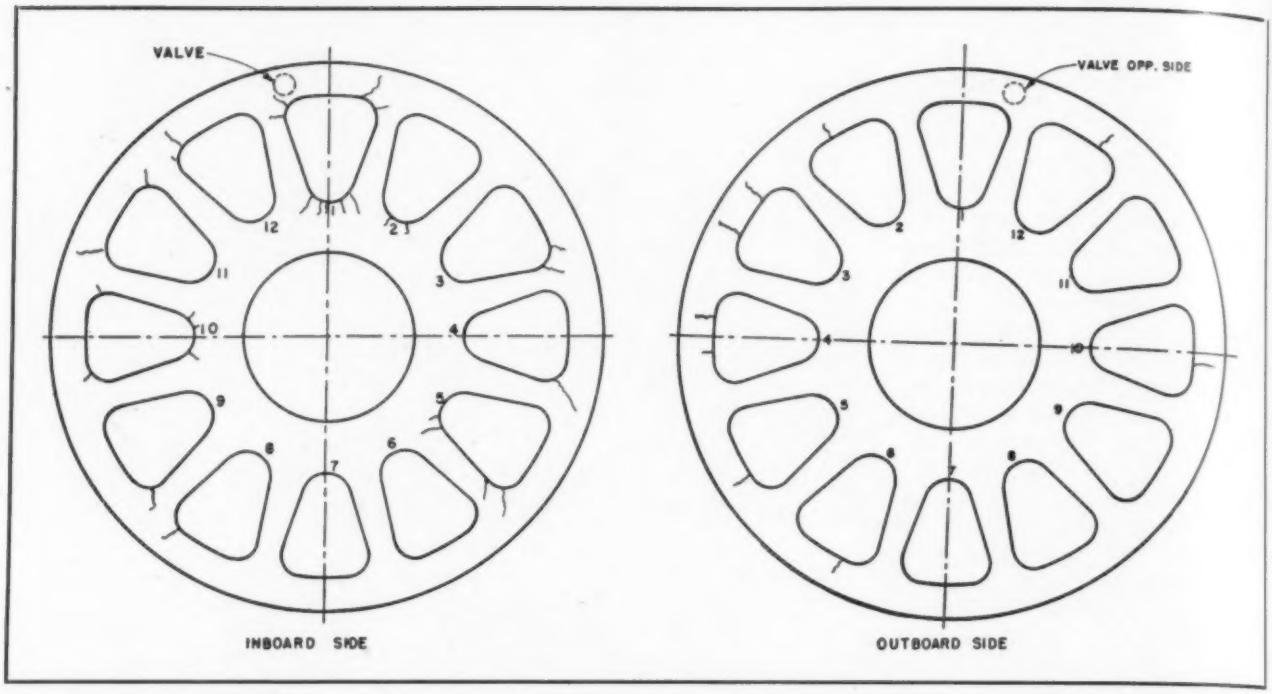


Fig. 8—Diagrammatic views of inboard and outboard sides of an aircraft wheel showing location of roll test failures.

dominated. Failure occurred at points of maximum tensile strength.

In Figs. 3 and 4 are fatigue failures induced in the laboratory in test bars and in a casting, respectively, by axial stresses, that is, stresses which are uniform across the cross section.

The general appearance of these failures shown in Figs. 1 to 4 is typical of all fatigue failures in cast magnesium. Fatigue failures (except in special cases where they are influenced by artificially induced, high surface stresses) always originate at the surface of the metal, whether stressing is due to bending or axial loading.

Failure Characteristics

Failures progress inward leaving interfaces marked with rings or bands concentrically arrayed about the point of inception of the failure. In Figs. 4 and 5 it will be noted that failure by fatigue originated at visible local defects which extended to the surface layer.

It is evident from these illustrations that fatigue failures always start at points in the surface of the metal and propagate inward with each successive cycle of stress. Fatigue strength of the metal is thus not as much a function of the aggregate net strength of the non-defective section as it is a function of the strength of any defective area. This can be thought of as a reason why the fatigue strength of materials is considerably below their static strength, as shown for a given cast magnesium alloy in Fig. 6.

Relative Importance of the Subsurface Metallurgy. Since the importance of the condition of the surface layers has now been emphasized, it is interesting to investigate what influence the condition of the metal below the surface may have on the serviceability of the metal.

Examinations of many service parts have shown repeatedly that failures may occur in uniformly stressed regions through sound areas even as close as a small fraction of an inch removed from areas containing subsurface microporosity in various amounts.

Figure 7 is a microsection through such a failure showing a fatigue crack which originated at a point only slightly removed from a region

Fig. 9—Sectioned out region of wheel in Fig. 8 showing fatigue failure interface and subsurface porosity. 6X.



of concentrated microspority located just below the surface.

Laboratory fatigue tests of the cantilever bending type of cast magnesium panels containing microporosity show similar results. With a few random exceptions, if failure does occur through microporosity or a defective region, such regions extend to the surface layer of the panels, as already shown in Fig. 5.

In Fig. 4 the same effect has been shown in fatigue tests on aircraft motor castings. The failures started at foundry defects at the surface of these particular parts in Fig. 4. Failures in other similar parts of this design during laboratory tests often occurred through sound metal adjacent to subsurface porous regions.

General Indications

The observations just discussed are not intended to obscure the fact that failures have occasionally been observed through subsurface porosity in uniformly stressed regions in test panels and parts.

However, the frequency of these occurrences is not only random with respect to any measurable variable in the porosity within the tolerable limits of porosity, but also is small when compared to the number of instances when failures have occurred at either defective surface regions or regions metallurgically perfect adjacent to internal defects.

If there were no instances of even random failure through subsurface porosity, this would indicate that subsurface porosity strengthens the cross section, which, of course, is not so.

Relative Importance of Design and Metallurgy. So far, the discussion has been concerned only with regions of cast metal which have been uniformly stressed. Failures which occur in uniform sections of service parts where design complications are not involved would approximate these conditions.

However, critical regions in service castings usually are at locations of complex design where non-uniform stressing, or stress concentrations, exist. Accordingly, it is well to consider applications which involve non-uniform stressing with regard to the influence of surface effects, internal porosity and design.

An ideal example for the discussion of this subject is a certain cast magnesium aircraft landing wheel which, in its early design stage, was roll tested to failure. Failures occurred in the wheel short of the desired life. The locations of the failures are shown in the diagrammatic views of this wheel in Fig. 8.

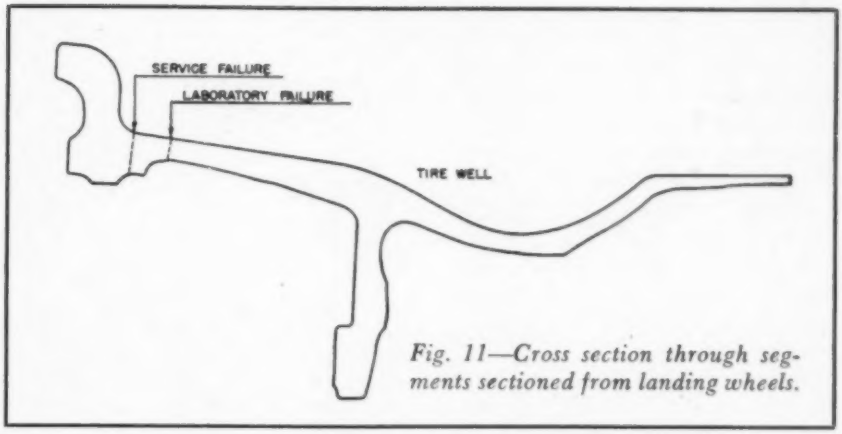


Fig. 11—Cross section through segments sectioned from landing wheels.

Observations

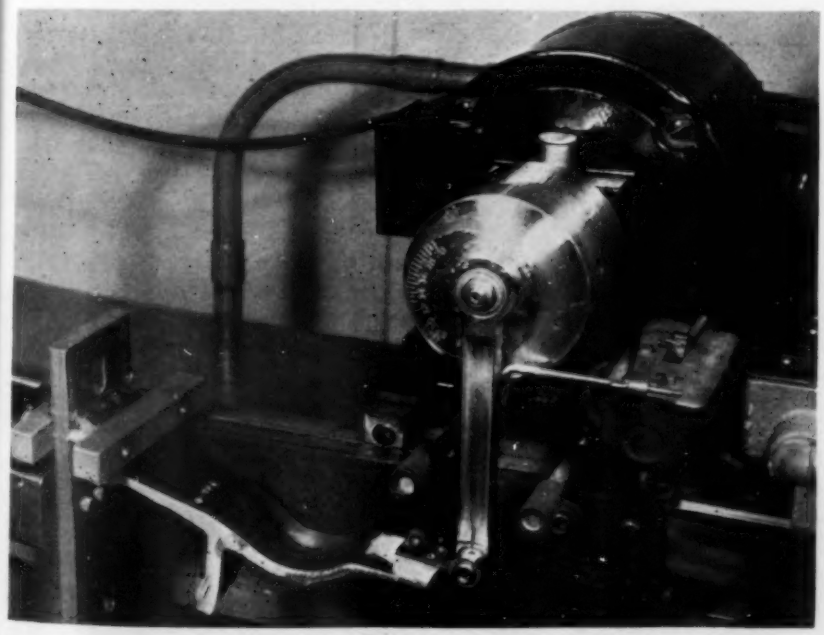
It will be noted that a relationship exists between the location of the failures and the design features on each side of the wheel. Failures were confined to the fillet regions around the spokes on the inboard side of the wheel, while they tended to be between the fillets on the peripheral side of the spoke windows on the outboard side of the wheel.

Radiographic examinations revealed that microporosity existed in the wheel in several areas in amounts which approach unacceptability. This was an early model of the wheel. The foundry, therefore, had not cast a sufficient number to completely solve its porosity problem. In a few instances the roll test failures protruded into regions of this porosity, as shown in the photomicrograph in Fig. 9.

In this figure the failure interface is the upper edge of the metal. The fatigue portion (straight portion) of the failure originated in sound metal at the surface on the right side and extended into the area of porosity at the left. The jagged portion of the failure is the interface formed when the specimen was statically broken after it had been subdivided from the wheel.

In appraising the presented information about this failure, the principal conclusion reached, on the

Fig. 10—Testing arrangement for fatigue tests on wheel segments.



basis of false conceptions about the factors affecting fatigue strength and serviceability, would have been that the porosity caused the failure. On this basis, efforts to make the wheel sound may have been deemed sufficient to give a serviceable wheel.

Actually, however, the porosity was of minor importance and perhaps even inconsequential. This can be reasoned, first, from the fact that the location of the porosity is random with respect to the location of the failures. (As already observed, the failures were oriented with quite definite relation to the design of the wheel.)

In the second place, previous experiences have indicated that internal porosity has an effect secondary in comparison to the influence of metallurgical conditions existing at the metal surface.

Failure in this wheel, then, was caused only by inadequacies in the design of the wheel, since failures occurred in sound as well as in porous metal. New wheels were produced after certain design changes had been made.

Additional Determinations

In the preceding experience with the landing wheel failure, it was stated on the basis of the tests on cast panels that the porosity in the wheel may not have been of even secondary importance. The results from a fatigue testing program which was carried out on a number of segments sectioned from a landing wheel of given design further substantiate this conclusion.

These segments were flexed to failure on plate bending fatigue machines, as shown in Fig. 10. At the point of failure in this test, which was in the proximity of the location of service fatigue failures (See Fig. 11) porosity existed in amounts corresponding to a variation in the estimated tensile strength of from 25,000 to 33,000 psi. from specimen to specimen.

The results of these tests are plotted in Fig. 12. It will be noted that the test points exhibit little scatter despite the variation in porosity. The scatter in results that does exist bears no relation to the amount of porosity present. In fact, the actual fatigue strength values for these specimens are of the same order as found in cast panels of the same alloy of the highest metallurgical and surface quality commercially obtainable.

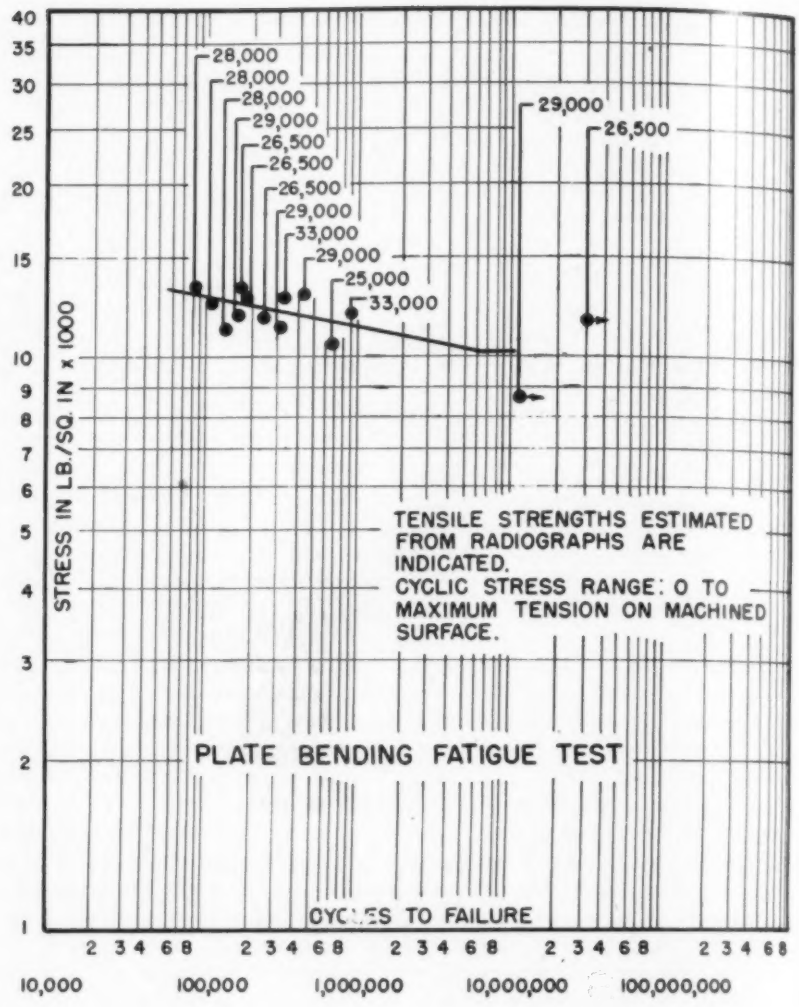


Fig. 12—Results of fatigue tests on wheel segments.

The conclusion to be made from the foregoing tests is that subsurface microporosity in amounts within the acceptable limits does not impair the bending fatigue properties of magnesium.

Porosity Standards

Several interesting relations between service failure and defects in metallurgy and design have been shown. However, it is not intended that confusion be left regarding how porosity should be considered in a casting on the basis of these recounted experiences. The relaxation of standards regarding porosity is not implied.

Such an action would be without due cognizance of either the influence of internal porosity on static properties or the effect of surface porosity on fatigue properties. In addition, the presence of internal porosity in various sections often is associated with defects at the surface.

In summary, it has been the intention to present information which will stimulate more genuine interest in porosity and its actual effect upon serviceability, and to point out that design and conditions at the surface of the metal are of primary importance on fatigue serviceability while internal metallurgy and static strength criteria are of secondary importance.

Conclusions

Fatigue failures originate at points of relative weakness in metal surfaces.

Service and laboratory experiences on parts uniformly stressed show that fatigue failures will be perfectly random in location with respect to internal microporosity or metallurgy; that is, the location of failure has no apparent relationship

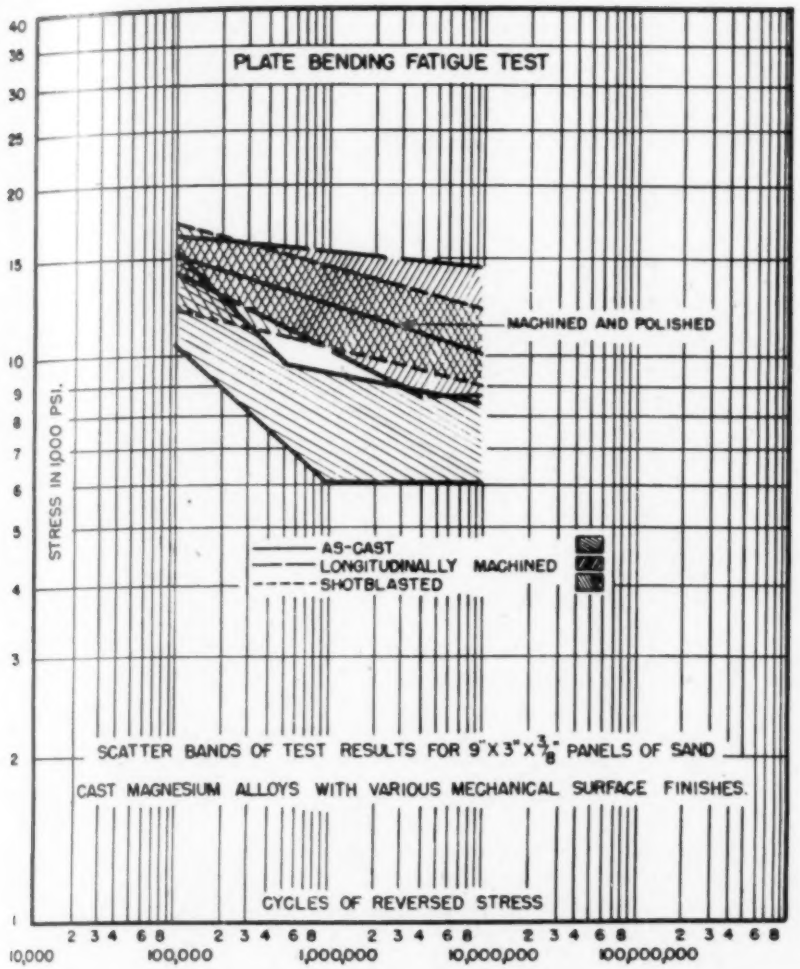


Fig. 13—Plate bending fatigue test results showing scatter bands. Test samples were sand-cast magnesium alloy panels with various surface finishes.

to the proximity of internal defects. The effect of internal metallurgy appears, therefore, to have little influence on serviceability compared to the effect of either the surface or design features. Failures frequently occur through foundry defects or porosity that extend to the surface of metal in the part.

In parts of complex design where stresses are not uniform and stress concentrations exist, the influence on the serviceability of the stress concentration associated with the design surpasses any influence that subsurface porosity may have.

DISCUSSION

Chairman: W. E. MARTIN, National Smelting Co., Cleveland.

Co-Chairman: A. W. STOLZENBURG, Aluminum Co. of America, Detroit.

J. J. MCLAIN[1]: I would like to compliment Dr. Found on his fine paper. Fatigue test results on magnesium castings are very important in the aircraft industry.

During the war our concern was to procure magnesium castings that would withstand the applied load. Sometimes we were successful and sometimes we were not. Surface defects have a deleterious effect upon the life of the casting. That, in itself, reflects back to the care and cleanliness that is used in the foundry operation. Of course, when a casting fails, it is always the foundryman that is to blame, but I am glad Dr. Found also brought out another factor, and that is design. The foundryman is not always to blame for failures.

Dr. Found brought out the fact that microporosity is a subject which cannot be ignored. One of our big problems in aircraft castings was procuring magnesium castings which would be consistently leak-proof.

Microporosity, judging from our investigations, at least, is caused by three main factors. One is gas, another is progressive solidification, and the third is alloy. We appreciate Dr. Found's paper and hope that he will continue with his work and give us more information.

R. A. QUADT[2]: In static stressing of aluminum and magnesium test bars, the effect of surface conditions is likewise extremely important on the properties obtained. In other words, for a test bar which has an inclusion at or near the surface, the effect on the mechanical properties is tremendous, while if that same inclusion is closer to the axis, it has little or no effect. This coincides with the experience you have had as to the effect of surface in fatigue.

CHAIRMAN MARTIN: Dr. Busk, you published a paper some time ago in which you studied the effect of microporosity on fatigue properties of magnesium casting alloys as evaluated by the R. R. Moore machine.

DR. R. S. BUSK[3]: That is right. In the present paper, Dr. Found discussed subsurface porosity when he described the effect as being very small. The bars which I experimented with were cut from panels which were uniformly filled with porosity. When they were machined that porosity came to the surface. What I was measuring at that time was the effect of surface porosity rather than subsurface porosity on fatigue.

CHAIRMAN MARTIN: Is microporosity as a generality, located in magnesium castings so that if too much metal is not machined away from the surface, the outer fibres will contain a relatively small amount of defect? Could this be a factor in the case of roll testing of wheels?

DR. FOUND: That is a difficult question. We are working on that now to get a quantitative relationship between these fatigue results and the proximity of various amounts of porosity to the surface.

CHAIRMAN MARTIN: Dr. Eastwood, does microporosity seem to confine itself to the center? In other words, is there a tendency to have a solid skin? What I am thinking about is that some value might be obtained from all this in design, in that you would try to minimize the amount of metal which you would machine away from the surface, so as not to expose subsurface microporosity, which then might reduce the fatigue life.

L. W. EASTWOOD[4]: I would say that inasmuch as we know that microporosity can cause shrinkage, that proves that the microporosity can extend to the surface. In general, microporosity will be more pronounced towards the center, or at least away from the casting surface.

CHAIRMAN MARTIN: In other words the less metal machined away the better the chances of having sound outer fibres to withstand fatigue loadings.

DR. EASTWOOD: Yes, I think that is true, but there are many other surface defects which may have an undesirable effect on endurance values.

MEMBER: We did considerable inspection of large magnesium castings. It was easy to see microshrinkage coming through to the surface of the castings, so I do not think it is safe to assume that microshrinkage does not come to the surface.

[1]Allison-Bedford Foundry, Bedford, Ind.

[2]American Smelting & Refining Co., Barber, N. J.

[3]Dow Chemical Co., Midland, Mich.

[4]Battelle Memorial Institute, Columbus, Ohio.

DR. EASTWOOD: I did not mean to imply that microporosity would never come to the surface. We can easily prove that it does. However, I think it is also true that, in general, you have more microporosity away from the cast surface than at the surface. But that does not mean that it never comes to the surface.

J. W. MEIER[5]: We made some experiments on microporosity along the same lines as experiments made at Battelle and in some instances the microporosity on the surface or just below the surface was more visible than in the center. I do not say it always was, but there were some cases where we had more microporosity at the surface than internally.

DR. EASTWOOD: We have to be careful that we are talking about microporosity and not surface shrinkage.

DR. FOUND: We have brought up the question of surface conditions which I hoped we would not get into too deeply. Surface is a very important factor in magnesium castings. Fatigue data in Fig. 13 shows the scatter bands for a large number of test pieces. These test pieces were in the shape of cast panels and they were flexed on a plate bending machine. The lowest scatter band represents four panels with as-cast surfaces. The panels were in H-alloy, C-alloy and R-alloy. The heat treatments included the as-cast, the solution heat-treated, and the solution heat-treated and aged states. There was absolutely no relationship between the composition and heat treatment in the results of these tests.

If the as-cast surfaces were machined off, the scatter band would be raised into the region bounded by the long dashed lines. It is evident that a considerable improvement was effected. In other words, panels with as-cast surfaces have a minimum fatigue strength of about 6,000 psi., whereas, if the surface is machined, this value is raised to about 9,000 psi. In other words, removing the cast skin is beneficial.

The third scatter band is for panels with as-cast surfaces which have been shot blasted. Included also is the fatigue curve that was established from solution heat treated panels in C-alloy after they had been machined and polished.

These additional data regarding machining or otherwise cold working the cast surfaces were presented in an effort to clarify any misconception that may have arisen regarding the serviceability of machined magnesium castings. As observed, machining does improve the fatigue strength of magnesium castings. Any implication that the machining may lead to weakening magnesium castings is not generally correct. A machined magnesium casting with porosity intersecting the surface is apparently stronger than a casting with an as-cast surface intersected by porosity in the same amount.

[5]Bureau of Mines, Ottawa, Ont., Canada.

CHEMICAL COMPOSITION of MALLEABLE IRON

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METALLURGICAL PRINCIPLES underlying the making of malleable iron have been described so often [1, 2, 3, 4] as not to require repetition here. The purpose of this paper is to consider the chemical elements in malleable iron with regard to their origin and effect.

Essentially, malleable iron is an alloy of carbon and silicon, the former being the element through which the properties of the product are principally controlled. The latter is used mainly to control graphitization.

Manganese, sulphur, and phosphorus are always present and require control. In addition, almost any elements capable of alloying with iron are likely to be present as "residuals," usually in small amounts, and a few elements not occurring naturally have been added either experimentally or on a manufacturing scale.

Elements exist in any commercial form of iron, either as metallic alloys, dissolved in one or more of the metallic phases, or as non-metallic inclusions, "sonims." The metals lead and silver seem to be almost completely insoluble in liquid iron, although they may be temporarily retained as a colloidal suspension. Mercury can be caused to form iron amalgams by electrolytic means, but they are completely unstable at even moderately elevated temperatures. Alloys of iron and zinc, cadmium and thallium are similarly unstable, permitting the volatilization of the non-ferrous constituent. The alkali and alkali earth metals, lithium, sodium, potassium, rubidium, caesium and magnesium, calcium, barium, and strontium are apparently removed, probably by oxidation, and remain, as alloys, only in minute amounts, if at all. The halides, fluorine, chlorine, bromine and iodine form chemical compounds with iron, which are not retained thereby. Attempts in the author's laboratory to obtain solutions of argon in solid iron failed and it is rather probable that argon and the other "noble" gases, helium, neon, krypton and xenon, could not be found in iron. Some elements are so rare that their effects are undetermined, although rare earths probably all behave like cerium and lanthanum.

Obviously very expensive elements not occurring naturally in iron are of little practical interest, although they may be worth studying for academic reasons.

In general, elements affect the properties of malleable iron in one or more of the following ways:

1. By dissolving in ferrite and hardening or strengthening it.
2. By combining with carbon to form stable compounds.
3. By dissolving in austenite and altering the response to heat treatment either with regard to graphite formation or the separation of cementite.
4. By forming compounds which, though inert themselves when present as sonims, alter the response to graphitization or the engineering properties by interrupting the continuity of the metal.

Effects of various elements in malleable iron on graphitization and mechanical properties are discussed on the basis of a thorough survey of the technical literature and on the author's own researches.

The quantitative aspects of the graphitizing reaction are complex but have been described elsewhere for all but sub-critical temperatures.[5, 6]

It must be borne in mind that the properties of a commercial alloy are not determined solely by chemical composition but are influenced also by manufacturing methods. To make two alloys of different compositions there must, of necessity, have been differences of practice, and if differences of results are observed, it is not necessarily true that they are due to the differences of analyses which were secured. For example: the addition of aluminum in small amounts increases the tensile strength of malleable and increases the graphitizing rate, but also reduces iron oxide to iron with formation of alumina. Are the two effects those of the alloyed aluminum, the addition of alumina, or the decreased oxygen?

Often the results of small experimental melts differ from those in larger commercial operations. Different melting practices are sometimes found to yield inconsistent results and the results of anneal differ both in speed and engineering properties with the environment. Possibly all these things are explicable in chemical terms, but, if so, much more exhaustive chemical in-

Presented at a Malleable Foundry Practice Session at the Fiftieth Annual Meeting of the American Foundrymen's Association, May 7, 1946, at Cleveland.

vestigation would be necessary than is usually possible. It perhaps is too much to hope that the author has weighed correctly all these factors, especially in quoted material.

With this much of an introduction we proceed to consider the elements in alphabetical order insofar as data are available. It presently will appear that the interest in many of them has been mainly from the viewpoint of effect on graphitization.

Aluminum. Aluminum combines so easily with oxygen that it is never found alloyed with iron unless purposely added. The addition to liquid iron alloys of amounts of aluminum leaving only a few hundredths per cent of that element in the metallic state will convert all the oxygen present into Al_2O_3. As pointed out earlier, it is impossible, therefore, save under very unusual and purely experimental conditions, to distinguish between the effects of deoxidation, metallic aluminum and aluminum as Al_2O_3.

Aluminum, when added to molten iron, increases the rate of graphitization, both on freezing and in annealing. It can, therefore, do great damage by causing mottling, and also favorably affect the annealability. The largest amount which may be present, as metal, without detriment by causing primary graphitization, is of the order of some few hundredths of one per cent. Very crudely, this is similar in amount to that needed to deoxidize completely the metal, which may be mere coincidence.

Experiments by C. H. Junge, under the author's direction, indicated a definite effect of aluminum on nodule number as shown below.

Effect of Aluminum on Graphite Nodule Number

Aluminum Added, per cent	*Nodules, per mm.²*
0.0000	1000
0.0063	130
0.0125	430
0.025	450
0.050	1100
0.100	1400

Note that the aluminum is that added, not that recovered. The amount recovered may be only, say, roughly, one-fifth that added in the light of other experience. An increase in graphitizing rate with successive additions after the first might be explained on the basis of nucleization but not the first step.

C. R. Austin[7] is of the opinion that in high carbon steels Al_2O_3 contributes more importantly to graphitization that aluminum.

Very little is known accurately about the accelerating effect of aluminum because of the small concentrations involved, the difficulties of their accurate determination, and the interference of Al_2O_3 and the removal of oxygen. Despite much intensive work, useful quantitative conclusions still are lacking. Possibly aluminum is equivalent to between one and one-half and six times or more its weight of silicon so far as secondary graphitization is concerned. An amount of 0.14 per cent of aluminum to silicon free white cast iron was found, in the author's laboratory, to permit graphitization in a normal, though rather slow, cycle. This is so vague that one may be content with calling it a "powerful" graphitizer.

Primary Graphitization

Aluminum is also a primary graphitizer. Hruska[8] noted that white cast iron became gray when from 0.1-1.0 per cent aluminum was added. He found small additions, 0.01-0.05 per cent, to improve the mechanical properties, but larger additions progressively to deteriorate the product. Such effects certainly exist but are inclined to be erratic. Table 1 gives data on a 32510 grade malleable and on a sub-standard product.

The data are from the author's laboratory and illustrative but not necessarily typical. Much depends upon the state of the metal to which aluminum is added and something even on the oxidizing conditions in annealing.

Antimony. Antimony usually is not suspected as present in malleable although it might enter from white metals not separated from scrap. An addition to iron in fractions of per cents it is quantitatively recovered. Metal containing 0.06 per cent of antimony broke with a "short" grayish fracture, suggesting that produced by the presence of tin. This suggests a possible relation to sulphur, to be discussed under *Tin.* This is not improbable in view of other chemical similarities between the sulphides of the two metals.

In view of this detrimental condition, no detailed study of any effect on physical properties was made, nor is the writer familiar with any discussion in the literature.

Arsenic. Arsenic is very commonly present in commercial irons. Amounts up to some hundredths of a per cent may be present without being noticed. Even 0.37 per cent added to normal iron produced no significant change in properties.

Beryllium. Beryllium is both readily oxidized and volatile. An addition of one-half of one per cent as the element to molten cast iron yielded only something like 0.01 per cent of the element in the resulting material.

Experiments were made adding one and one-half per cent copper, containing six per cent beryllium, to molten white iron and a similar amount of pure copper to another lot of material for comparison. The presumably beryllium bearing metal was ludicrously bad, with 16,400 psi. tensile strength, 0.3 per cent elongation, and 150 Brinell hardness number. The hard iron showed considerable primary graphite.

Here is a suggestion that if the alkali earths could be alloyed with iron by protecting them from oxidation by alloying first with other metals, radical, though not necessarily useful, effects on graphitization might be secured.

Solubility in Iron

Although bismuth is supposed to be practically insoluble in iron[9], Boeghold[10] says that 0.005 per cent bismuth added to an iron containing 2.75 per cent carbon, 1.65 per cent silicon, and 0.40 per cent molybdenum, which would have been gray without the bismuth, caused it to become perfectly white in a one and one-fourth in. diameter section. Anneal at 1700° F. (925° C.) was complete in five hours instead of nine hours or ten hours required for 1.25 per cent silicon iron, the highest which would have frozen white in this section. He suggests the use of 0.002-0.02 per cent bismuth as addition.

The intended mechanism is plainly a suppression of primary graphitization in the presence of enough silicon greatly to accelerate annealing. It is not suggested that bismuth, as such, is an accelerator. A white cast iron containing 0.19 per cent bismuth, properly introduced, showed no evi-

dence of any greatly changed properties.

Boron. Boron is one of the most interesting elements. When added to white cast iron in large amount, 0.88 per cent boron in the metal, it made the product completely unannealable. The microstructure of the resulting white cast iron was that of a hypereutectic alloy, suggesting an enormous shift to the left of its eutectic point.

Dr. Richard Walter[11] suggested to us in a personal communication the use of smaller amounts of boron. Tests with additions (not to be confused with recoveries) of 0.1 per cent boron and 0.04 per cent boron showed an enormous increase in graphitizing rate measured at 900° C.

Graphitization Acceleration

Commercial development[12] has indicated that an addition of 0.001 per cent boron sufficiently accelerates graphitization to serve as useful insurance in annealing without visible alteration of properties. Larger amounts, up to 0.003 per cent, have sometimes been used, especially in the presence of chromium, up to 0.12 per cent. There is, however, some danger of impairment of properties.

It has been originally thought that much boron might form films of iron boride around cementite grains and so inhibit annealing, as discussed under sulphur for iron sulphide films. Alternatively iron carbide containing boron in sufficient amount might be stable and not metastable, a condition known to exist for the oxy-carbide. Amounts of boron approaching the danger point from the viewpoint of lowered ductility and strength result in a greatly increased number of graphite nodules, especially near the surface of castings.

Any mechanism accounting for both acceleration and retardation must evidently be quite complex.

Cadmium. The very small amounts of cadmium which can be retained after considerable additions of that element, and which amount to only a few hundredths of a per cent, seem to exercise no visible effect.

Cerium. Cerium is one element of the group beginning with lanthanum, the 57th element, and ending with cassiopium, the 71st, known as the rare earths. They are chemically so nearly alike as almost to be impossible of separation by ordinary reagents and are quite similar in chemical behavior to boron and aluminum. Many of them are among the rarest metals on earth and none occur naturally in iron alloys.

Cerium alloyed with lanthanum and some other rare earths is an article of commerce, "Misch Metal," obtained as a by-product in the recovery of thoria from monazite. It has been proposed as a scavenger for removing oxygen, sulphur, and nitrogen from molten metal. The alloy contains 50 per cent cerium and the remainder is predominantly lanthanum. Working with the alloy[13], it is not possible to separate the effect of the different elements of the group, which is, however, not very necessary in view of their similarity.

Table 1

TENSILE PROPERTIES OF STANDARD AND SUB-STANDARD MALLEABLE IRON WITH ALUMINUM ADDITIONS

	Sub-Standard Iron		*Standard Iron*	
Aluminum Added, per cent	*Tensile Strength, psi.*	*Elongation, per cent*	*Tensile Strength, psi.*	*Elongation, per cent*
None	48700	7.8	50600	11
0.006	51100	12.8	53300	15
0.013	50200	11.3	53800	18
0.025	49300	15.5	52800	17
0.050	48700	13.8	52600	18
0.100	48600	17.0	50300	16

Alloying Hard Iron

Alloys were made adding one-half of one per cent and five per cent of "Misch Metal" to crucible melts of hard iron. The first two were found to contain 0.22 per cent and 0.36 per cent cerium, respectively. After a commercial anneal their fracture was nearly normal except for a few glittering spots which under the microscope looked like cementite. The last alloy containing 2.48 per cent cerium was extremely sluggish even at high temperature and closely resembled the two others.

Even large amounts of alkali earths do not alter the general character of the fracture. An additional alloy was made without the "Misch Metal" but with 0.10 per cent manganese extra. This is the amount equivalent to the sulphur content of the iron (as MnS). The properties were similar to those of the metal treated with the alkali earths and it was concluded that the small effect noted was only that of the manganese released when cerium (or lanthanum) substitute for that element in the sulphide.

It also was found that when hard iron is impaired in annealability by tin, a sufficient addition of cerium overcomes the difficulty.

Chromium. Chromium almost universally is present in commercial malleable iron, which it enters via steel scrap, which even though not known to contain chromium invariably does so. Since this residual chromium is accidental in origin, no fixed ranges can be given. White cast iron for malleablizing may contain from 0.01 per cent chromium or less, up to 0.13 per cent or more, even when the element is not added purposely.

It is without doubt the most active direct retarder of graphitization which can be encountered in ordinary commercial practice. What constitutes a dangerous amount has been the subject of much controversy, probably because its effect is progressive; each increment of chromium producing some additional retardation, but none producing a discontinuous jump. The late Enrique Touceda regarded even less than 0.01 per cent with suspicion, while others consider 0.04 per cent or 0.05 per cent as unimportant. The length of the annealing cycle perhaps also is an important factor in forming such an opinion.

H. H. Johnson, in the author's laboratory, found that 0.305 per cent chromium added to a hard iron containing only traces of that element increased the time to precipitate a given amount of graphite, either at 900° C. or 710° C., to about three times that to produce the same result in the absence of any but traces of chromium. The

iron contained 2.64 per cent carbon, 0.95 per cent silicon, 0.31 per cent manganese, 0.069 per cent sulphur, and 0.184 per cent phosphorus. It is not to be understood that the ratio is always exactly three to one. This figure is given merely as the ratio in the time ordinate to superimpose most nearly the time graphite curves of the two materials.

H. H. Johnson and G. M. Guiler, in this laboratory, also investigated the graphite formed in a given iron at various chromium concentrations in various times at various temperatures, with results shown in Table 2.

With many irregularities, there is an evident progressive decrease in carbon as one reads down every column. One knows nothing about how chromium would affect the well established form of the time—graphite curve[14]—which makes evaluation difficult. Furthermore, the addition of solid metal removes hydrogen (q.v.) with improvement in graphitization.

The preceding statements may serve as the best quantitative picture that so far can be drawn. There can be little interest in using chromium to alter the mechanical properties of malleable in view of its powerful effect on the retention of combined carbon. Any usable chromium additions cause this retention to a degree which far outweighs the effect of an insignificant amount of alloy.

Nevertheless, pearlitic malleable containing chromium and an abnormal amount of silicon has been proposed[15]. Additions from 0.22 per cent to 1.14 per cent chromium are described coupled with *additions* of 0.08 to 1.00 per cent silicon as a silicon zirconium manganese alloy, apparently usually in an amount from two-thirds that of the chromium addition to equality with that addition. All the material contained combined carbon. The tensile strength obtained varied rather indirectly from about 63,000 to 80,000 psi. The elongation decreased steadily with increasing chromium from 18.5 per cent with no addition to 2.5 per cent at 1.14 per cent carbon with an *additional* 1 per cent silicon.

Various attempts have been made to counteract the effect of chromium by other elements without much success. Goode, in the data given above, did not obtain the ability to graphitize completely (he may not have wanted to) by adding silicon. Other accelerators have been tried unsuccessfully; only boron, already described, having given any signs of utility and even then at some risk of impairment of quality.

Although high chromium alloys lose much of that element when melted under oxidizing conditions, small concentrations are not further reduced by any melting process suitable for the production of malleable.

Anna (Nicholson) Hird, in the author's laboratory, found that the carbides isolated electrolytically from white cast iron containing about two per cent chromium were relatively very low in that element. Hedberg, in the same laboratory, found that the chromium to iron ratio of the material dissolved electrolytically from a white cast iron containing 0.28 per cent chromium was only about one-third that of the alloy as a whole.

Cobalt. It was concluded by Sawamura on the basis of dilatometric curves taken at rather rapidly rising temperature that cobalt accelerates annealing[16]. The method is open to considerable criticism. Two per cent cobalt added to white iron in the author's laboratory did not visibly affect the results. Quite possibly slight acceleration of graphitization would have been overlooked. Cobalt does not seem to have attracted experimentation from the viewpoint of quality improvement. Possibly there was an assumption that the cheaper similar element, nickel, would do all that could be expected of cobalt.

Columbium. Columbium often accompanies tantalum. The very small amount of information available is to be found under this element.

Copper. Copper has attracted more attention than any other metallic alloying element. Since it favors graphitization and improves mechanical and chemical properties, much laboratory and operating attention has been given to it.

Copper in Iron Alloys

Copper is always present in commercial iron alloys in amounts of the order 0.05 to 0.10 per cent. It originates in the iron ore and/or in the contamination of scrap by copper base alloys. When added to iron, copper is recovered completely and it is not removed in melting. Sawamura[16] reported it as favoring graphitization, and Smith and Palmer[17], by isothermal dilatometric means, determined that an iron in which the first stage of graphitization was complete in eight hours at 925° C. would in the presence of 1.33 per cent copper complete this stage in about five hours, at 1.73 per cent copper in four hours, and in the presence of 2.94 per cent copper in two and one-half hours.

It is to be observed that these figures represent, approximately, a linear relation of logarithm of time. Numerically the logarithm of the time for completion of graphitization is reduced by about 0.166 for an increase of one per cent of copper (i.e., the time is divided by about 1.47). After completion of the first stage the times to complete the second

Table 2

GRAPHITE FORMATION IN IRON WITH VARYING AMOUNTS OF CHROMIUM

Temperature .... *900° C.*	*780° C.*	*740° C.*		
Time ..... *5½ hr.*	*62 hr.*	*150 hr.*	*200 hr.*	*300 hr.*
Chromium, per cent	*Graphite Formed, per cent*			
0.008	0.39	0.54	0.12	0.37	2.60
0.018	0.38	0.74	0.09	0.67	—
0.027	0.38	0.53	0.07	0.36	—
0.035	0.25	0.50	0.13	0.54	—
0.045	0.26	0.75	0.12	0.55	—
0.053	0.28	0.79	0.10	0.35	2.43
0.061	0.26	0.36	0.09	0.32	—
0.069	0.24	0.51	0.13	0.44	—
0.076	0.25	0.59	0.07	0.39	—
0.089	0.28	0.58	0.14	0.21	—
0.110	0.28	0.51	0.08	0.39	2.10
0.140	0.26	0.38	0.07	0.27	—
0.260	0.10	0.23	0.05	0.20	1.46

stage at 725° C. were 13 hours, six hours, five and one-half hours and four and one-fourth hours, respectively. These data do not follow the logarithmic relation suggesting a more rapid effect of copper when low than when high. It is, of course, possible that the fact that cooling to 725° C. was not instantaneous, introduced complexities of interpretation.

Fisher and Shumar, with this author[21], found close quantitative parallelism between the effects of copper and nickel. An increment of one per cent in copper multiplied the graphitizing rate at 900° C. by about three, and the effect of copper concentration on graphitizing rate was substantially linear. The effect of graphitizing rate at 700° C. was substantially that described for this temperature range under *Nickel*. The addition of copper up to one per cent, generally speaking, first increased the nodule number at about 0.50 per cent and then again decreased it, if the nodules were germinated at 900° C. For nodules germinated at 700° C. there was an irregular increase in number throughout.

Smith and Palmer also described (loc. cit.) the precipitation hardening by copper, a subject also investigated in the author's laboratory[18].

Malleable Iron Hardness

Thus ordinary malleable iron, air cooled from 700° C., had a Brinell hardness of 142, with or without subsequent tempering for five hours at 500° C. After air cooling from 725° C. and thus tempering, its Brinell number rose to 149. After reheating to 740° C. and air cooling, its Brinell number was 143, which was raised to 176 by tempering at 500° C.

The same metal to which one per cent copper was added had a Brinell number of 173 after quenching from 725° C. and tempering at 500° C. for 5 hours; 153 on air cooling from 740° C. and 181 on quenching from 740° C. and tempering.

In the paper cited, Smith and Palmer plotted tensile data for metal air cooled from various temperatures and tempered four hours at 500° C. These properties were not greatly altered (total carbon 2.40 per cent) unless the critical point was exceeded.

Lorig and Smith[19] showed that 1.7 per cent copper was equivalent to about 0.20 per cent silicon so far as tendency to mottle was concerned, independently of section. They found that, without precipitation hardening, iron containing about 2.35 per cent carbon was but little strengthened by copper up to 2.75 per cent. The yield strength was raised about 7000 psi. by 2 per cent copper and the elongation affected irregularly, but on the whole not improved. This iron contained about 1.10 per cent silicon.

An iron of similar carbon, averaging 1.25 per cent ± in silicon had its tensile strength increased by nearly 10,000 psi., and its yield strength a similar amount, by one per cent copper, but not thereafter again, with no very useful effect on elongation.

An iron containing 2.75 per cent carbon and around one per cent silicon was raised 15,000 psi. in tensile and yield strength by 1.5 per cent copper; its elongation was slightly improved.

Lorig and Smith thought that 0.70 per cent copper was required for precipitation hardening.

Rusting Rate

A copper content of around 0.25 per cent has been found to reduce the attack of malleable iron by rust[20]. This amount of copper reduces the rate of rusting under industrial atmosphere attack by about two mg. per cm.^{2} per yr. It is much more efficient in resisting attack by acids. Attack by moist steam is so slow in any event that no determinable differences due to copper could be found in one year, no losses having occurred in either case.

Manning and Rosenthal[21] report that pearlitic malleable containing 0.80 per cent copper lost only about one-third as much weight in a given exposure to five per cent sulphuric acid solution as did a similar alloy without copper.

According to Wolf and Meisse[22], malleable iron containing either one per cent or two per cent copper corroded in the smoke of locomotive stacks about 0.7 times as fast as the unalloyed metal.

Copper is frequently used in connection with molybdenum, a matter to be touched on when considering that element.

Gallium. British haematite pig usually contains this element in amounts up to 0.003 per cent.[23] It was thought at one time that British pig iron, even of appropriate analysis, was not suitable for making black heart malleable. However, 0.48 per cent gallium added to white cast iron did not produce any annealing difficulty. Spectographic investigation of the resulting metal at the National Bureau of Standards showed the presence of gallium. Exact quantitative estimation was not attempted but an amount of some tenths of a per cent was thought probably present.

Gold. There can be no commercial interest in the iron-gold alloys. Hard iron with an addition of 0.47 per cent gold produced a peculiarly sticky, infusible melt. No evidence of any effect on graphitization was noted. Sawamura[16] regarded it as a slight accelerator.

Hydrogen. White cast iron may contain from 0.0002 per cent to 0.0015 per cent hydrogen by weight. This is derived from moisture in the blast and rust in the charge. Addition of any solid ferrous material, such as ferro-alloys, to liquid cast iron, lowers the hydrogen content. Experience with steel melting suggests that hydrogen can be burned out under conditions which oxidize carbon. Hydrogen is largely eliminated in annealing; all malleable iron having roughly a concentration of 0.00013 per cent hydrogen.

Effect of Hydrogen

It is difficult to conduct experiments in which hydrogen content is the only variable affecting annealing. However, it has been shown that, in general, hydrogen greatly retards "first stage" annealing. Graphitization is rapid below say 0.0005 per cent hydrogen; slow above about 0.0008 per cent and erratic between.

The number of nodules of graphitic carbon is related to hydrogen content in such a manner that the average amount of graphite per nodule after a given heat treatment decreases linearly as the logarithm of the hydrogen concentration increases. This relationship involves both graphitizing rate and nodule number. It indicates that had the nodule number remained constant, instead of decreasing as the hydrogen increases, graphitizing rate would still have decreased with increasing hydrogen. This points to an effect of hydrogen on the migratory rate of carbon. The criterion

described above is much more consistent than any correlation of nodule number and concentration.

It proved difficult to gain any consistent tensile strength information, but apparently hydrogen decreased the tensile strength very markedly but is of less effect on elongation. All the above data were reported by the author, with Guiler and Barnett[24].

Lorig and Samuels[25] found that molten white iron treated by bubbling hydrogen through the liquid, was somewhat improved in annealability and showed an increased number of graphite nuclei. This contradiction to the previously reported results of Schwartz, Guiler and Barnett may perhaps be explained by the existence of a stream of gas in Lorig and Samuels' work, whereas at National Malleable and Steel Castings Company the hydrogen originated in the melting operation.

Heat and Graphitizing Treatments

Heat treating the white cast iron at 600° F. (315° C.) for eight hours before the graphitizing treatment, decreased the size of the graphite nodules in all cases but much more in metal through which hydrogen had been bubbled than in iron not subjected to that heat treatment.

Lanthanum. See Cerium.

Lead. Lead does not alloy with iron but may be present in colloidal suspension. Attempts to add lead to white cast iron, either as metal or as the oxide (which is easily reducible and should be converted to the metal by the carbon of that metal), have yielded low and irregular recoveries and no noticeable effect on the product.

The improvement in machinability observed when lead is added to steel would be obscured in malleable iron by the presence of temper carbon.

Lithium. Attempts to add lithium to white cast iron, even in spectroscopic traces, have so far proved unsuccessful.

Magnesium. Magnesium apparently cannot be measurably alloyed with iron[26]. Experiments in the author's laboratory failed. Additions of copper-magnesium alloys have been tried, though not so high in copper as the beryllium alloy discussed under *Beryllium.*

Manganese. As is well known, manganese is a desired element in malleable iron. Its effect in counteracting sulphur will be discussed under *Sulphur.*

Anna (Nicholson) Hird, in the author's laboratory, found that when cementite was isolated electrolytically from ordinary white cast iron, only the manganese of MnS remained with the cementite; the remainder passing into acid solution with the ferrite. If very large amounts of manganese, approximating ten per cent, are alloyed with the hard iron, the isolated carbides were much richer in manganese than the alloy as a whole.

In this section, only the effect of the manganese not combined with sulphur will be considered. This amount may be from 0.10 per cent up. Experiments with *excess* manganese (above 1.7 x sulphur) ranging from 0.30 per cent to 0.71 per cent indicated that there was no accurately demonstrable effect of manganese on graphitization above the A_1 temperature, i.e., "first stage" graphitization. The effect on second stage graphitization is known to be considerable.

Manganese Additions

It is not unlikely that the effect of manganese depends on how and when it was added, for Johnson, in the author's laboratory, demonstrated that there might even be an acceleration of graphitization in both stages when ferro-manganese was added just before pouring. An effect of deoxidation is suspected.

Forbes[27] reports on a possible increase in mottling on adding ferro-manganese.

Kikuta[28] was of the opinion that the logarithm of the time to complete first stage graphitization increases linearly with the manganese content. His data support this opinion for manganese in excess of MnS. However, he says that the initial rate of graphitization at 925° C. is not much affected by manganese content but the time for completion is, since small particles of cementite (?) remain persistently unaltered. An alloy containing 3.16 per cent manganese did not graphitize at all in 25 hours at 925° C., although one of 2.10 per cent manganese completed the reaction in 22 hours.

Kikuta also attempted some study of second stage graphitization. Very great retardation was observed, although the data justify no quantitative conclusions. The reaction becomes so slow at 1.00 per cent manganese as to be of little practical interest.

Forbes[27] reports the combined carbon content of three alloys differing importantly only in manganese content each after three "Z" metal heat treatments. These combined carbons, listed in the order Mn = 0.34, Mn = 0.71, Mn = 1.00 per cent, and, for the three heat treatments, 0.070, 0.255 and 0.500 per cent; 0.140, 0.235 and 0.430 per cent; and 0.050, 0.260 and 0.545 per cent. The retarding effect of manganese on second stage annealing seems indicated. On the other hand the writer clearly remembers at least one instance in his early experience when a high carbon iron which had accidentally been made with 0.77 per cent manganese annealed satisfactorily in the then standard rather long cycle.

Forbes, Paulson and Minert[49] reported that after a preliminary period, varying from three to ten hours, depending on manganese content, the amount of carbon graphitized at 1300° F. (705° C.) increased as the logarithm of time. Doubling the time, at any stage after the logarithm relation is established, increases the graphite by about 0.14 per cent, regardless of the manganese content.

Graphite Content

This relationship implies that the graphite content does not approach zero asymptotically, but goes on at a finite rate and then stops suddenly, a condition at variance with most opinions on graphitization. Indeed it is a contradiction to an earlier figure in the same paper which suggests an asymptotic approach to zero of combined carbon.

Three points on a combined carbon-time curve are but a slight foundation for determining the form of a curve, especially considering the well known uncertainties in determining combined carbon in the presence of much graphite. Being so bold as to reinterpret the data of Fig. 5 of the Forbes, Paulson and Minert paper on the assumption that the curves are logarithmic after 12 hours, one finds the observations not prohibitively out of line with the principle that the logarithm of the graphitizing rate is approximately a linear function of manganese con-

tent. The departures are in the direction of more rapid retardation per unit of manganese as the concentration of that element increases.

No investigation has detected any definite effect of manganese up to say 0.75 per cent or 1.00 per cent on the solubility of carbon in gamma iron[27]. Manganese additions are well known in the art of making pearlitic malleable when they are used to retard the second stage of anneal.

The direct effect of manganese on mechanical properties is so small as to have escaped observation. Its indirect effect by retaining combined carbon is the only one of practical importance. McMillan[29] suspects excess manganese of causing a decrease in elongation.

The source of manganese is the ore from which the iron was made. As discussed under *Sulphur,* the usual manganese content approximates three times the sulphur, except in arrested anneal pearlitic malleable where it may be some 0.50 per cent higher.

Molybdenum. Molybdenum ordinarily occurs in malleable iron in amounts of a few hundredths of a per cent, but can be picked up in larger amounts when molybdenum steel scrap is used. It is occasionally purposely added in quantities up to 0.50 per cent, or even higher, especially in the presence of copper. Molybdenum can be added either as ferro-molybdenum or by adding calcium-molybdate or similar compounds to the slag.

Schwartz, Johnson and Junge[30] found the times for completing one-half the possible graphitization at 900° C. shown in Table 3. The last result in Table 3 casts some doubt upon the whole matter.

At 700° C. a large portion of the time-combined carbon curve is logarithmic in form, i.e., the logarithm of the combined carbon decreases linearly with time.

The reaction rate $\left(\frac{d \log \frac{C_0}{C}}{dT}\right)$ when C_0 is the initial combined carbon and C the combined carbon at time T at 700° C., T being on the logarithm part of the curve, was:

Molybdenum, per cent	*Time, hr.*
0.00	30
0.23	20
0.45	15

Table 3

TIME IN HR. FOR COMPLETING HALF THE POSSIBLE GRAPHITIZATION AT 900° C. IN MALLEABLE IRON CONTAINING MOLYBDENUM

Fuel Melted Iron		*Electric Melted Iron*	
Molybdenum, per cent	*Time, hr.*	*Molybdenum, per cent*	*Time, hr.*
0.00	8	0.00	8
0.27	13	0.23	12
0.65	35	0.45	11

In fuel melted iron the addition of molybdenum was observed first greatly to increase and then greatly to decrease the number of graphite nuclei germinated at 900° C., whereas in electric melted metal germinated at 700° C. the number of nuclei was somewhat independent of the molybdenum content. Note that any solid addition to molten iron decreases its hydrogen content.

Attempts to calculate the effect of molybdenum on migratory rate seemed to show that on fuel melted iron at 900° C. the addition of 0.27 per cent molybdenum decreased the migratory rate to three-fourths of its original value and the addition of 0.65 per cent molybdenum decreased it to less than one-third its original value. For electric melted iron the migratory rate at 700° C. was little affected by molybenum content, being apparently slightly increased by 0.45 per cent molybdenum, which seems somewhat incredible.

An anonymous source furnished the data in Table 4 showing the effect of molybdenum on the mechanical properties of malleable iron with and without copper. The metal contained 2.32 per cent carbon, 1.14 per cent silican, 0.075 per cent sulphur, and 0.342 per cent manganese.

Amount of copper up to one per cent may be used with molybdenum between 0.25 per cent and 0.50 per cent.

Nitrogen. Nitrogen is found in all white cast iron, originating in the pig iron, and, in duplexed or cupola iron, from the coke. The pig iron got its nitrogen from the same source. Some fixation of atmospheric nitrogen also seems to occur.

The nitrogen content of cupola metal, in the duplex operation, has been found to vary from about 0.0056 to 0.0138 per cent. Averages from different cupolas vary from 0.0083 to 0.0124 per cent.

The metal, while being held in the air furnace, seems to approach an intermediate equilibrium value somewhere between 0.0098 and 0.0116 per cent.

The product of the Brackelsberg furnace contains normally about 0.0090 per cent nitrogen, and this metal, held in the holding furnace, drops to between 0.0056 and 0.0080 per cent.

During muffle annealing the nitrogen content drops to about 0.0031 per cent.

Nitrogen is introduced readily into white iron by melting under potassium ferrocyanide. A certain crucible melt made without such addition contained 2.06 per cent carbon

Table 4

EFFECT OF MOLYBDENUM ON MECHANICAL PROPERTIES OF MALLEABLE IRON CONTAINING 2.32% C, 1.14% Si, 0.075% S, 0.342% Mn

Molybdenum, per cent	*Copper, per cent*	*Tensile Strength, psi.*	*Yield Strength, psi.*	*Elongation, per cent*	*Brinell Hardness No.*
Nil ?	Nil ?	56,000	37,600	26.5	120
0.40	Nil ?	61,000	41,200	25.5	129
0.53	Nil ?	62,700	41,400	25.0	137
0.67	Nil ?	65,200	42,200	21.0	137
0.77	Nil ?	66,000	42,200	22.5	143
0.84	Nil ?	66,300	41,900	21.0	143
0.40	0.30	60,200	41,300	22.5	137
0.50	0.40	64,400	44,700	21.0	137
0.51	0.50	64,600	44,900	20.5	146

and 0.0134 per cent nitrogen. The same melting stock, melted under ferrocyanide, produced 2.36 per cent carbon and 0.0300 per cent nitrogen. Both irons annealed in a normal cycle, although the iron higher in nitrogen had much larger temper carbon nodules.

Mechanical effects of nitrogen are obscured in commercial operations by other variables. It is suspected that nitrogen *might* strengthen the iron and reduce ductility but no convincing facts are available.

Nickel. Nickel occurs in small amounts in almost any ferrous product into whose manufacture steel scrap enters. These "residual" concentrations of nickel are of little interest since this element is an accelerator of graphitization[16] and the small amounts have trivial effects on properties.

First Stage Graphitization

Fiordalis and Trinter with the present author[31] found the effect of nickel on first stage graphitization to be similar to that of an equal weight of copper. Nickel increases the number of graphite nuclei germinated at 900° C. somewhat but rather irregularly. So far as could be observed the rate of graphitization increased approximately linearly with nickel concentration. An increase of one per cent in nickel multiplied the graphitizing rate at 900° C. by 2.9 and at 700° C. by 2.2. Thus the graphitizing rate for any nickel content is that without nickel + 1.9, or 1.2 times the per cent of nickel.

Schwartz pointed out in an earlier paper[32] that in the presence of nickel the graphite-time relation, instead of following largely the relation that the graphite formed, is proportional to the one and one-half power of time, a much lower power approximating 0.7 was indicated. The data of reference 31 did not confirm such conclusion.

In this author's laboratory a silicon-free alloy containing 0.92 per cent nickel annealed about as readily as a nickel-free alloy containing approximately that amount of silicon. A nickel content of 0.52 per cent did not permit complete commercial anneal in the absence of silicon.

No references to the effect of nickel on mechanical properties have been encountered by the writer in any journal still accessible to him. The British Iron and Steel Institute abstracts an article by Rebecca Hall[33] containing a statement that "in small amounts nickel raises the tensile strength and ductility." Fairly extensive experimentation under the author's direction was inconclusive.

Oxygen. Oxygen always is present in white cast iron despite its high carbon content, which apparently should reduce the oxygen content to very low values. We will not discuss the methods for determining oxygen. Using over-killing with aluminum and gravimetric determination of the resulting Al_2O_3 as a means of analysis, the following observations have been made by G. M. Guiler in the author's laboratory.

Duplexed hard iron may contain from 0.004 to 0.019 per cent oxygen. Oil fired duplexing is accompanied by a higher oxygen content than coal firing. The oxygen content near the end of a day's run in duplexing, when metal has stood long in the furnace, likely is quite low. In a duplexing operation in which the metal was held for about one hour (average) in the air furnace the oxygen content of the cupola metal was reduced about 0.008 per cent in the air furnace.

Non-Metallic Inclusions

By volatilizing the iron in chlorine, non-metallic matter can be isolated from hard iron or malleable iron. Since it usually is necessary to ignite this residue to remove graphite, it is by no means certain that the elements are in the same state of oxidation or combination with one another as in the original metal. The amount of this residue from hard iron is somewhat variable but is of the order of 0.12 per cent; that from malleable iron made by annealing the same hard iron may be three or four times as great. The ignited residues from hard iron apparently consist of SiO_2, P_2O_5, Fe_2O_3 with only traces of manganese. After anneal, P_2O_5 disappears and large amounts of Mn_3O_4 are found. There is also a considerable increase in total amount of Fe_2O_3 and SiO_2. During annealing the metal absorbs oxygen.

It is usually possible to demonstrate a higher SiO_2 content near the surface of a casting than at the center. An amount of SiO_2 equivalent to 0.10 per cent silicon in the iron is not unusual. Such oxidation is suspected of contributing to the formation of rims. Attempts to identify particular mineral species by x-ray spectrographic methods thus far have been fruitless.

Boegehold[34] presents information that solid iron oxide added to liquid white iron promotes primary graphitization. He points to surface oxidation of "drops" or "skulls" in ladles as a source of such oxide (Fe_3O_4?). He also pointed out that the addition of solid iron oxide to molten iron produced a decrease in temper carbon nodules which was taken to mean a decreased annealing rate. It thus appears that iron oxide promotes graphitization during freezing and retards it during annealing, a most undesirable combination.

Mechanical Properties

Nothing is known of its direct effect on mechanical properties, although an increase in ductility from properly controlled decarburization is possible. On the other hand many cases of "low" rims can be found when the strength and ductility of metal oxidized in annealing is less than that at the center where no oxidation in the solid state has occurred.

Palladium. The addition of one per cent palladium produced no visible effects.

Phosphorus. The source of phosphorus is the calcium phosphate in the remains of marine animals which lived in the water from which the iron ores were deposited.

The phosphorus content of malleable iron usually depends upon the amount of steel scrap in the mix. It usually falls between about 0.08 per cent and 0.19 per cent. Kikuta[28] found that phosphorus, even up to one per cent, accelerated first stage graphitization, though somewhat irregularly. The relation may be somewhat logarithmic but the data are not completely consistent. Kikuta regards the effects of sulphur and phosphorus to be additive in some degree. Effect of phosphorus on second stage graphitization proved negative up to about 0.45 per cent; at 0.79 per cent graphitization could not even be started at 710° C.

Phosphorus increases the tensile strength and yield strength of malleable iron and affects the elongation but little. There is not sufficient data for precise statements but the increase in tensile strength for an

increase of 0.05 per cent phosphorus may be 1,000 to 3,000 psi., and the increase in yield strength 2,000 to 4,000 psi. The change in elongation may be of the order of one per cent.

The principal limitation on phosphorus content is its tendency to produce white, inter-crystalline fractures, especially in the presence of silicon.

Silicon and Embrittlement

W. R. Bean[35], on the basis of Humphrey notch bar tests, showed that iron containing over 1.5 per cent silicon and 0.10 per cent phosphorus, or over 0.50 per cent silicon and 0.26 per cent phosphorus, had a white fracture as usually annealed. L. H. Marshall[36] showed this embrittlement to be related to the time the iron remained at about 900° F. (480° C.), and that the effect could be removed or prevented by quenching from near 1200° F. (650° C.).

While this treatment is effective, it has been found in the author's laboratory that such treatment is not completely effective in rendering high phosphorus and/or silicon material the equal of low phosphorus and/or silicon material at sub-zero temperatures. Phosphorus increases the fluidity of white cast iron, but only about one-twentieth as much as an equal weight of carbon.

Platinum. Platinum behaved like an equal weight of gold (q.v.).

Selenium. Selenium behaves, atom for atom, like sulphur. Since its atomic weight is 79 and that of sulphur 32, it is $\frac{32}{79}$ or about one-half as potent as sulphur. In view of its higher atomic weight, a given concentration of selenium is equivalent to a mass of metallic selenide less than the mass of a sulphide from the same concentration of sulphur. This should reduce the quantitative effect of selenium, compared with sulphur.

Silicon. Silicon is derived from the silicious material accompanying iron ores, or, in the case of ferro-silicons, directly from silica sand.

This element has excited the greatest interest of all since it furnishes the most common commercial means of controlling graphitizing rate and tendency to mottle. The silicon content considered desirable seems to have increased steadily. A quarter of a century ago[4] silicon contents as low as 0.45 per cent were referred to, and concentrations over one per cent were considered applicable only to very low (2.10 per cent) carbon iron. McMillan[37] now recommends silicon contents from 1.60 per cent to 1.90 per cent for carbon contents from 2.00 to 2.40 per cent. One element should go up as the other goes down to keep the sum at about 3.90 per cent.

Gladhill and White[38] probably were the first to publish a truly quantitative study of the effect of silicon on graphitizing rate. They found that the time required to complete graphitization at 1650° F. (900° C.) was about eight hours for 0.90 per cent silicon, 18 hours for 0.83 per cent silicon and 23 hours for 0.60 per cent silicon, whereas at 1500° F. (815° C.) the corresponding times were 37 hours, 80 hours and 130 hours, respectively. It is this writer's recollection that in a French paper, no longer accessible to him, these authors expressed the opinion that equal increments of silicon multiplied the graphitizing rate by equal amounts. The above figures do not fit this hypothesis particularly well. This relation has been found to apply reasonably well in experiments by Hird and Guiler in this laboratory. Hird and Wahrenberger, under the author's direction, found that the graphitizing rate of an iron containing 1.13 per cent silicon was 3.45 times as high as that of one containing 0.69 per cent silicon at 900° C., and 18.4 times as high at 780° C.

Silicon on Mottling

The effect of silicon on mottling has not been very accurately investigated, perhaps because primary graphitization is influenced so much by other conditions. McMillan's constant of 3.90 points to equality of effect. Graphs[4], based on earlier work by Thrasher, seem to confirm this constancy for small cross sections, although a sum of carbon and silicon of 3.60 was indicated. This equality, with smaller and smaller sums, exists throughout that diagram at reasonably low total carbons (under 2.50 per cent) and for very heavy sections was given as about 3.20. The author's operating experience is that a smaller drop in silicon accomplishes more than a larger drop in carbon in overcoming mottling. Observations on mottling are confused easily by vagaries of temperature, melting stock, cooling rate, etc., and hence prove to be erratic.

It generally is difficult to obtain data on the effect of silicon on mechanical properties which exclude the effect of carbon. It would seem that silicon, being soluble in ferrite, should harden the matrix of malleable iron. The writer's data, based on a statistical study of many thousands of air furnace heats, indicate that, in general, up to about 2.75 per cent carbon, changes in silicon content from 0.65 to 0.95 per cent have little effect on tensile strength. There is, however, a downward trend of a few per cent in elongation. McMillan (loc. cit.) reports a tensile strength of 59,700 psi. and an elongation of 14.8 per cent for the iron he described. No data are given, however, for metal of similar carbon and lower silicon, so no comparison for the effect of silicon is possible. He also makes the statement that some primary graphite may be tolerated.

The effect of silicon on "galvanizing embrittlement" has been discussed under *Phosphorus.*

Segregation of Silicon

Anna (Nicholson) Hird[51], in the author's laboratory, found that silicon segregated toward the last freezing metal and was found more largely in connection with the cementite, the slower the metal had frozen.

M. M. Austin at the same time pointed out that finding the silicon with the cementite did not necessarily mean that it was a constituent of that phase. It is not likely that any silicon was with the carbide as SiO_2, but the absence of a silicide of iron as a discreet phase was not proved.

Kasper and Mertz, with this author[50], concluded that silicon, like carbon, improved the machinability of malleable iron. Contrary to much popular opinion, the effect of silicon on fluidity is negligible.

Sulphur. Most of the sulphur of malleable iron originates in the fuel used at various stages of its making. Unlike most of the other elements the effect of sulphur depends entirely upon the form in which it is present. The discovery of the neutralizing effect of manganese on sulphur by Alfred Hammer in the

late 1870's[4] is probably the first application of metallurgical science to malleable manufacture. Hayes and Flanders[39] discussed the matter without any apparent recognition of earlier knowledge on the subject.

Sulphur and its kindred elements, selenium and tellurium, are so situated in the periodic table that they readily form stable chemical compounds. These compounds are held in solution in liquid iron, and, on cooling, separate either in the liquid or solid states. If they separate as solids they occur in the frozen metal as "ideomorphic" crystals, i.e., in fragments which have, more or less perfectly, taken the external crystal form peculiar to their composition. If they separate as liquids they coalesce into drops of practically spherical form which, if their melting point is above that of the metal, solidify and are entrapped in the freezing metal. If their melting point is below that of the metal, liquid is finally trapped between growing dendrites and form "allotriomorphic" inclusions, i.e., inclusions whose form is imposed on them by the surrounding crystals. Frequently these are thin films which prevent carbon migration and hence retard graphitization.

Stability of Sulphides

Levy[40] found the iron sulphide (of pure iron carbon sulphur alloys) in the eutectic into which it introduced "intervening layers." It has been found in the author's laboratory[13] that the range of stability of the sulphides, beginning with the least stable, is iron, manganese, tin (and antimony?), lanthanum and cerium. The sulphides of tin, and probably antimony, behave like those of iron, being film formers, while those of manganese and the rare earths are without effect. Aluminum sulphide is as stable as manganese sulphide, and is not a film former. In view of the limited amounts of aluminum which are permissible, the resulting effect is not detectable. Copper sulphide probably is more stable than manganese sulphide and is a film former. This could account for occasional observations that copper added to high sulphur irons retards graphitization. It might do so by liberating a relatively large amount of manganese from combination with sulphur.

If all the sulphur is to be present as a compound, say MnS, an excess of the metallic element is necessary for such reversible reactions as

$$FeS + Mn = MnS + Fe$$

will reach equilibrium if:

$$\frac{(MnS)\,(Fe)}{(FeS)\,(Mn)} = k$$

no matter what the several concentrations actually are. Now the concentration of iron (Fe) is nearly the same in all irons (about 0.96). The desirable concentration of iron sulphide, FeS, is some small constant amount. Nearly all the sulphur in the iron is to be MnS, hence MnS is practically $\frac{55}{32}(S)$ when S is the sulphur content of the iron as MnS and since FeS is small, practically the sulphur content of the alloy. Taking all these constants into the constant k':—

$$\frac{(S)}{(Mn)} = k'$$

whence it follows that the ratio of Mn to S to achieve a given small residue is constant. Its actual value is about 3. If the valence of the other metal is not equal to that of iron, the relation may be more complicated.

The allotriomorphic sulphides are all retarders of graphitization. Whether because of this retardation or because of some more direct mechanism unknown to the author, iron containing more sulphur, by weight, than manganese, as in white heart malleable, produces when graphitized a very dense, "spherulitic" form of temper carbon shown in Fig. 1, entirely unknown in black heart practice, except perhaps at very low annealing temperatures. Thanks are due to A. E. Peace of Ley's Malleable Castings Company, Ltd., of Derby, England, for the specimen from which Fig. 1 was made.

Calcium and sodium form stable compounds with sulphur, which readily float out. Thus soda-ash is an active desulphurizer. In basic electric furnace practice, white reducing slags containing calcium carbide, the sulphur usually is not higher than 0.04 per cent and may be brought below 0.01 per cent. In conventional air furnace melting, sulphurs are likely to be about 0.07 per cent using coal containing not over one per cent sulphur. All duplexed iron is high in sulphur containing from 0.10 per cent to over 0.20 per cent, unless desulphurization is practiced. The surface of malleable castings is always higher in sulphur than the center, pointing to a sulphur pick-up in annealing. The amount of this pick-up cannot be recorded exactly but in ordinary cases the surface may be a few hundredths of a per cent higher than the center.

The effect of sulphur on annealing, being in part dependent on the distribution in space of the sulphides, cannot be measured satisfactorily. Qualitatively it is known that sulphur rather energetically retards graphitization, during freezing and in annealing. However, even white heart malleable containing one or two-tenths per cent sulphur combined with iron still graphitize to some extent.

The effect of sulphur on the time graphite curve apparently is to prolong the so-called "incubation" period[41]. This should mean a retardation of migratory rate as compared with crystallization velocity of graphite or rate of solution of cementite[5].

The idiomorphic sulphides affect tensile properties about as much as an equal volume (about one-half their weight) of carbon. This has been confirmed roughly, in this laboratory, for manganese sulphide.

Tantalum. Tantalum is not to be



Fig. 1—Spherulitic temper carbon in high sulphur-low manganese white heart malleable, also showing some sulphides. 500X. Very light Nital etch. (Eleanor M. Smith, in the author's laboratory.)

expected in malleable unless purposely added. The addition of 0.46 per cent tantalum and 0.093 per cent columbium to molten white iron produced no detectable effects.

Tellurium. The presence of tellurium in malleable, except when purposely added, has not been reported. With regard to the formation of metallic compounds it behaves, atom for atom, like sulphur, and hence is $\frac{32}{127.6}$ times as potent. See also the section on *Selenium.*

Boegehold[42] finds that small amounts of tellurium, ranging from 0.002 per cent to 0.02 per cent, increase the silicon content which can be tolerated without mottling, and by virtue of this higher silicon the metal is more annealable. The relation of tolerable increase of silicon to tellurium addition is of the order of 0.08 per cent silicon per 0.001 per cent tellurium.

Omitting Silicon Addition

A. E. Peace has informed the writer that as the result of several years of experience it is his custom to add 0.003 per cent tellurium and 0.30 per cent silicon to metal for castings prone to mottle. The end is successfully accomplished but the graphitizing rate is approximately unaltered. The omission of the silicon addition results in hard "rims." No data are available regarding the effects of tellurium on mechanical properties.

Tellurium. Incomplete work in the author's laboratory indicates that tellurium may have a profound favorable effect on the form of graphite nodules and thus improve the mechanical properties.

Thallium. Thallium probably does not occur accidentally in malleable iron. The addition of one-half per cent thallium to molten hard iron produced no result. In the absence of satisfactory analytical procedures it is not known whether any was recovered in the metal.

Thorium. Thorium probably does not occur accidentally in malleable iron. The addition of one per cent thorium to white cast iron produced no results. No evidence of its presence could be had analytically nor was the product sufficiently radioactive to fog a photographic plate on long exposure.

Tin. Tin is encountered occasionally in amounts of several hundredths of one per cent, having been accidentally introduced in malleable iron when scrap contained bearing metal or tinned parts. It is extremely objectionable not because of any direct effect but by combining with the sulphur of manganese sulphide to form an allotriomorphic sulphide as explained under *Sulphur.* When sulphur is stabilized by cerium and lanthanum, tin, at least in small quantities, appears to be without effect on the metal.

Titanium. Most iron ore contains a little titanium, some of which is recovered in the blast furnace. Melted under oxidizing conditions titanium disappears rather easily from the metal. In the presence of about 0.10 per cent silicon, metal containing 0.15 per cent titanium graphitizes as fast as titanium free metal containing 0.60 per cent to 0.75 per cent silicon[40]. No attempt has been made to trace the effect of such amounts of titanium on mechanical properties. Qualitatively it is believed that less tendency to mottle is encountered for equal annealability, using titanium rather than silicon.

Tungsten. Tungsten is not normally encountered in malleable iron. It can be introduced into molten cast iron as ferro-tungsten but difficulty is encountered because the alloy has a high melting point and easily covers itself with tungstic oxide when heated. Alloys have been made in the author's laboratory containing up to 2.88 per cent tungsten. In the absence of silicon this amount of tungsten did not produce any significant amount of graphite in ordinary annealing. Added to normal white iron 1.29 per cent tungsten did not retard annealing.

There is believed to be a paper by Anson Hayes and a co-worker, dealing with the joint effect of tungsten and manganese, which has not been traceable through the usual library approaches. Data as to the mechanical effects of tungsten on malleable iron are lacking.

Occurrence in Malleable Iron

Uranium. Uranium is another element not to be expected to occur accidentally in malleable iron. Investigation in the author's laboratory has disclosed that in absence of silicon, even traces of uranium will permit graphitization to proceed at rates comparable with those caused by about 0.75 per cent silicon. A concentration of 0.3 per cent uranium and no silicon caused very rapid graphitization.

Uranium is oxidized easily when added to white cast iron, only a minor percentage of the addition being recovered. The element apparently is a more powerful graphitizer than titanium so far as may be judged by qualitative evidence.

Vanadium. Vanadium is easily oxidizable and not normally present in malleable iron. It forms very stable compounds with carbon and perhaps iron. This carbide remains unaltered in heat treating and is of little mechanical effect except that it supposedly increases the metals' serviceability as a bearing[45]. Its use is suggested more for pearlitic than normal malleable.

Zirconium. Zirconium is an element not occurring naturally in malleable iron and is oxidized easily. It is a very active graphitizer.

Irons containing 0.60 per cent silicon and 0.21 per cent zirconium had properties characteristic of "low" iron, that is, iron containing too much carbon and silicon. Even 0.035 per cent and 0.055 per cent zirconium in the presence of 0.045 per cent silicon annealed readily and with some tendency to be "low."

Having described what seems to be known about the individual elements, it seems important to attempt some generalization. For these to be of any interest we must go beyond the hackneyed platitude that chromium, manganese, etc., retard graphitization because they form stable carbides, which says that they promote stability because they are stable and adds nothing to our comprehension.

Under *Sulphur,* the effects of certain elements which act solely because they form compounds of certain general characteristics, has been sufficiently discussed. These are the following:

Elements Forming Inert Compounds

S	Mn
Se	Sn
Te	Sb
B* (in large amounts)	La-Ce
V*	Cu ? (when it retards)

*Boron and vanadium are here added to the metals considered under *Sulphur* because it seems highly probable that so far as they form carbides these com-

pounds remain present as inert phases and otherwise contribute nothing in either direction to graphitization.

The following elements appear to affect the graphitizing by their presence, while not themselves forming separate phases. They are given in the order of their atomic diameters, when known. These diameters are shown and also the group in the periodic table where the elements are found. The group number is also the number of valence electrons.

Elements Hindering Graphitization

Element	*Atomic Diameter (A°)*	*Group*
H	?	I
O	?	VI
Cr	2.49	VI a
Mn	2.58	VI a
Mo	2.72	VI a

Elements Favoring Graphitization

B	—	III
P	—	V
Be	2.24	II
Si	2.35	IV
Ni	2.49	VIII a
Co	2.51	VIII a
Cu	2.55	I b
Al	2.86	III
Au	2.88	I b
Ti	2.90	IV a
Bi	3.10	V b
Zr	3.18	IV a
U (more than W = 2.75)		VI a

For comparison, carbon is in group IV with an atom diameter of 1.42 and iron in group VIII a, with a diameter of 2.48.

Probably we do not know enough about the participation of the elements hydrogen and oxygen to engage in much discussion. Except hydrogen, all the retarders are in group VI, which also contains sulphur, and in sub-group "b," the elements selenium and tellurium. All the metallic retarders have atom diameters larger than iron. They all have diameters near enough to that of iron to permit formation of solid solution, (within 12 per cent or 15 per cent according to Hume-Rothery[46]), and these solutions are known. Two of them are known to be present in greater concentration in the carbides at low concentrations than in the ferrite.

All the metals in group I which can alloy with iron are accelerators. Beryllium is the only metal of group II which has been kept alloyed with iron.

Boron and aluminum are the only elements in group III, about which much is known, though Ga in III b is nearly neutral.

All the elements of group IV which have been alloyed with iron are accelerators except that tin (Sn) has its effects obscured by sulphur.

Phosphorus and bismuth in group V, may be compared with V whose carbides are inert and neutral. Tantalum in sub-group "a," and with neutral arsenic and antimony whose effect is obscured by sulphur in sub-group "b." Vanadium is alone as a retarder in group V. Tungsten in group IVa is inert, molybdenum and chromium are active retarders and the early elements in group VI and those in group VIb are compound forming.

In group VII manganese is the only alloyable element and is a retarder. Group VIIIa does not occur in iron and group VIIIb contains iron and the accelerators cobalt and nickel, also the noble metals whose effects are weak and not thoroughly explored.

Most of the accelerators have larger atoms than iron but a few have smaller atoms. Within group VIa, however, the retarding effect first diminishes to zero, and the accelerating effect then makes its appearance as the atomic weight increases. It is very likely that the uranium atom is larger than tungsten in this group.

It seems fair to assume that four valence electrons are favorable to an atomic structure which will alloy and favor graphitization. As the number falls first to three and then to two and one, the ability to alloy decreases, but where preserved the elements remain graphitizers.

Atomic Weight or Size

Five valence electron elements are not always accelerators but may be neutral or compound forming. Elements containing six or seven valence electrons are either retarders *per se* or compound formers, with exception of neutral tungsten and accelerating uranium. It has been pointed out already that in group VI the effect on graphitization is related to atomic weight and/or atom size.

The metals containing a complete outer shell of eight electrons are accelerators or neutral behaving somewhat like the adjacent sub-group Ib. In this area there seems to be a tendency for the effect on graphitization to decrease with atomic weight.

It is quite possible that we must view the effects of boron and bismuth, which are used in traces only, from the viewpoint of nucleus formation. Indeed bismuth can scarcely exist in the metal in any other form. This mechanism for boron and aluminum is pretty well established since both these elements definitely increase the number of temper carbon nodules. A similar increase has been noted as an effect of nickel and of copper. Hydrogen, chromium, and molybdenum have been shown to decrease nodule number, but manganese, also a retarder, to increase it. None of these changes are so enormously marked as in the case of the accelerators boron and aluminum.

The accelerating elements in groups II, III, and IV are also very active deoxidizers. One is prone to consider the formation of nuclei of their oxides as an important factor in explaining their effects on graphitization. Manganese, however, is an active deoxidizer and apparently enucleates the metal it retards.

Barnett, under the author's direction, has shown[47] that graphitizing rate may be proportional to the square root of the nodule number. The theoretical value would be the two-thirds power[5].

Element Distribution

Of the limited number of elements investigated, silicon, chromium, and manganese seem to distribute themselves between iron and carbide in such a manner as to concentrate in the latter phase. This is true, however, of chromium and manganese only if the concentration of these elements is rather large—decidedly larger than it would be in white iron for conversion to malleable. For small amounts of manganese and chromium the distribution seems to be in the opposite direction.

Of the alloys favoring graphitization, titanium, silicon, and aluminum restrict the gamma iron field. Chromium does also, a fact which Zener[52] regards as anomalous.

The elements manganese, nickel, copper, and zirconium expand the gamma region. Apparently little consistency can be found between the effects of the elements on graphitization and effect on the gamma loop.

The effects on primary graphitization, while not neglecting an effect

on graphitizing rate, cannot be studied without reference to susceptibility to supercooling which involves the question of how crystals begin to grow from the liquid, and cannot be discussed here.

Those elements which retard graphitization must, in practical cases, be present in relatively small amount or produce pearlitic malleable. In the latter case their effects on mechanical properties are overshadowed by the effects of the retained combined carbon. That problem should be studied from the viewpoint of the relation of combined carbon in various metallographic forms to properties.

The very active accelerators must again in practice be present in quite limited amount or mottling will be produced and the effect on mechanical properties is that of flake graphite and not of the particular alloy.

This leaves us with a very limited number of metallic alloys whose effect on mechanical properties could have any real interest. These are the slight retarders manganese, molybdenum, the accelerators silicon, copper, nickel, and cobalt and the neutral elements tungsten. The two last mentioned have not been attractive on economic grounds. Copper and molybdenum have been successfully used as described under these headings. The effect of silicon has been discussed under that heading. No formal knowledge has been available concerning nickel or the small range of manganese content which would not interfere with annealing.

It seems almost self evident that any of these elements should strengthen and stiffen ferrite. The observations regarding copper, molybdenum and silicon—the latter within a range consistent with graphitizing conditions—at least qualitatively confirm this conclusion.

Mechanical Properties

Alloy elements might, of course, also affect mechanical properties by controlling the form and number of temper carbon nodules. These characteristics are, however, largely related to heat treatment.

Schindler and Elliott, with the author[48] showed that, in a given iron, decreasing the nodule number per mm.^{3} from 41,000 to 135 altered the tensile strength but little, but a further decrease to 30 increased it 2,000 or 3,000 psi. A reduction in the number of nodules per unit volume, however, increased the elongation. At an average nodule number of 35,000 per cubic millimeter the elongation was about 14 per cent and at an average of 80 about 21 per cent. An acceleration of annealing rate due to an increase in nodule number may be accompanied by corresponding effects.

In this paper an attempt has been made to summarize the facts, known to the author, regarding the effects of chemical composition on qualities of malleable iron. A sincere attempt has been made to survey the literature with reasonable completeness. Apologies are due to any authors whose work has escaped notice, either through ignorance, inadvertance or because the available library facilities did not disclose it. In a few cases references deemed repetitions were omitted. Continental references were in large measure neglected, fearing possible confusion with white heart malleable. It is the writer's hope that this summary and particularly the appended references may facilitate the work of those who may wish to work in the field of alloying malleable.

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DISCUSSION

Chairman: A. M. FULTON, Northern Malleable Iron Co., St. Paul, Minn.

Co-Chairman: C. F. JOSEPH, Saginaw Malleable Iron Div., General Motors Corp., Saginaw, Mich.

J. E. REHDER[1]: Was silicon largely in the carbide, segregated with the carbide?

DR. SCHWARTZ: If you separate the ferrite phase from the carbide phase by electrolytic current passing from the hard iron into the solution, the silicon is very largely left with the carbide. However, it might not be in the carbide. It might be a silicide which was left with the carbide, but so far as where it is when you analyze the carbide, it forms part of the carbide.

That study was made many years ago by Mrs. Hird, then Anna Nicholson, at our laboratory. One of the men in the laboratory suggested what I just said. The silicon does not have to be in the carbide to be found with it analytically. The only thing that is known is that it is located in the carbide. Whether mechanically or chemically is not known.

C. F. JOSEPH: How much of the various elements could be in an iron and still not affect the iron to any great extent as far as mottling or graphitization is concerned? Very often I have been asked how much chromium, molybdenum, or nickel could be in an iron without changing the iron from the standpoint of graphitization in the hard and the annealed state.

DR. SCHWARTZ: The opinions about chromium have ranged from the late Dr. Touceda's idea that even thousandths of a per cent were detrimental up to perhaps 0.1 per cent. My guess is that in commercial practice chromium is not determined as a routine check. We make commercial iron and do not recognize that you may have 0.05 per cent chromium. I think if the iron has 0.10 per cent chromium it will be so bad you will know it by the usual mechanical tests. If the chromium content is 0.02 or 0.03 per cent I do not think it will be evident unless an analysis for chromium is made.

The percentage of molybdenum could be quite large. I would not hesitate to put 0.5 per cent molybdenum in iron and still expect to do ordinary annealing.

There seems to be a trend for nickel, even if it does not mottle, to help produce irregular forms of graphite. On the other hand, that cannot always be so, because some good results were obtained using nickel. We made several tons of experimental high nickel alloys when we were working with them, and I concluded that it was not worth further effort.

Boron is an accelerator in iron with 0.001 per cent boron and even more so with 0.003 per cent; so much more that quality of the iron can deteriorate. Boron counteracts the chromium to some extent. With 0.003 per cent boron in iron, 0.1 per cent chromium in the iron may anneal. A boron content of 0.25 per cent in iron anneals faster than no boron at all. It does not produce good iron. One per cent boron in iron does not anneal under any circumstances. I think the carbon is permanently in the carbide state.

I doubt if one would put enough tungsten in any commercial iron to have to consider what might happen.

Probably 1.5 per cent of copper in iron would improve it, causing it to anneal better and be stronger. Copper may be added accidentally as a copper-tin alloy. When tin is put in, the tin takes the sulphur from manganese.

Tin sulphide forms films. Copper sulphide is a fairly stable compound which may perhaps remove sulphur from manganese sulphide but it makes the idiomorphic form and does no harm, except as it eliminates manganese. This is largely conjectural. Manganese is not specifically an active retarder in the first stages of annealing. It is in the second stages of annealing.

The effect on primary graphitization is usually less marked. The mechanism for affecting primary graphitization may have to include an effect on the tendency to supercool. It is my impression that such elements as chromium and molybdenum must be present in much larger amounts to decrease mottling than to retard annealing.

ERIC WELANDER[2]: What is the most desirable range of phosphorus content?

DR. SCHWARTZ: For ordinary malleable iron, 0.18 per cent.

MR. WELANDER: What happens to the iron when the phosphorus drops below 0.18 per cent?

DR. SCHWARTZ: The tensile strength drops.

MR. WELANDER: What will happen to the elongation?

DR. SCHWARTZ: Not much. There will be no improvement.

DAVID TAMOR[3]: Often in heavy sections metallurgists will add carbide stabilizers to prevent mottling. They may not know how much residual chrome is in the melt and will end up with an iron that may or may not be capable of annealing.

It would be interesting to know, with 0.05 per cent chrome for instance, what the limitations on a carbide stabilizer are.

DR. SCHWARTZ: I do not know why anybody would want to put carbide stabilizers in heavy sections, when they could leave the silicon out. It is analogous to making a man sick and then administering medicine to make him well, both at the same time.

At the present time nearly any scrap iron you analyze contains nickel. I have no idea what the average residual nickel is, but I am quite sure that the amount makes no substantial difference.

Almost all scrap iron contains chromium. I receive from our plants analyses which read "Chromium Nil," but I am very suspicious of these results. I think it means the chemist got so little chromium he did not want to report it.

I would not be concerned if it was 0.02 per cent, but I might be concerned if it was 0.05 to 0.10 per cent. I would be concerned about where the purchasing department was buying the scrap.

If you are going to combine chromium and nickel you have a major research on your hands, because you would have the independent variation of two different variables.

I believe there is molybdenum in most malleable scrap now. There is enough molybdenum in steels. Those small amounts will not oxidize, so no doubt we are picking up molybdenum.

Possibly what one ought to do when he has unexplained misanneals is to determine all the elements which can be tramp alloys and study the results.

You will always have copper. I certainly would not start on the molybdenum-chromium idea without wanting to consider whether copper was always

present because we know that you can use 0.5 per cent molybdenum and 0.5 per cent copper or one per cent molybdenum and 0.5 per cent copper and get along.

So many of the other elements occur seldom in malleable iron. Tin occurs often. I would be concerned about tin as much as about chromium concentration-wise. Tin is not so scarce in malleable iron because bearing metals are frequently mixed in scrap. When the tin gets into the iron it stays there.

W. G. FERRELL[4]: Dr. Schwartz, you stated that the aluminum was an accelerator.

DR. SCHWARTZ: Very much so.

MR. FERRELL: We find it to have just the opposite effect in 1.40 to 1.60 per cent silicon iron. We cannot dissolve the pearlitic matrix with the use of aluminum in malleable iron. We have never been able to find out why.

DR. SCHWARTZ: No reason occurs to me why the presence of these silicon contents should alter the effects. It is true that if aluminum is present in amounts of several per cent it is no longer an accelerator but this is of academic interest only to the malleable founder.

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GAS ELIMINATION IN NON-FERROUS ALLOY CASTINGS

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GAS IN METALS is a subject which usually brings on a lively debate in any gathering of foundrymen. The following observations and preliminary experiments are offered as more fuel to the debate, and as a partial answer to a furnaceman's statement that in melting metal, and in spite of the utmost care, an occasional heat of metal will be gassed, otherwise known as a "wild heat." What, then, can be done to eliminate the gas and how much gas can be eliminated?

It is the belief among foundry furnacemen that commercially gas-free metal can be produced in the foundry by melting metal under a neutral or slightly oxidizing atmosphere.

Men operating the oil-fired Schwartz furnace in this foundry endeavor to melt the metal rapidly, under a slightly oxidizing flame, at the same time avoiding an excessively high temperature. The metal is removed from the furnace as soon as possible after the proper temperature has been reached. This procedure has been largely successful in producing a metal low in gas.

It is believed by many that gas will escape from molten metal under proper conditions. Professor Cecil Desch, of Sheffield University, England, made the following statements regarding his observations on the gas content of liquid metal: "Unlike other liquids, molten metals dissolve more gas as the temperature rises, so that the more the metal or the alloy is overheated before casting, the more gas will be dissolved; and so such solutions easily remain supersaturated. An overheated metal usually will contain more gas at the moment of pouring than one that has been heated only to proper pouring temperature."

Gas Given Off

He also states: "In some cases the gas would be given off on cooling, before the casting temperature was reached, and it was only in the case of rapid cooling that one would expect to get any supersaturated solution."

In the melting and casting of aluminum bronze, Professor H. C. H. Carpenter, of Imperial College of Science and Technology, England, made the following statement: "After the metal had been poured and cleared of its dross, it would be allowed to remain in the crucible as long as possible before pouring. With the fall of temperature the solubility of gas diminishes. Corresponding to this there is a liberation of dissolved gas."

Gas in non-ferrous alloys may be eliminated in large part by holding the gas-contaminated metal for a sufficient period at a temperature above its melting point in an atmosphere relatively free of the contaminating gas. Experiments also indicate that a drop in temperature during the holding period is not necessary to secure elimination of a large part of the gas.

Oil-Fired Schwartz Furnace Observations. Early in the summer of 1942 a large, covered, well insulated thermal ladle holding 1000 lb. of metal was put into use at the "Yard" foundry. Experiments were made in removing all of the metal from the Schwartz furnace at one time, at the same time keeping the metal sufficiently hot so that all would pour.

It was observed on occasional heats that the risers on the first molds poured would mushroom, and that the risers on the last molds would have a good shrink. This indicated that the gas probably was escaping from the metal during the 10 to 15 min. required to pour the ladle of metal.

On some occasional gassed heats another phenomenon was observed. When light castings and heavy castings were poured at the same time, the risers on the light castings would shrink, and the risers on the large castings would mushroom. This indicated that a mass effect, on the metal solidifying rate, is a factor in the liberation of gas.

Electric Furnace Observations. When melting a copper-nickel alloy in an indirect-arc, rocking electric furnace, and making no effort to control the gas content since the metal was to be remelted, the following observations were made:

Power was turned off when the metal was at the proper temperature. The first ladle of metal tapped

NOTE: The opinions or assertions contained herein are those of the author and are not to be construed as official or reflecting the views of the Navy Department or the Naval Service at large. Presented at a Brass and Bronze Session of the Fiftieth Annual Meeting, American Foundrymen's Association at Cleveland May 8, 1946.

was wild with violent effervescence, and the ingots into which the metal was poured mushroomed badly. The second ladle, with only a slight effervescence, produced ingots with only a slight mushrooming.

Shrinkage

The third ladle was quiet and produced ingots with a good shrink. This would indicate that the gas was escaping during the time that the metal was held in the furnace. The total time from the moment the power was turned off to the finish of pouring was about 10 min.

In taking samples of monel metal from the electric furnaces the following observations were made. When two small ladles of the same size were used simultaneously to dip metal from the furnace, the metal in the ladle made of graphite would solidify rapidly, forming a dense test button with a good shrink.

The metal in the well-dried ladle made of fire clay would solidify more slowly, forming a test button with little if any shrink, and often would mushroom. From this it may be concluded that a metal containing some gas will produce a satisfactory casting if solidification is sufficiently rapid.

Test Casting. The following castings were adopted as the standard test castings:

The test casting called the large test casting was a truncated cone 6 in. high of 3¾-in. diameter at the top and 2-in. diameter at the bottom. The cone had a poured open top and was covered with dry charcoal to promote feeding. This arrangement should produce sound metal in the bottom half of the casting if the metal is gas free. A small, flat bar casting, 2x1x½ in., poured vertically and called the small test casting was also tested.

Gassing Metal. The metal was melted in a small, gas-fired crucible furnace, the atmosphere of which could readily be made oxidizing or reducing. It was possible to produce gassed metal a large part of the time by melting and overheating in a reducing atmosphere. This method was somewhat uncertain because sometimes the metal, when poured into the large test casting, would not mushroom.

In order to produce badly gassed metal consistently, the following procedure was adopted. When the metal was melted a slightly damp lump of fire clay was floated on top of the liquid metal for about 3 min., by which time the metal had absorbed sufficient gas so that it would mushroom badly when poured into the large test casting.

Test Procedure. A heat of hydraulic bronze taken at random from the Schwartz furnace was tested for gas elimination, using the difference in specific gravity as a measure of the gas eliminated. Metal was taken from the metal stream coming from furnace, entering thermal ladle.

Large test casting *A* was poured at 2200° F. The shrink on test casting was normal. The fracture of this casting was slightly spongy and had many scattered golden spots, and the specific gravity was 8.71.

Large test casting *B* was poured at 2090° F. after holding metal in the ladle for 12 min. The shrink on test casting was normal. The fracture was slightly spongy with some scattered golden spots, and the specific gravity was 8.79.

Regular castings were poured 5 min. after the metal entered the ladle. None of the risers of the regular castings mushroomed. The metal made satisfactory castings.

Several 35-lb. heats of metal were melted in the gas-fired, laboratory crucible furnace, and gassed or not as the occasion demanded. The first test casting was poured. The remainder of the metal was treated and the second and third castings poured. The castings then were sectioned and examined for differences.

Samples

Cross sections and fractures were taken from the similar locations in each set of castings. The etching solution used on the specimens was ammonium persulphate. For complete details, see the accompanying photographs, photomicrographs and descriptions of tests.

Shrinkage Porosity Pattern Due to Lack of Feeding in Hydraulic Bronze Casting. To observe the difference in properties between poorly-fed and well-fed castings, a 35-lb. heat of hydraulic bronze was melted in a gas-fired, laboratory crucible furnace, under a slightly oxidizing flame. The temperature of the metal in the furnace did not exceed 2200° F.: pouring temperature, 2100° F.

Two truncated cone test castings (Figs. 1 to 8) were poured out of the same crucible of metal, and with as short a lapse of time as possible between pourings. To prevent feeding, the top of casting (Fig. 1) was frozen as soon as possible by pouring water on top of the casting.

Feeding

To promote feeding, the top of casting (Fig. 2) was covered with dry charcoal as soon as possible.

To observe the porosity pattern, the castings were split vertically. The specific gravity was determined on a half cross section taken 2 in. from the bottom. The cross section was broken so that the fracture could be examined. Specific gravities of the half cross sections shown in Figs. 1 and 2 were 8.82 and 8.88, respectively.

Gas Porosity Pattern. The gas porosity pattern generally produced in hydraulic bronze is similar to that shown in Figs. 9 and 10. The photograph (Fig. 9) is of a cross section of a riser just above the casting. The metal from which this casting was poured was melted in an oil-fired Schwartz furnace, and evidently was gassed in melting, because all of the risers on the castings poured from this heat of metal mushroomed.

Similar castings poured from another heat of metal did not have mushroomed risers. The molds were made from the same sand, by the same molder. The gas porosity pattern often produced in manganese bronze is shown in Figs. 11 and 12.

Figure 11 shows a cross section of a riser just above the casting. The metal in this casting was melted in an oil-fired Schwartz furnace. The metal was gassed accidentally by pouring into a newly lined ladle which was not thoroughly dry. The risers on castings poured from the same heat of metal, but from a different ladle, did not mushroom.

A marked difference in gas porosity patterns produced in the different alloys is evident.

Gas Elimination in Aluminum Bronze. To determine the effect of time and temperature on the elimination of gas from molten aluminum bronze, the following experiment (Figs. 13 to 16) was undertaken.

A 35-lb. heat of aluminum bronze

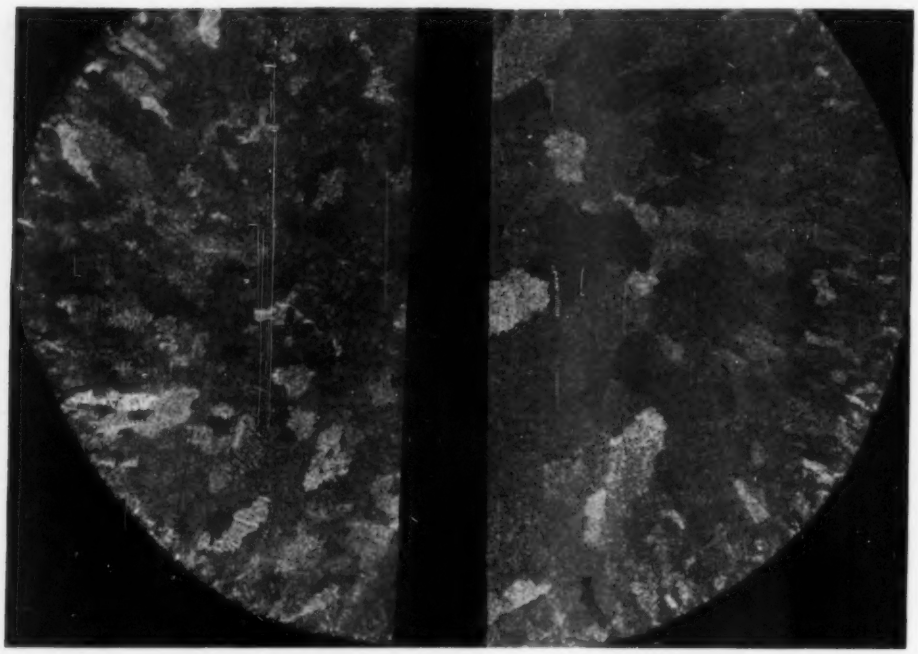


Fig. 1 (left)—A half cross section of the poorly fed casting, showing considerable fine porosity in center. Strongly etched with ammonium persulphate. About 2X.
Fig. 2 (right)—A half cross section of the well fed casting, showing little if any porosity. Strongly etched with ammonium persulphate. About 2X.



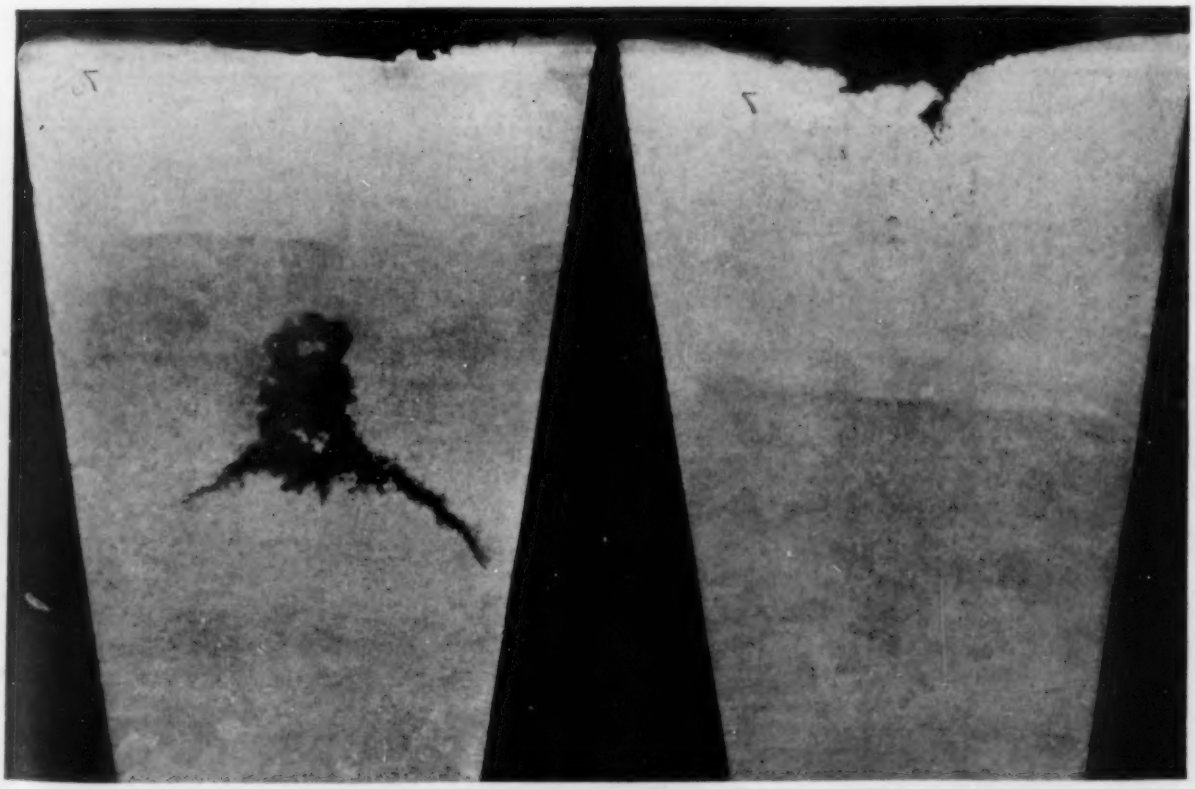
Fig. 3—An enlargement of the half cross section of the poorly fed casting (Fig. 1) showing considerable porosity. Strongly etched with ammonium persulphate. About 25X.

Fig. 4—An enlargement of the half cross section of the well fed casting (Fig. 2), showing little if any porosity. Strongly etched with ammonium persulphate. About 25X.



Fig. 5 (left)—A fracture of the half cross section of the poorly fed casting (Fig. 1). The fracture was spongy and had a color range from blue-gray through golden to dark brown. About 1¼X. Fig. 6 (right)—A fracture of the half cross section of the well fed casting (Fig. 2). The fracture was solid and dense and was blue-gray in color. About 1¼X.

Fig. 7 (left)—A vertical section of the poorly fed casting, showing gross porosity. Slightly etched with ammonium persulphate. About ⅞X. Fig. 8 (right)—A vertical section of the well fed casting, showing lack of gross porosity. Slightly etched with ammonium persulphate. About ⅞X.



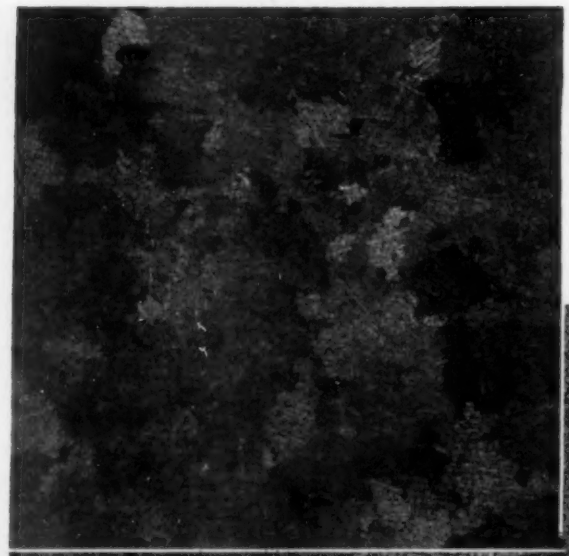
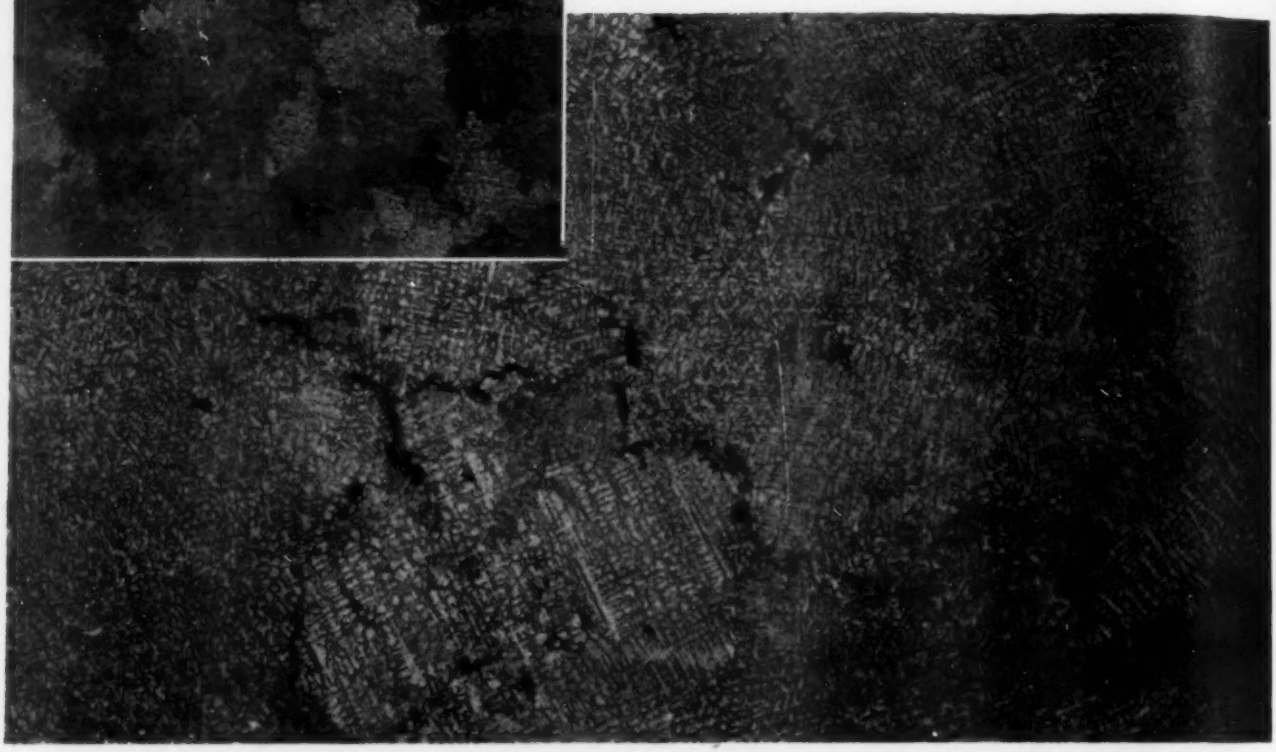


Fig. 9 (top, insert)—Cross section of a riser, just above the casting, that mushroomed badly. The metal hydraulic bronze was melted in an oil-fired furnace. Cross section strongly etched with ammonium persulphate. About 1½X. Fig. 10—Enlargement of the cross section shown in Fig. 9. Strongly etched with ammonium persulphate. About 25X. Note that porosity is largely intergranular, long tortuous interconnected voids. Few spherical gas holes have formed.



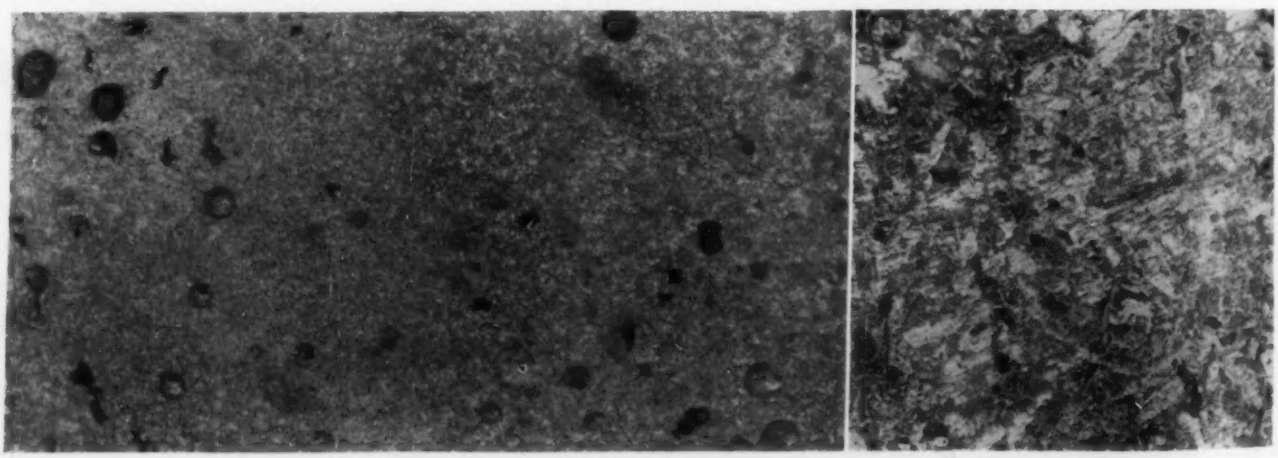


Fig. 11 (left)—Cross section of a riser that mushroomed badly. The metal manganese bronze melted in an oil-fired furnace. Cross section strongly etched with ammonium persulphate. About 1½X. Metal accidentally gassed by pouring into a new ladle that had not been thoroughly dried. Fig. 12 (right)—Enlargement of cross section shown in Fig. 11. Strongly etched with ammonium persulphate. About 25X. Note that porosity is largely spherical with little if any tortuous interconnected voids.

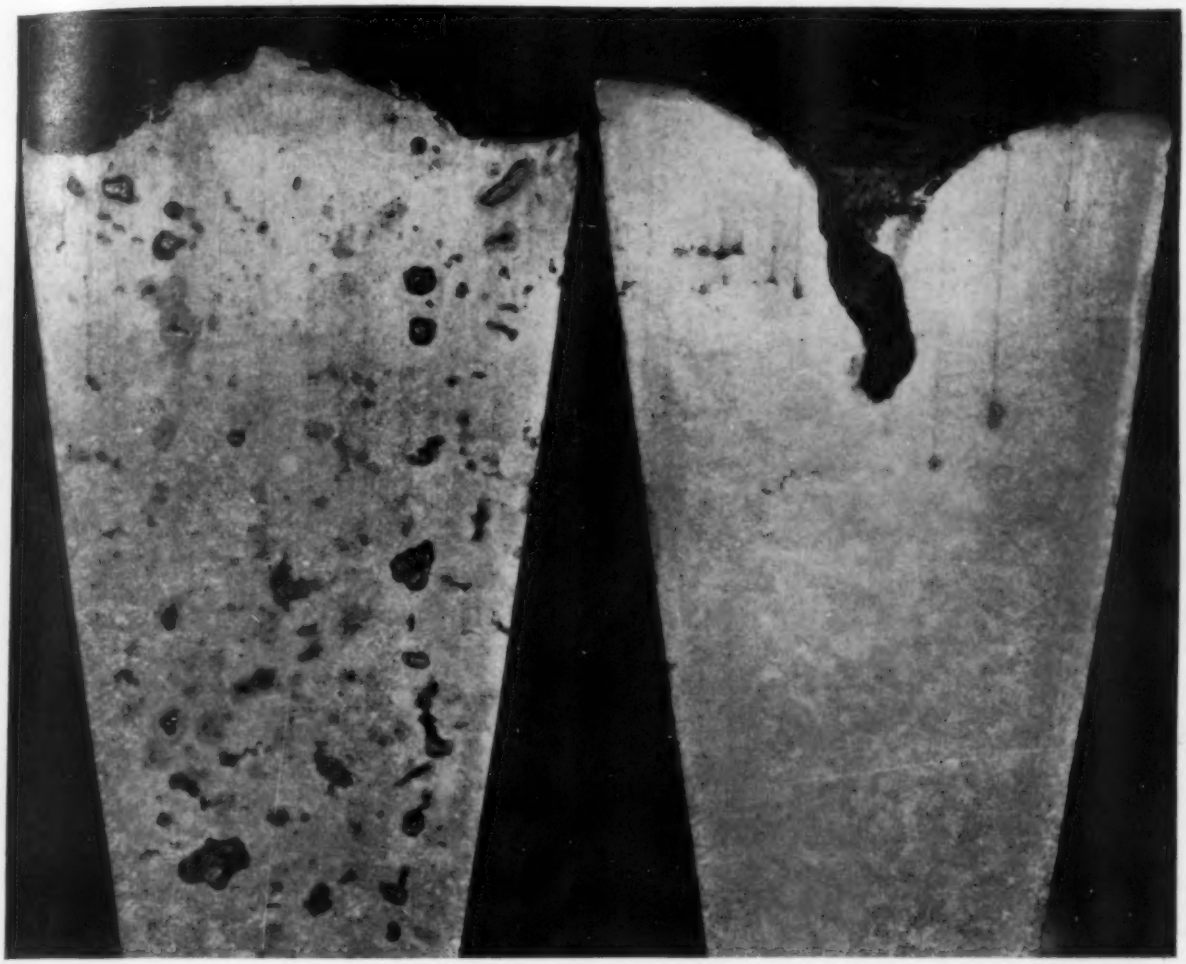


Fig. 13 (left)—Vertical section of the test casting poured at 2200° F. immediately after gassing. Lightly etched in ammonium persulphate. About 7/8X. Fig. 14 (right)— Vertical section of the third test casting poured at 1980° F. Lightly etched in ammonium persulphate. About 7/8X. Note marked reduction in porosity.

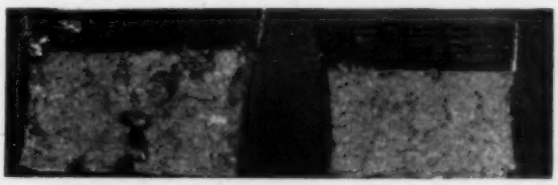


Fig. 15 (left)—Fracture of the half cross section of the badly gassed casting. Fracture was coarse and spongy. Interiors of the holes were a bright yellow color. About 1¼X. Fig. 16 (right)—Fracture of the half cross section of the third casting poured at 1980° F. Fracture had an even color distribution and showed little if any gas porosity. About 1¼X.

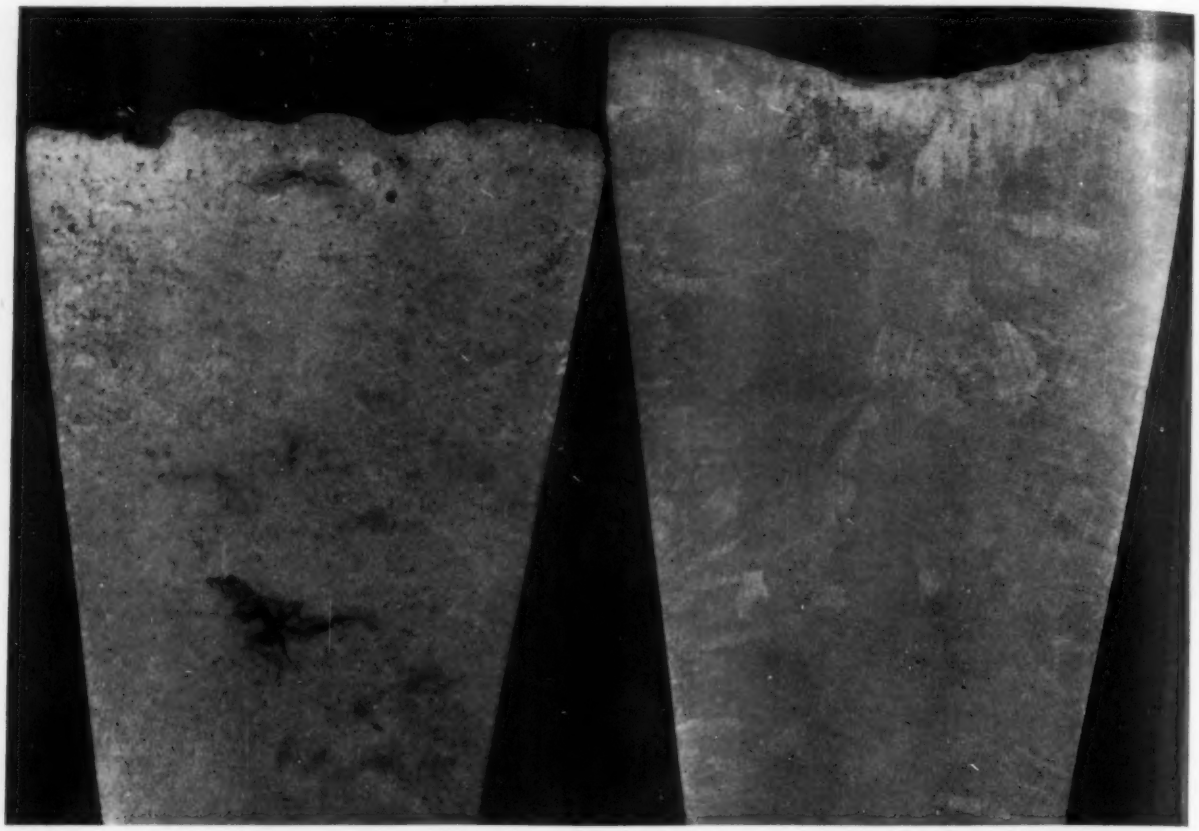


Fig. 17 (left)—Vertical section of the first test casting poured at 2360° F. immediately after gassing. Lightly etched with ammonium persulphate. About 7/8X. Fig. 18 (right)—Vertical section of the third test casting poured at 2050° F. Lightly etched with ammonium persulphate. About 7/8X. Note marked reduction in porosity.

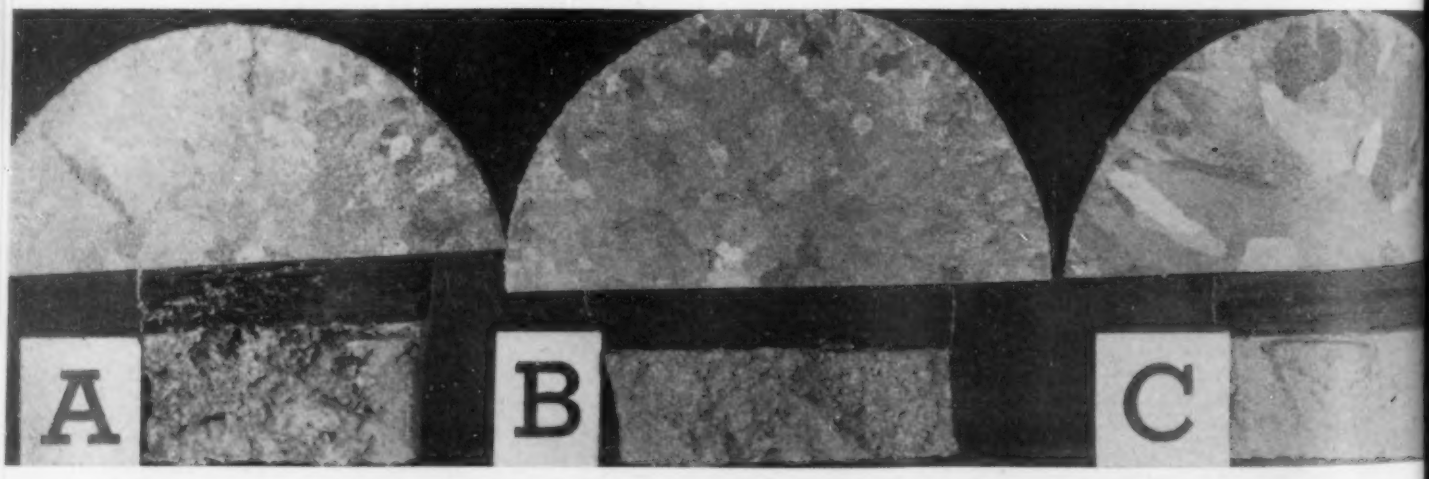


Fig. 19A—Half cross section and fracture from casting (Fig. 17). The half section shows considerable porosity. Fracture was coarse, spongy, and badly discolored. Specific gravity—8.13.

Fig. 19B—Half cross section and fracture from casting (no photograph of vertical section). Half section shows considerable porosity. Fracture was coarse and discolored. Specific gravity—8.44.

Fig. 19C—Half cross section and fracture from casting (Fig. 18). The half section showed little if any porosity. Fracture was a fine blue-gray. About 1¼X. Specific gravity—8.82.

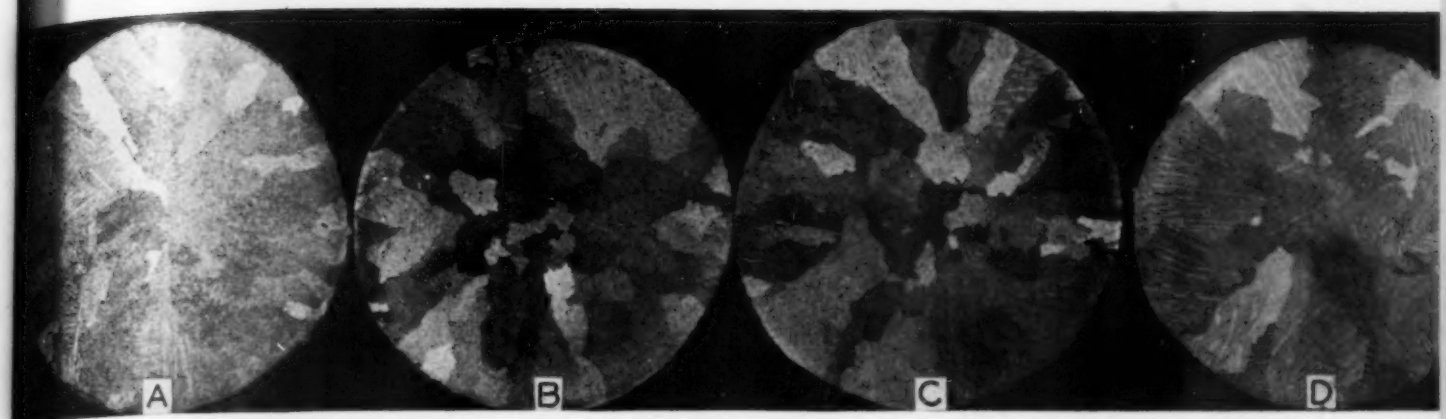


Fig. 20A-B-C-D—Cross sections of test castings. Strongly etched with ammonium persulphate. About 1X. Note gradual diminution of porosity from A to D.

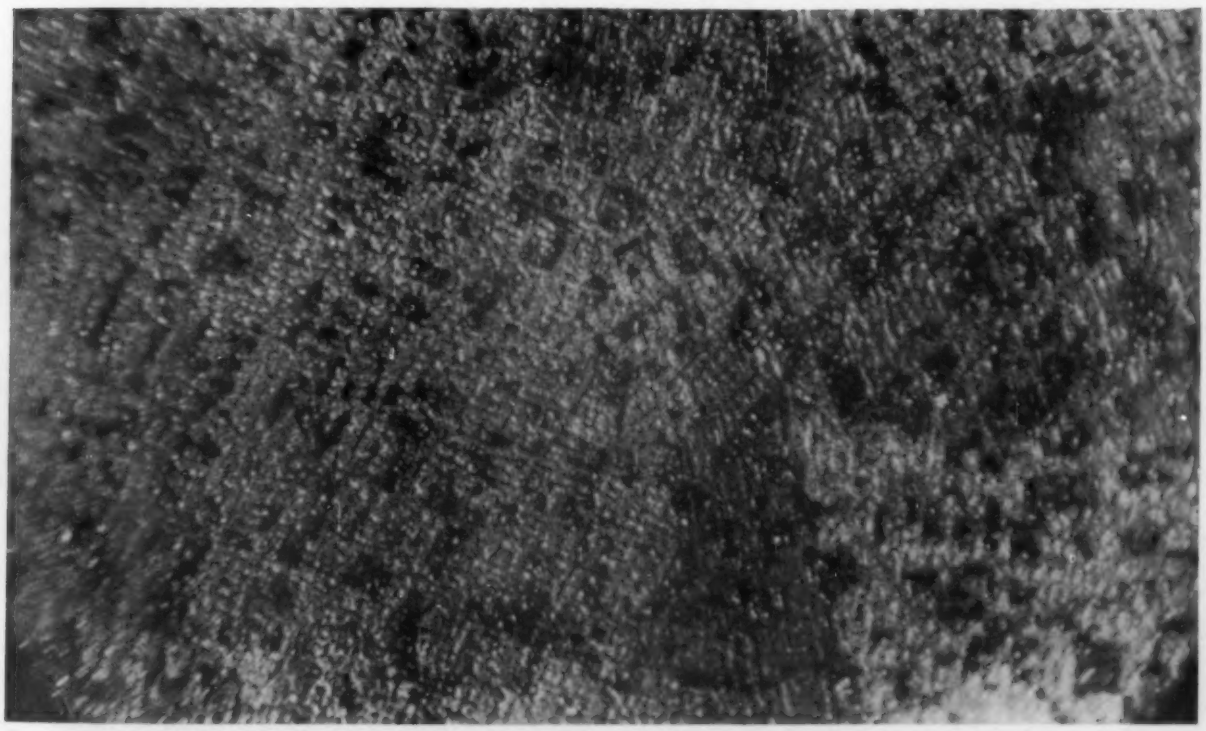


Fig. 21—Enlargement of the cross section of first test cone (Fig. 20-A) which mushroomed badly. Strongly etched with ammonium persulphate. About 25X. Porosity is distributed within the grains and at grain boundaries. All the dark areas are voids.

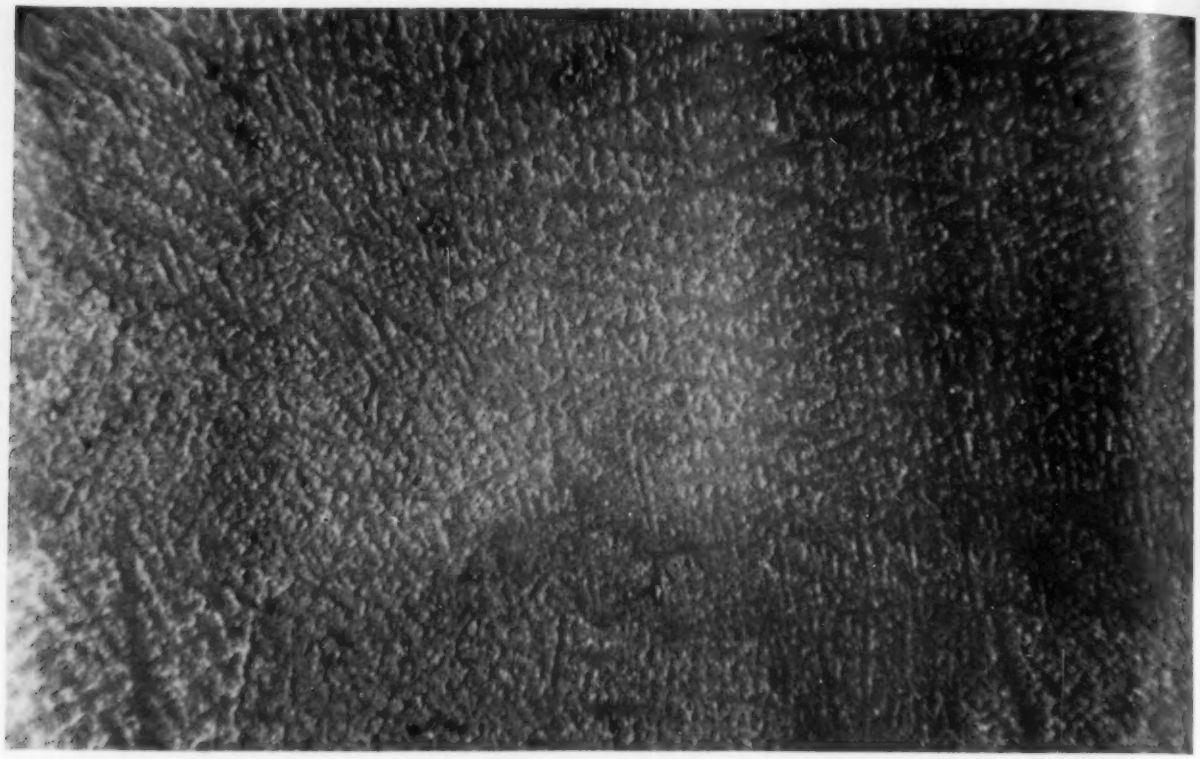


Fig. 22—Enlargement of the cross section of the fourth test cone (Fig. 20D) which had a good shrink. Strongly etched with ammonium persulphate. About 25X. Marked reduction in porosity comparing Fig. 20, A and D.

was melted in a gas-fired, laboratory crucible furnace. The metal was brought to a temperature of 2230° F. and gassed by floating a lump of damp clay on the surface of the metal for 3 min.

The first test cone (Fig. 13) was poured immediately after gassing. Pouring temperature was 2200° F. The casting mushroomed badly. The remainder of the metal was returned to the furnace and held for 5 min. with the furnace fuel turned off.

Effect of Furnace Cooling

The second test cone was poured at 2100° F. The casting mushroomed badly. No photographs were taken of this casting because it appeared similar to that shown in Fig. 13. The remainder of the metal was returned to the furnace and held for 6 min. with the furnace fuel turned off.

The third test cone (Fig. 14) was poured at 1980° F. after a total holding time of 11 min. The casting had a normal shrink.

In order to observe the porosity the castings were split vertically. The specific gravity was determined on a half cross section taken 2 in. from bottom of the casting. The half cross section was broken so that the fracture could be examined. Specific gravity of the test casting shown in Fig. 13 was 6.91, and that of casting shown in Fig. 14, 7.61.

A large part of the gas evidently will escape from molten aluminum bronze if the time is sufficient, and with the possible assistance of a drop in temperature.

Gas Elimination in Hydraulic Bronze. To determine the effect of time and temperature on the elimination of gas from molten hydraulic bronze, the following experiment was undertaken.

A 35-lb. heat of hydraulic bronze was melted in a gas-fired, laboratory crucible furnace. The metal was brought to a temperature of 2400° F. and gassed by floating a lump of damp clay on the surface of the metal for 3 min.

The first test cone (Fig. 17) was poured at a temperature of 2360° F. immediately after gassing. The casting mushroomed badly. The balance of the metal was returned to the furnace and held 4 min. with the furnace fuel turned off.

The second test cone (no photographs) was poured at a temperature of 2240° F. The casting mushroomed less than that shown in Fig. 17. The remainder of the metal was returned to the furnace and held for 10 min. with the furnace fuel turned off.

The third test cone (Fig. 18) was poured at a temperature of 2050° F. after a total holding time of 14 min. The casting had a normal shrink.

Observations

To observe the porosity the castings were split vertically. The specific gravity was determined on a half cross section taken 2 in. from the bottom of the casting. The half cross section was broken so that the fracture could be examined. Specific gravities of half cross sections

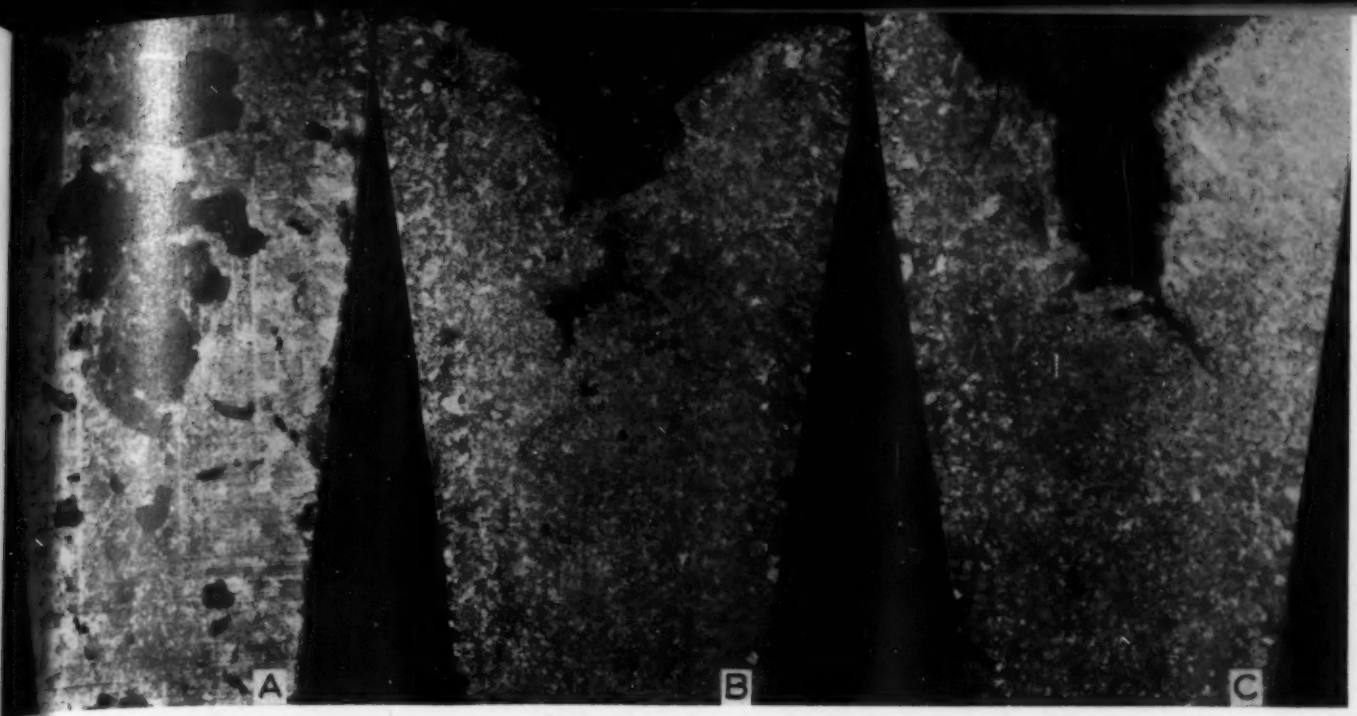


Fig. 23A—Vertical section of the first test casting poured at 2775° F. Strongly etched with ammonium persulphate. About 7/8X. Note gross porosity and mushrooming.

Fig. 23B—Vertical section of the second test casting poured at 2795° F. Strongly etched with ammonium persulphate. About 7/8X. Note fine scattered porosity and the shrinkage.

Fig. 23C—Vertical section of the third test casting poured at 2790° F. Strongly etched with ammonium persulphate. About 7/8X. Note absence of visible porosity. Markedly more shrinkage than in Fig. 23B.

of the three test cones (Fig. 19, left to right) were 8.13, 8.44 and 8.82.

A large part of the gas evidently will escape from molten hydraulic bronze if the time is sufficient, and with assistance of a temperature drop.

Gas Elimination in Monel Metal (various pouring temperatures). To determine the effect of time and temperature on the elimination of gas from molten monel metal, the following experiment was undertaken.

A 200-lb. heat of monel metal pig was melted in a 350-lb. capacity indirect-arc, rocking electric furnace. The metal was brought to a temperature of 3050° F. The first test cone (Fig. 20-A) was poured immediately at a temperature of 3000° F. The casting mushroomed badly. The remainder of the metal was held in the furnace for 5 min. with

Fig. 24A—Half cross section taken 2 in. from bottom of first test casting (Fig. 23A) poured at 2850° F. Strongly etched with ammonium persulphate. About 1¼X. Note the gross porosity and almost complete absence of fine porosity.

Fig. 24B—Half cross section taken 2 in. from bottom of second test casting (Fig. 23B) poured at 2775° F. Strongly etched with ammonium persulphate. About 1¼X. Note fine porosity and some reduction of gross porosity.

Fig. 24C—Half cross section taken 2 in. from bottom of third test casting (Fig. 23C) poured at 2795° F. Strongly etched with ammonium persulphate. About 1¼X. No gross porosity and only a small amount of fine porosity concentrated in central area of casting.



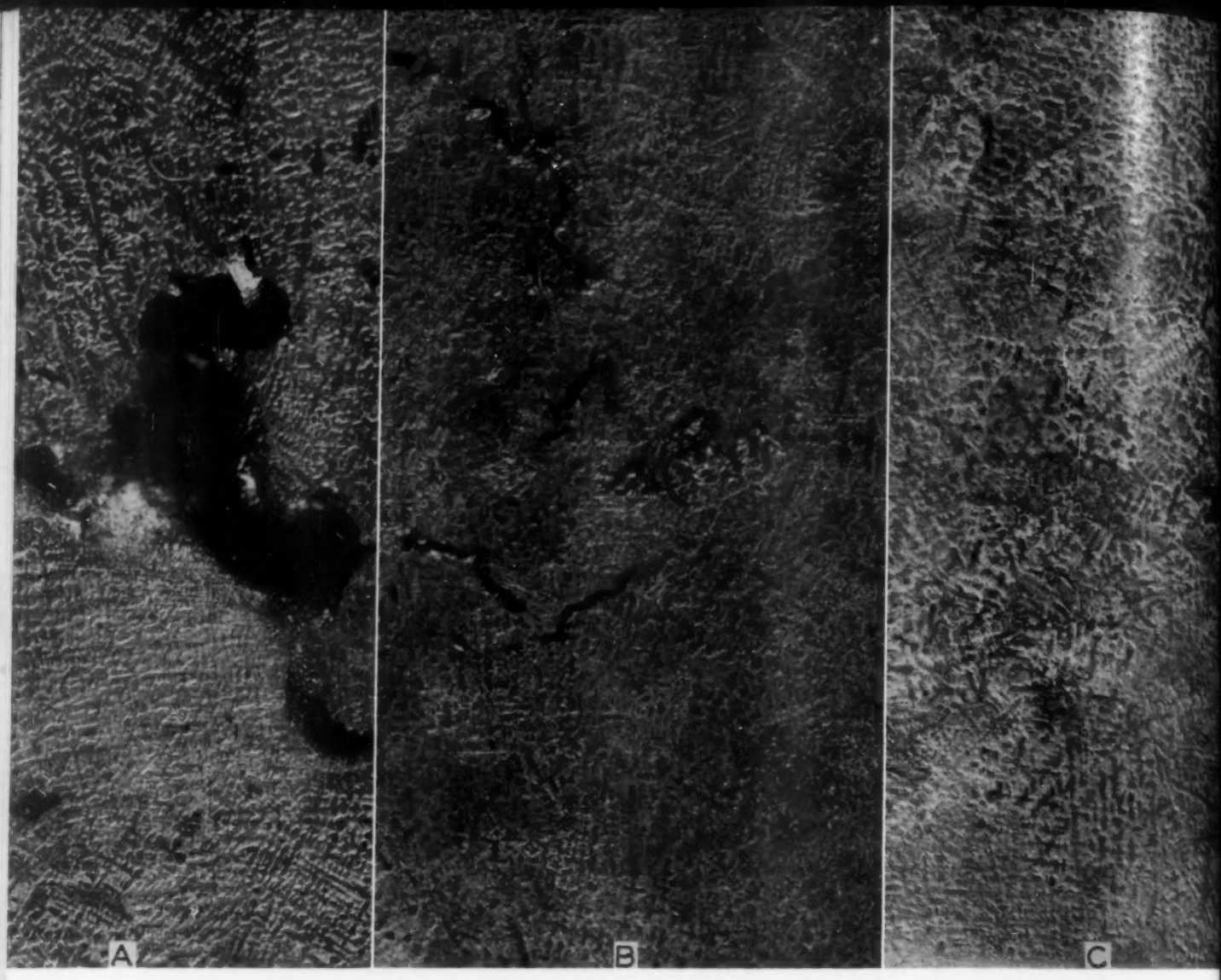


Fig. 25A—Enlargement of half cross section (Fig. 24A). About 16X. Considerable gross porosity. The dark dendritic patterns are not voids.

Fig. 25B—Enlargement of the half cross section (Fig. 24B). About 16X. Considerable gross porosity but finer than that shown in Fig. 24A. Voids are the elongated tortuous dark areas.

Fig. 25C—Enlargement of the half cross section (Fig. 24C). About 16X. Porosity is of the fine interdendritic tortuous type. Voids are the fine thin elongated tortuous dark areas.

all the furnace power turned off.

A second test cone (Fig. 20-B) was poured. The casting mushroomed less than the casting shown in Fig. 20-A. The remainder of the metal was held in the furnace for an additional 5 min. with power turned off.

A third test cone (Fig. 20-C) was poured. The casting mushroomed less than that of Fig. 20-B. The remainder of the metal was held in the furnace for an additional 5 min. with power turned off.

A fourth test cone (Fig. 20-D) was poured at a temperature of 2710° F. The casting had a good shrink.

Temperature Control

Total holding time was 15 min. The total drop in pouring temperature was 290° F. The optical pyrometer was used in determinations.

Cross sections were taken 2 in. from the bottom of each casting and etched in ammonium persulphate. Specific gravities of cross sections A and D of Fig. 20 are 8.04 and 8.69. Figures 21 and 22 are enlargements of A and D sections of Fig. 20.

A large part of the gas evidently will escape from molten monel metal if the time is sufficient, and with the possible assistance of a drop in temperature.

Gas Elimination in Monel Metal (constant temperature). To determine the effect of time, while holding the temperature constant, on the elimination of gas from molten monel metal the following experiment was undertaken.

A 35-lb. heat of monel metal pig was melted in a magnesia crucible in an electric induction furnace. The metal was brought to a temperature of 2850° F. and gassed by floating a lump of damp clay on the surface of the metal for 15 min. Magnesium as a deoxidizer was added at rate of 1½ oz. per 100 lb. of metal.

The first test cone (Fig. 23-A) was poured immediately after adding the magnesium at a temperature of 2775° F. The casting mushroomed badly. The top of the casting was rough and drossy. The balance of the metal was held in the furnace for 10 min. with just sufficient power on to maintain the temperature.

A second test cone (Fig. 23-B) was poured at a temperature of 2795° F. The casting had a marked

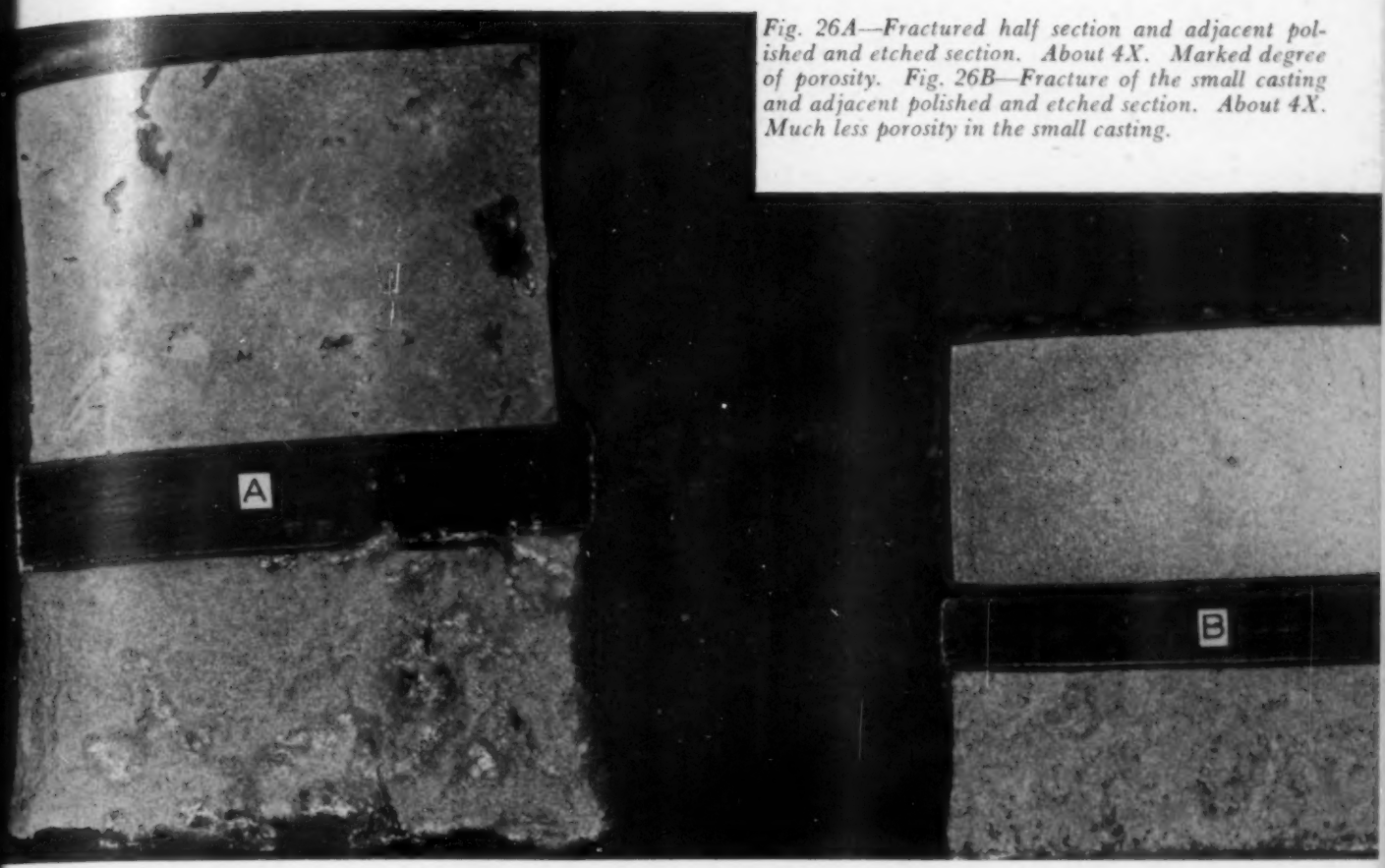


Fig. 26A—Fractured half section and adjacent polished and etched section. About 4X. Marked degree of porosity. Fig. 26B—Fracture of the small casting and adjacent polished and etched section. About 4X. Much less porosity in the small casting.

shrink. The top of the casting was free from dross. The balance of the metal was held in the furnace for an additional 10 min. with just enough power on to maintain the temperature.

The third test cone (Fig. 23-C) was poured at a temperature of 2790° F. The casting had a marked shrink, larger than that shown in Fig. 23-B. The top of the casting was free from dross.

Total holding time was 20 min. There was a temperature increase of 20° F. An optical pyrometer was used to determine temperatures.

To observe the porosity the castings were split vertically. Specific gravities determined on half cross sections taken 2 in. from bottom of casting (Fig. 24-A, B, C) were 8.273, 8.715 and 8.774, respectively. Figure 25 shows enlargements of the sections shown in Fig. 24.

A large part of the gas evidently will escape from molten monel metal if the time is sufficient at a constant temperature, and with the possible assistance of the stirring action in the furnace.

Mass Effect on Gas Elimination in Aluminum Bronze. To determine the effect of the solidifying rate on the elimination of gas from molten aluminum bronze, the following experiment was undertaken.

A 35-lb. heat of aluminum bronze was melted in a gas-fired, laboratory crucible furnace. The metal was brought to a temperature of 2200° F. and gassed by floating a lump of damp clay on the surface of the metal for 3 min. A mold was poured at a temperature of 2120° F., making the large truncated cone and the small flat bar, 2x1x½ in. (Fig. 26-A, B). The cone solidified slowly and the small bar solidified rapidly. The cone (Fig. 26-A) mushroomed badly. The small bar (Fig. 26-B) had a shrink in the top.

Porosity Observed

To observe the porosity in the castings, the cone was split and a half cross section was taken 2 in. from the bottom of the casting. The half section was broken, so that the fracture could be examined. A section adjacent to the fracture was polished and etched. The specific gravity was determined on the half cross section.

The small casting was broken transversely at about the center. A section adjacent to the fracture was polished and etched. The specific gravity was determined on the bottom half of the casting, that of the cone (Fig. 26-A) being 7.10, and that of the small casting (Fig. 26-B) 7.60.

With a metal of a given gas content more porosity can be expected in heavy castings than in light castings.

Mass Effect on Gas Elimination in Hydraulic Bronze. To determine the effect of the solidifying rate on the gas elimination from molten hydraulic bronze, the following experiment was undertaken. A 35-lb. heat of hydraulic bronze was melted in a gas-fired laboratory crucible furnace.

The metal was brought to a temperature of 2450° F. and gassed by floating a lump of damp clay on surface of molten metal for 45 sec.

A mold was poured at a temperature of 2300° F., making the large truncated cone (Fig. 27-A), and the small flat bar (Fig. 27-B), 2x1x½ in. The cone solidified slowly and the small bar solidified rapidly.

The remainder of the molten metal in the crucible was held for 3½ min., until the temperature dropped to 2000° F., at which time

the same kind of a mold was poured, making a large truncated cone (Fig. 27-D), and the small flat bar (Fig. 27-C).

To observe the porosity in the castings the cones were split and half cross sections were taken 2 in. from the bottom. The half sections were broken, so that the fractures could be examined. Sections adjacent to the fractures were polished and etched. The specific gravity was determined on the half cross sections.

Castings Broken Transversely

The small castings were broken transversely, at about the center. Sections adjacent to the fractures were polished and etched. Specific gravities were determined on the bottom halves of the castings, being 8.35, 8.60, 8.75 and 8.51, respectively, for the sections shown in Fig. 27-A-B-C-D.

The large cone (Fig. 27-A) mushroomed badly. The fracture was spongy and badly discolored. The polished section had considerable porosity.

The small bar casting (Fig. 27-B) had a slight shrink. The fracture was spongy and discolored, but less so than the fracture on section A. The polished section had much less porosity than the polished section A.

The large cone (Fig. 27-D) had a slight shrink. The fracture was slightly spongy and discolored, but less so than section A. The polished section had considerably less porosity than the polished section of section A.

The small bar casting (Fig. 27-C) had a good shrink. The fracture was slightly spongy, much less so than the fracture on section B. The polished section had much less porosity than the polished sections B and D.

Indications are that a large part of the gas can be eliminated from metals by holding the gas-contaminated metal for a sufficient time above its melting point in an atmosphere relatively free of the contaminating gas.

It is not necessary to have a drop in temperature during the time lapse to eliminate a large part of the gas.

A not unusual occurrence in a foundry is to pour a heavy casting and a light casting from the same ladle of metal, and to have the risers on the heavy casting mushroom and risers on the light casting shrink.

Melter vs. Molder

This brings forth a dogmatic statement from the furnaceman: "The molds are too wet, rammed too hard, or sand too fine." The molder retorts with: "The metal is burned and full of gas."

A possible explanation is that the small casting solidified rapidly enough to hold a large portion of the gas in solution so that no porosity resulted. The heavy casting solidified slowly enough to allow the gas to migrate and concentrate in the last metal to solidify, with resulting porosity and mushrooming of risers. The above mentioned mold conditions could also cause the metal to mushroom.

DISCUSSION

Chairman: D. FRANK O'CONNOR, American Saw Mill Machine Co., Hackettstown, N. J.

Co-Chairman: HERMAN SMITH, Federated Metals Div., American Smelting and Refining Co., Pittsburgh, Pa.

STANLEY R. KEITH[1]: How is the magnesium put into the monel metal? What manner was used in getting it into the molten metal?

MR. DALBEY: We introduce the magnesium into the metal by putting the required amount of stick magnesium in the crook on one end of a half inch steel rod. The crook containing the magnesium is then plunged into the metal. It takes considerable dexterity on the part of the furnace operator to get the magnesium into the metal, and not have it burn on the surface of the molten metal.

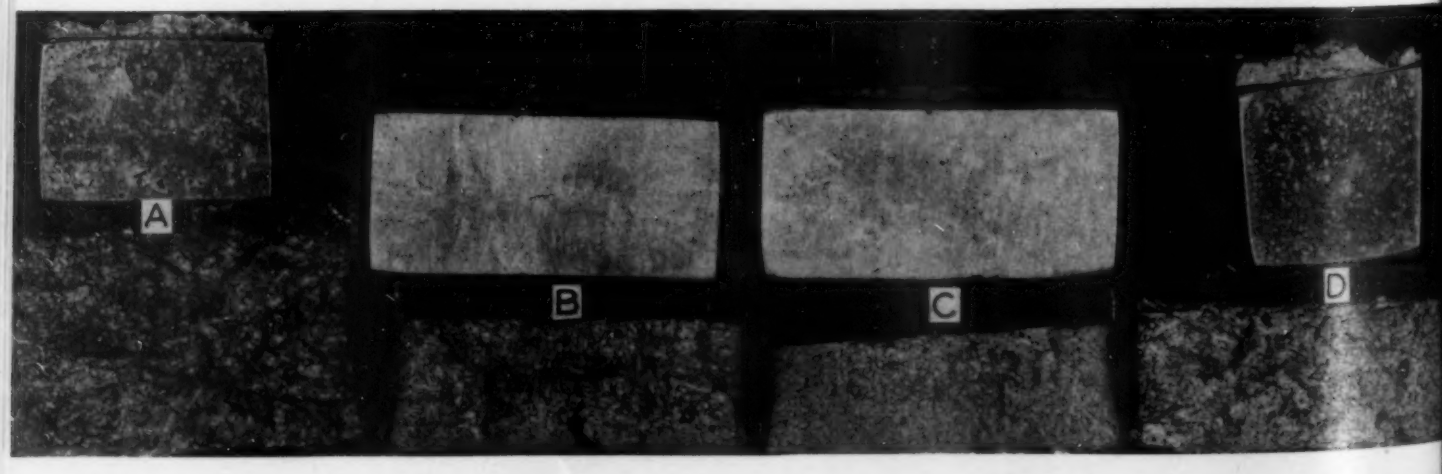
[1]Submergent Alloys Corp., New York, N. Y.

Fig. 27A—Fracture and adjacent polished sections from truncated cone test casting. Fracture was badly discolored.

Fig. 27B—Fracture and adjacent polished section from small flat bar casting. Fracture was less discolored than cone casting (Fig. 27A).

Fig 27C—Fracture and adjacent polished section from small flat bar casting. Fracture only slightly discolored, less than that of the cone casting (Fig. 27D).

Fig. 27D—Fracture and adjacent polished section from truncated cone test casting. Fracture was less discolored than that shown in Fig. 27A.



PREVENTING VEINING AND PENETRATION ON CASTINGS MADE IN SYNTHETIC SANDS

R. E. Morey
and
J. R. Kattus

Veins and penetration are surface defects which frequently occur on castings produced in synthetically bonded sand. They are frequently found on lead and tin bronze castings, and occasionally on iron and steel castings. To find methods for preventing their formation, an investigation was made to determine the mechanism of veining and penetration. Phosphor-bronze was used in experimental work because the defects are most prevalent in castings made with this alloy.

Veining of bronze castings is caused by (1) sweating of the metal, which is an extreme case of inverse segregation, and (2) cracking of the mold or core. In the late stages of solidification dissolved gases come out of solution and force the low-melting-point constituent of the alloy toward the surface of the casting through interdendritic passageways. Frequently, this constituent exudes from the surface of the casting and enters cracks in the mold or core to form veins. Cracking of molds and cores is believed to be caused by the thermal expansion characteristics of silica. Penetration defects occur when the exuded metal enters interstices among the sand grains.

Elimination of most veining on bronze castings can be accomplished by proper melting practice because this tends to reduce the amount of dissolved gases in the molten metal. Sand techniques which help to eliminate veining by preventing the mold from cracking were developed. Washes containing finely divided refractories stop penetration by filling intertices in the surfaces of molds and cores.

The mechanism of veining on gray iron castings is believed to be similar to that of veining on bronze castings, but it is believed that veins on steel castings form only at locations where hot tears occur.

INCREASED USE of synthetic sands throughout the foundry industry is a convincing indication that their advantages over natural sands in many cases are widely recognized. In order to fill the need of advanced bases, repair ships, and other naval establishments for simplicity in sand practice, a synthetic molding sand and a core sand were developed at the Naval Research Laboratory and recommended for use in making molds and cores for most of the commonly used foundry alloys.

The molding sand mixture consists of washed silica sand with a median size of 180 microns (corresponding approximately to A.F.A. grain size of 80) plus additions of 3.0 per cent bentonite, 1.5 per cent dextrine, 0.5 per cent corn flour, and 3.5 per cent moisture. A core sand employing the same base sand bonded with 1.0 per cent core oil, 1.0 per cent corn flour, and 4 per cent moisture was recommended.

It was realized that the base sand is coarser than normally would be used in making a synthetically bonded sand for use with non-ferrous metals. In order to use one sand for all metals and thus eliminate the necessity of maintaining stocks of two or more kinds of sand on repair ships and at advanced bases, it was felt that a slight sacrifice in surface finish could easily be made. This compromise on grain size permitted the use of the single grade of washed silica sand in the mixture to be used for all metals which normally are required for emergency ship repair.

With few exceptions, sound engineering castings of all the commonly cast metals except magnesium have been produced consistently in these sands. However, difficulty often was experienced in producing bronze castings free of the surface defects known as veining and penetration. These flaws are rarely found when natural sands are used. Veins are extraneous fins of metal which sometimes protrude at various places on the unfinished casting.

Penetration is the name applied

Presented at a Sand Research Session of the Fiftieth Annual Meeting, American Foundrymen's Association, at Cleveland, May 8, 1946. Published with permission of the Navy Department without endorsement of statements and opinions of the authors, R. E. Morey and J. R. Kattus, Lt. (j.g.) U.S.N.R., Steel Castings Section, Division of Physical Metallurgy, Naval Research Laboratory, of the Office of Research and Inventions.

to surface roughness produced where molten metal is forced into the interstices between the sand grains for a considerable distance. It varies in severity from slight roughness of the casting surface to extreme cases where the metal has penetrated a distance of an inch or more. These extreme cases, where sand grains are completely surrounded by metal, are frequently called "burnt-on sand."

Preventing the occurrence of these defects is a matter of importance since castings are frequently scrapped because of them or, at best, time and equipment are required to remove them. Therefore, an investigation was conducted to find, if possible, the causes of veining and penetration and means for preventing their occurrence.

Observations Regarding Veining and Penetration. While veining has been known to occur in many of the sand casting alloys, the most serious cases are found in certain lead and tin bronzes. Of the ferrous metals, gray cast iron has the greatest tendency to vein, but veins are occasionally found on steel castings. Aluminum and magnesium alloys have the least tendency to form veins.

Veining is often confused with penetration. This confusion arises because the two defects frequently occur simultaneously. However, differences have been noted at the Naval Research Laboratory and also have been recorded in the literature[1].

That molds and cores sometimes crack when suddenly heated is a fact known to most foundrymen. This is attributed to dimensional instability caused by thermal stresses in the sand resulting from temperature gradients. Veins form when molten metal flows into such cracks. They form most frequently at "hot spots" in the sand such as re-entrant angles, but occasionally are found on flat surfaces. Penetration occurs when the metal is fluid enough to flow into interstices between the sand grains.

Penetration Causes

Coarseness, soft ramming, and poor flowability of the sand are the main causes of penetration. In most cases it may be eliminated by the use of a wash in which a refractory of fine particle size is used as the base. This prevents penetration by filling the interstices between the sand grains but it does not stop the mold from cracking when hot metal is poured into it and hence has little influence on veining. Figure 1 shows a phosphor bronze casting which illustrates this fact. Half of the mold for this casting was painted with silica flour wash and the other half was untouched. The line of demarcation can clearly be seen where penetration was prevented by the wash, but veining was not retarded.

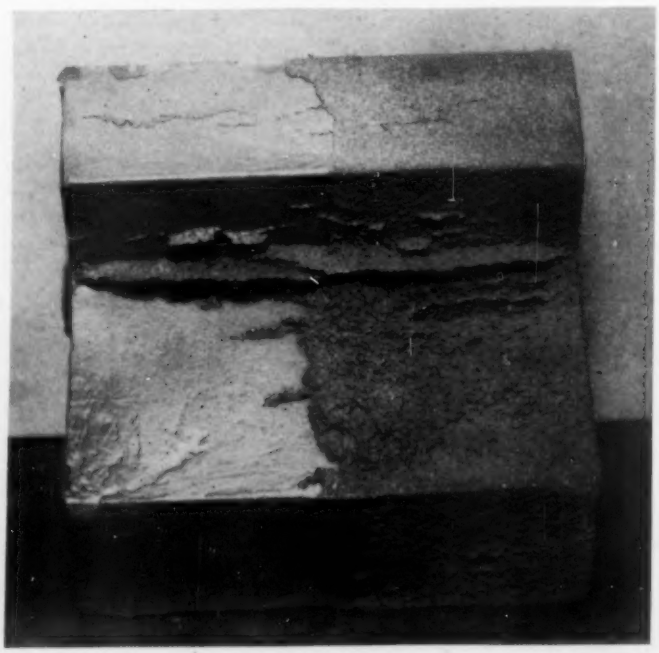


Fig. 1—Phosphor bronze casting. Left half of mold was painted with a silica flour wash, the other half was not.

Suggested Mechanisms of Veining. No satisfactory explanation of the mechanism which causes veining has ever been proposed. For a crack or fissure to be produced into which metal can flow, the surface of the sand must be in tension. A compressive stress only tends to bond the sand into a more compact mass. When a specimen of rammed sand is heated in a furnace, a temperature gradient exists from the surface to the interior of the body, and the surface expands more rapidly than the interior to produce compression in the surface and tension in the interior.

If the furnace is controlled at some constant temperature below 573° C. (1063° F.), which is the inversion point of alpha quartz to beta quartz, the surface of the specimen comes to the furnace temperature and expands a definite amount. At some later time, the center of the specimen also reaches the furnace temperature and expands an amount equal to the expansion at the surface of the specimen. When this condition is reached, the temperature gradient is gone and the stress gradient also disappears.

Simple heating, therefore, will not produce the necessary tensile stress to cause veining. Conditions existing during cooling can be neglected because it has been observed that, at elevated temperatures, cracks develop in test specimens during the heating period. These facts indicate that some other conditions must exist for veining to occur.

Granular Sand

Since the sand is a granular material of low strength, it is conceivable that some adjustment takes place in the grain positions or the clay layer which relieves the compressive stress at the surface during the early stages of heating. A temperature gradient then exists with little or no stress gradient, as shown in *C* of Fig. 2. This condition persists as long as the temperature con-

tinues to rise provided the surface does not reach 573° C. (1063° F). However, as the temperature rises the sand is developing hot strength to become a unified mass instead of a granular material.

Then when the sand at the surface reaches a temperature of 573° C. (1063° F.) it suddenly stops expanding and begins to contract slightly, as shown by the expansion curve for sand in Fig. 3, while the sand in the interior of the specimen is still expanding because it has not reached the inversion temperature.

The slight contraction at the surface accompanied by expansion in the center tends to create tensile stresses at the surface. This is believed to be the condition which initiates the cracks which, in a mold, permit the molten metal to enter the sand and form veins. As the isothermal surface at 573° C. (1063° F.) progresses inward, the depth of the crack is gradually increased.

Another suggestion which has been made is that the organic material in the sand mixture burns in a manner similar to that in which a piece of paper contracts when it burns. If a specimen is exposed to an elevated temperature in the dilatometer for a period of not over 4 min. and is then removed and broken open, it will be found that there is a burned or black layer of sand on the surface of the specimen while the center remains unchanged.

Cracking

This might be expected to cause a change in length of the surface layers and cause some tension which would produce cracking. However, this theory is discredited by tests upon sillimanite and periclase refractories which indicate that the nature of the refractory as revealed by its characteristic curve probably is responsible for the forces which starts cracks in the specimen.

Experimental Procedure. Experience in pouring all types of metal at the Naval Research Laboratory has shown that the worst instances of veining are consistently found on phosphor bronze castings. This alloy is known to foundrymen for its property of "attacking the sand." Veining and penetration often disfigure the castings beyond repair. The use of phosphor bronze in experiments on veining produces the most severe test of the sands used; therefore, any means of stopping veining in this alloy would do at least as well as other copper base alloys. For this reason bronze of the following composition was used in all tests: 89.5 per cent copper, 10 per cent tin, and 0.5 per cent phosphorus.

A simple test block (illustrated in Figs. 4 and 5) was designed so that three different sand mixtures could be tested in one casting. Three standard 2x2-in. test specimens,

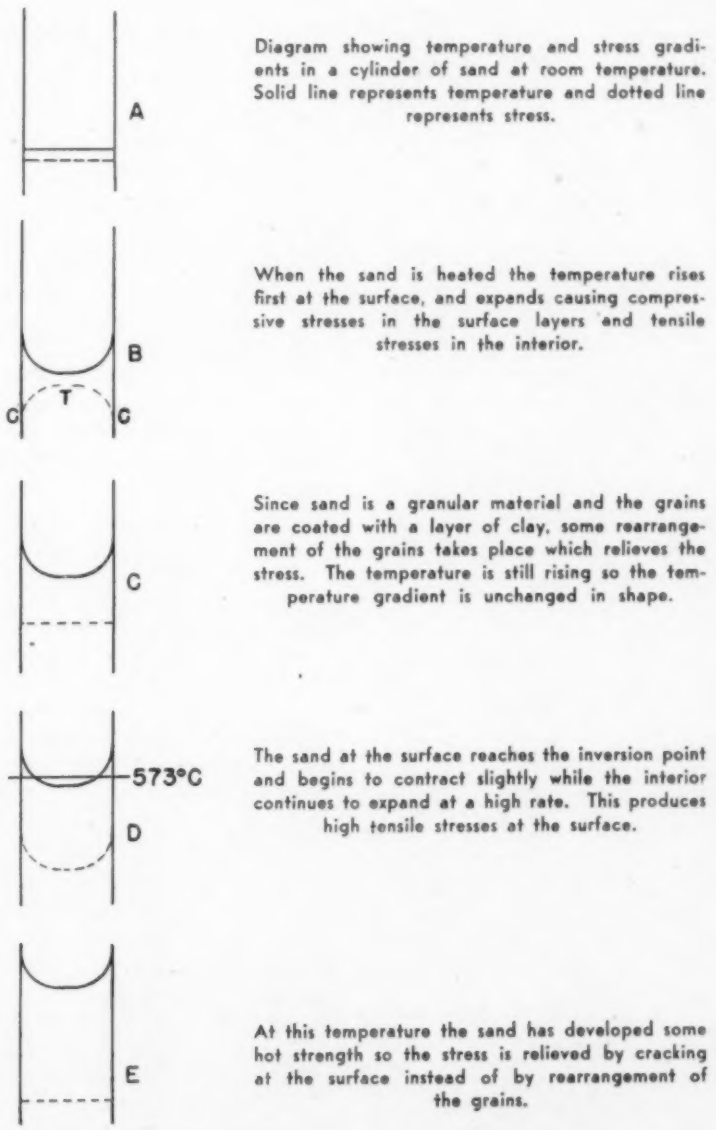


Fig. 2—Diagram shows temperature and stress gradients in a cylinder of sand.

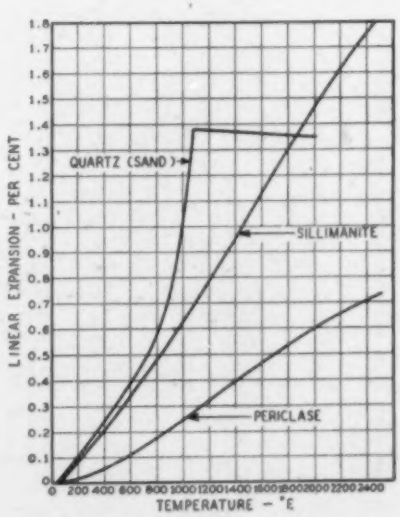


Fig. 3 (right)—Expansion curves for sand, sillimanite, and periclase.

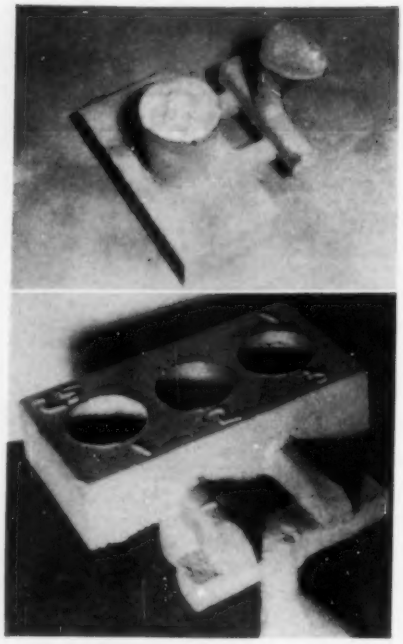


Fig. 4 (top)—Photo shows type of casting used in these tests. Fig. 5 (bottom)—Same as Fig. 4.

made from three different mixtures, were inserted into the bottom of the mold so that one-half of each specimen projected into the molten metal. This type of test is one of the most severe possible, since the sand under test is nearly surrounded by metal and the heat flow is restricted.

Most mixtures tested were mulled in 2000-gram batches in a 15-lb. capacity laboratory muller. The regular molding sand and core sand, which were used for the purpose of comparison, were taken from sand bins in the foundry after they had been mulled in a 200-lb. production muller. Specimens made of clay-bonded molding sands were dried at 105° C. (220° F.), and those made of core sands were baked at their normal baking temperatures before being placed in test molds.

Induction Furnace Used

All heats were melted in an induction furnace. After the data from each heat were recorded, the same metal was remelted for use in subsequent tests.

Test Results. If the first possible mechanism stated in the foregoing is a valid explanation for the cracks which form on mold surfaces as their temperatures are raised, refractories which have smooth expansion curves should be free from cracks and thus prevent veining when used as molding materials. Two such materials, namely, periclase (MgO) and sillimanite ($Al_2O_3.SiO_2$) were tested. Their expansion curves are shown in Fig. 3. Samples of sand and of these materials having equivalent particle size were bonded similarly and tested in experimental castings. Figure 6 shows the results obtained. From left to right the cores were sand, sillimanite, and periclase. Both periclase and sillimanite stopped veining completely; thus the foregoing theory is given support.

Ability of periclase and sillimanite to undergo shock heating without cracking is illustrated in Fig. 7. These three dilatometer specimens were subjected to a temperature of 1093° C. (2000° F.) for a period of 3 min. Both periclase and sillimanite maintained smooth, continuous surfaces, but large cracks were developed in the sand. The specimens all had been coated with silica wash in order to show the cracks more clearly. Most of the sand mixtures used in these experiments were subjected to this test. It was found that, in general, the degree of veining of the test casting was related to the amount of cracking developed by the shock-heating test.

Earlier in this paper it was stated that veining is most serious in lead and tin bronzes, while other metals show much less tendency to form veins. From some heats of lead and tin bronzes, castings are made which are free from veins, although the chemical composition and pouring temperature apparently are the same as those of similar heats from which castings with many veins are produced. This indicates the existence of some property characteristic of each heat which determines whether or not veining occurs.

Early in these experiments it was noted that in most heats the phenomenon of tin sweating occurred. Sweating is an extreme case of inverse segregation in which the low melting point constituent of an alloy exudes from the surface of the casting late in solidification. Figure 8 shows a typical example of tin sweat which has exuded from the top of a riser. The fact was soon established that the serious cases of veining are associated with this phenomenon because invariably the heats that sweated were the ones in which the bad cases of veining occurred. Two castings which illustrate this fact are shown in Figs. 9 and 10.

In both cases the chemical compo-



Fig. 6—Test casting shows results produced by test cores of (left to right) sand, sillimanite and periclase.

Fig. 7—Test cores after being subjected to a shock heat test of 2000° F.



Table 1

COMPOSITION OF VEIN, EXUDED METAL AND BODY OF CASTING

	Composition, per cent		
	Cu	Sn	P
Veins	84.2	14.3	1.5
Exuded Material	79.0	17.3	2.0
Body of Casting	89.8	10.3	0.6

sition of the heats, pouring temperature, and type of molding sand were the same. The tin sweat in the casting shown in Fig. 9 was pronounced, while in the casting shown in Fig. 10 it was completely absent. It will be noted that veining in the casting which sweated is many times worse than in the one in which sweating did not occur.

Veining and Sweating

In order to further associate veining with sweating, a sample of material which was exuded from the top of a riser and actual veins which were chipped from castings were analyzed. The results, given in Table 1, show that both the veins and the exuded material are high in tin and phosphorus and, therefore, are low-melting-point constituents.

As a final experiment to prove that sweating is responsible for veining, two identical castings were poured from the same heat. One was shaken from the mold 4 min. after pouring, and the other was allowed to cool in the mold. The casting which was removed from the mold while hot had solidified sufficiently to retain its shape, but no veins had formed. However, shortly thereafter, sweating began to take place with liquid metal exuding from numerous places on the casting. The casting that cooled in the mold formed many veins. These castings are shown in Figs. 11 and 12.



Fig. 8—Top of riser shows low-melting constituent which has exuded (tin sweat).

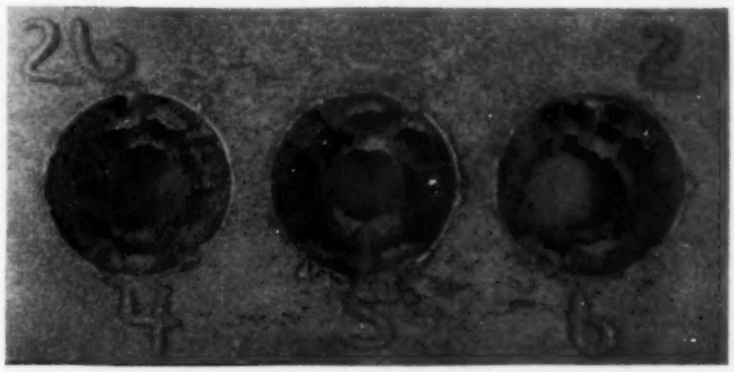


Fig. 9—Phosphor bronze casting on which tin sweat was pronounced; veining associated with this phenomenon.

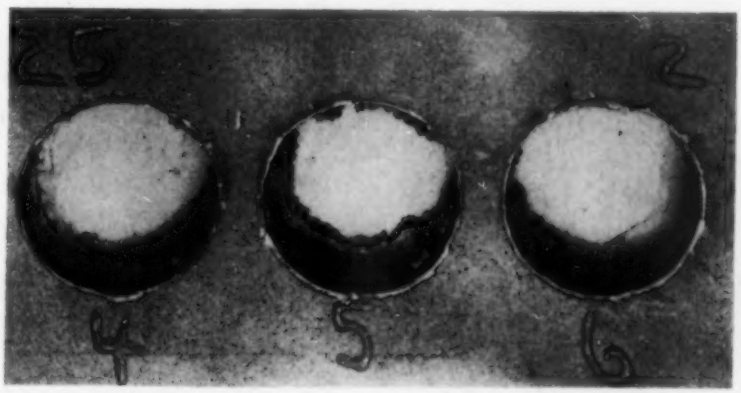


Fig. 10—Phosphor bronze casting on which tin sweat was absent.

Sweat Beads

Many beads of sweat are visible on the casting shown in Fig. 11, which was shaken out while hot. The obvious conclusion may be drawn that metal also exudes from the casting shown in Fig. 12 but, having no other place to go, was forced into cracks in the sand and thus formed veins. It may be noted that the surface of the bottom casting is much rougher than that of the top casting. Thus it appears that the exuded metal also enters interstices among the sand grains and causes penetration.

Many theories have been proposed to explain inverse segregation[6,7,12]. Data obtained in these experiments on phosphor bronze support that of Gender[7], who states that gases which are dissolved in the liquid metal come out of solution during the later stages of solidification and force the low-melting-point constituent toward the surface through interdendritic passageways.

Density determinations were made on six test castings from different heats, three of which sweated and veined whereas the other three did not. The chemical composition, pouring temperature, and position of the samples for density determina-

Table 2

DENSITY DETERMINATIONS

Density, grams per cc.	
Castings with tin sweat	Castings with no tin sweat
7.974	8.351
8.165	8.375
7.975	8.329

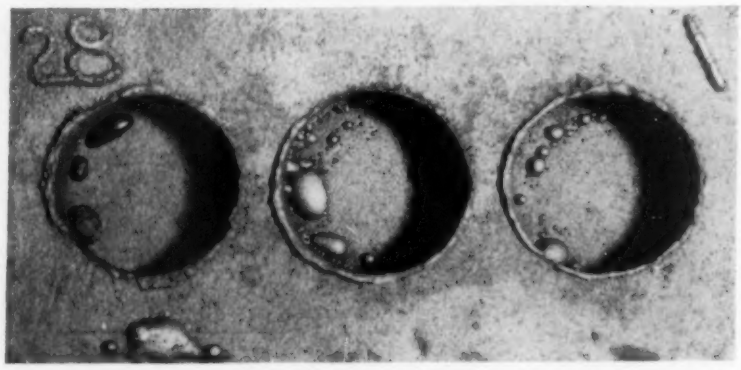


Fig. 11—Casting shaken out from mold 4 min. after it was poured. Note globules of sweated metal but absence of veining.

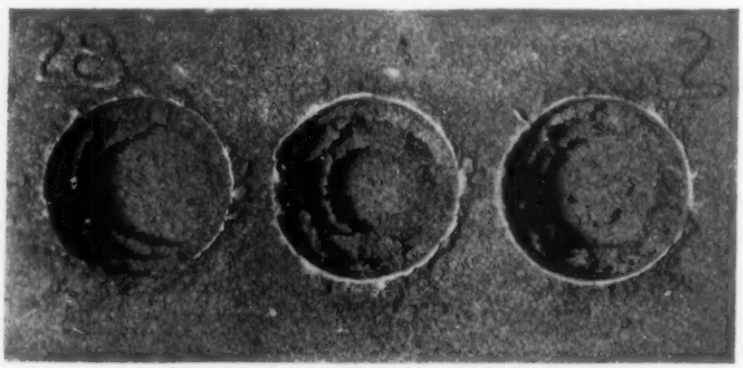


Fig. 12—Casting from same heat as casting in Fig. 11; cooled overnight in the mold before shaking out. Note veining but absence of globules of sweated metal.

tion were the same for each test.

Results given in Table 2 show that the castings which sweated and veined are considerably less dense. This can be attributed only to entrapped gases; therefore, it appears that at least for the extreme case of inverse segregation known as sweating, Gender's theory is valid.

The fact that severe cases of veining generally are found on lead and tin bronze castings can be readily explained by the theory of sweating. These alloys have wide freezing ranges, which is a condition that must exist in order for sweating to occur. This condition promotes the type of freezing in which coring and dendritism are prominent. When freezing progresses in this manner, numerous interdendritic passages form, through which the low-melting point constituent can flow to the surface.

Veins on Gray Iron and Steel Castings. It is possible that the theory of sweating also explains the veins which frequently are found on gray iron castings. This alloy has a wide freezing range, and research conducted at the Naval Research Laboratory indicates that its manner of freezing is similar to that of tin bronzes; i.e., during the initial stages of solidification, many dendrites form throughout the casting with liquid metal filling the interstices. Although the amount of gas dissolved in cast iron may be negligible, a phenomenon occurs which might have a similar effect in causing sweating. This is the formation of graphite flakes. It is conceivable that the growth of the flakes forces liquid metal to the surface in the same way that the rejection of dissolved gases causes sweating bronzes.

Veins occasionally occur on steel castings, but they differ from those found on bronze castings in both appearance and location. They are small and seldom occur at re-entrant angles. Because of the narrow solidification range of steel, its manner of solidification differs from that of tin bronzes and cast iron.

Instead of the formation of dendrites throughout the casting, the solidification of steel begins with the formation of a solid skin of metal in contact with the mold and progresses inward with a relatively sharp division between the solidified layer and the liquid metal. For these reasons, sweating and inverse segregation do not occur in steel, and veins must form by some other mechanism.

Bore Cracks

Studies of bore cracks, hot tears of a specific type, found in cast steel fittings and valve bodies[11], show that veins on steel castings frequently are found following the contour of these cracks. Experimental observations led to the explanation that hot tears form in the solid skin of metal while the interior of the casting is liquid. The sand adjacent to the casting adheres to the metal and also tears. Molten metal from the interior of the casting flows through the crack in the solid skin and into the adjacent crack in the sand to form a vein. The problem of eliminating veins on steel castings, therefore, is secondary to the problem of eliminating hot tears. If the proper precautions to prevent hot tears are observed, no veining occurs.

Evidence indicating that this theory of veining is applicable to bronzes is lacking. The experiment in which a test casting was shaken out 4 min. after being poured precludes this possibility. One of the necessary conditions for the formation of hot tears in steel castings is friction between the sand and the casting. There were no exudations from the surface of this casting until it had been out of the mold for a minute and, therefore, the casting was free of any frictional stresses imposed by the sand.

It is, therefore, reasonable to conclude that the cause of exudations was not hot tearing. The exudations, which took place in this experiment, occurred at numerous scattered points, and it is not likely that they were connected by cracks. Microscopic and macroscopic examinations revealed no cracks in the vi-

cinity of veins on the bronze castings.

Prevention of Veining by Proper Melting. The evidence indicating that the formation of veins on bronze castings is one of the effects of sweating and that this is caused by gassiness makes it apparent that melting practice is an important factor to be considered. If gases can be excluded from the melt, veins seldom are formed.

Cast Bronze Research

Melting practice and gas porosity of cast bronzes have been the subjects of much research[8,9,10]. Although there was no systematic investigation of them made during these experiments, the results obtained suggest a few of the principles that should be followed to obtain good castings:

1. Heats of virgin metals should be melted, allowed to solidify, and then remelted before being cast. In general, heats made from pig or scrap are less gassy than those melted from virgin metals.
2. The use of clean, high grade raw materials is desirable.
3. An oxidizing atmosphere is best for melting.
4. If deoxidation is necessary, the minimum amount of deoxidizer should be used and it should be added just before pouring.
5. The temperature of the melt should be kept as low as is practical. The maximum temperature reached rather than the pouring temperature is most critical. Gas solubility increases with temperature, therefore the highest temperature attained determines the amount of gas dissolved. As the temperature falls, dissolved gases do not come out of solution until sufficient time elapses for equilibrium to be established. Most of the gas dissolved during melting is retained in the metal until solidification is in progress.

Influence of the Sand Mixture on Veining. Although it has been shown that sweating is responsible for the serious cases of veining, it is also true that the mold or core must crack before veins can form; therefore, the condition of the sand is an important factor in veining. Numerous sand mixtures were tested with experimental castings. It should be remembered that the test was one of the most severe possible.

The metal used has a greater tendency than any other to form veins;

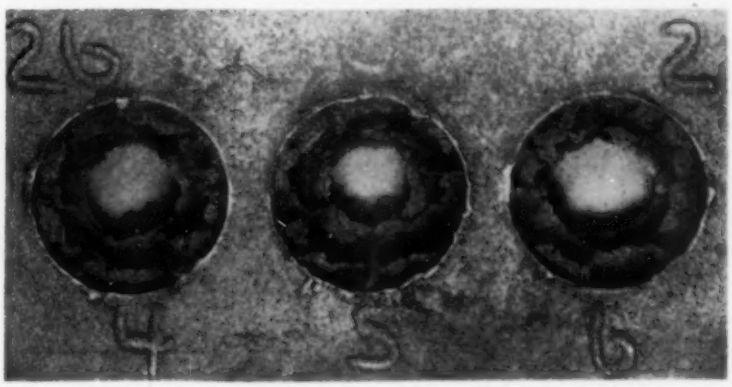


Fig. 13—Effect of different amount of core ramming.

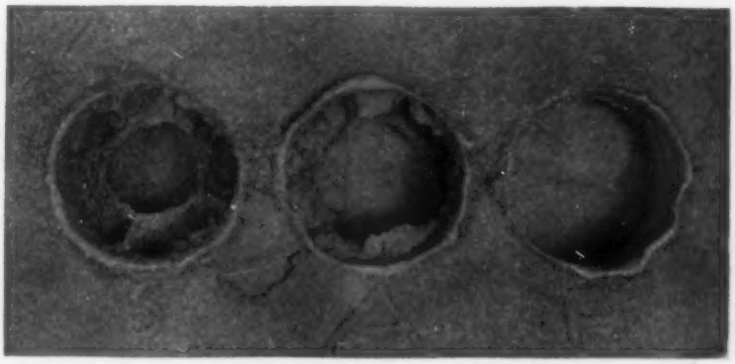


Fig. 14—Veining can be decreased by use of fine sand.

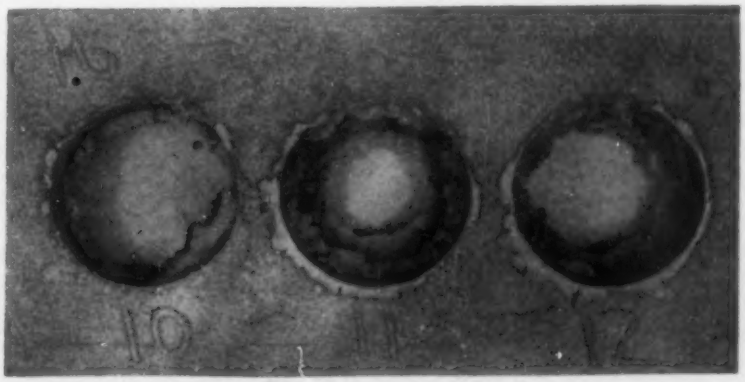


Fig. 15—Test casting shows effect of sand grain distribution.

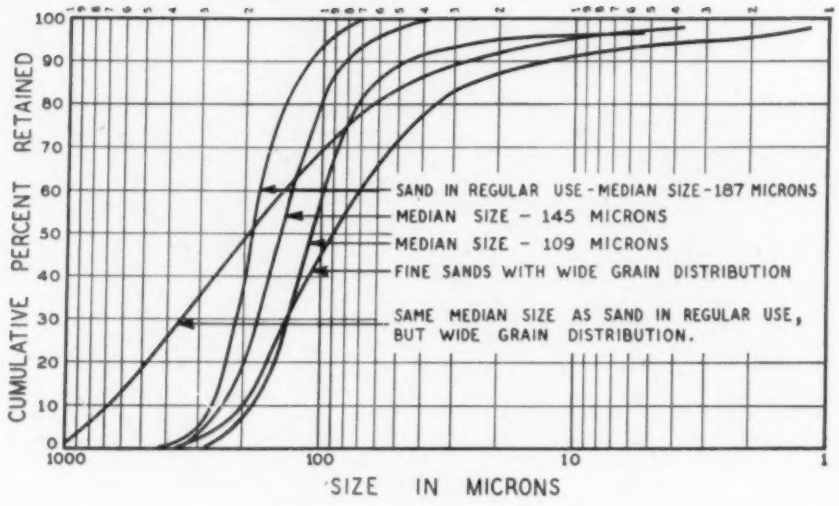


Fig. 16—Cumulative curves of sands used for castings shown in Figs. 14 and 15.

no attempt was made to employ the best melting procedure, and the shape of the test casting is one with which veins may be formed easily. Any conditions which are effective for preventing the formation of veins should be even more effective under less severe conditions. Most of the test cores were painted with silica wash in order to eliminate penetration so that the veining could be clearly seen. The pouring temperature in all tests was 2000° F.

Ramming. The effect of different amounts of ramming is negligible. This is illustrated in Fig. 13, which shows a casting in which three cores of molding sand were rammed differently. Cores No. 4 and 5 (as marked on the casting) were rammed ten times and one time, respectively, while No. 6 was made by ramming ¾ of its length nine times, then adding the remainder of the sand and ramming once again in order to form an unevenly rammed specimen.

Grain Fineness and Distribution. Veining can be decreased by the use of fine sand. This is shown in Fig. 14. The three cores used in this casting were molding sands with median sizes of 187, 145, and 109 microns, respectively, from left to right. These sizes are approximately equivalent to A.F.A. grain fineness numbers 80, 108, and 135.

Core No. 10 in Fig. 15 was a fine synthetic molding sand with a wide grain distribution; No. 11 was the regular molding sand which has a well assorted grain distribution; and No. 12 has the same median size but a wide grain distribution. The wide distribution gives slightly better results. Cumulative curves for these sands are shown in Fig. 16.

Facing Materials. Numerous facing materials and binders were tested. No correlation was found between any of the ordinary physical properties of molding sands and the degree of veining. However, one property which has a great influence on veining was discovered. This is the property of plasticity at elevated temperatures. Certain materials, when added to the sand mix, render a rammed body of the sand capable of being deformed at high temperatures without breaking. This enables the thermal stresses in the sand to relax by plastic deformation rather than by cracking, and hence prevents veining.

Wood Flour

The best material which has been found for imparting plasticity at elevated temperatures is wood flour, but it is useful only when employed in conjunction with a clay such as bentonite. It can not be used with success in core sands. Apparently, there is a fluxing reaction between the wood and the clay which produces plasticity. Wood flour increases collapsibility and ease of shake-out, but it creates gas and lowers dry strength.

Other materials which produce this property in sand are ordinary fluxes such as borax or sodium carbonate. However, these materials flux directly with the sand and produce a hard glass which makes shake-out and cleaning difficult. They also increase penetration.

In Fig. 17, the left-hand and center cores consisting of regular molding sand with additions of 5 per cent wood flour and borax, respectively, are compared with one made of regular molding sand with no additions. Some penetration can be seen where the core mixture containing borax was used, but veining was stopped completely by both wood flour and borax.

Hot Compression

The property of plasticity can best be investigated by means of a hot compression test. This consists of compressing a dilatometer specimen after soaking at temperature for 12 min. Ordinarily the specimen remains relatively rigid until a certain load is reached, and then it collapses into two or more pieces. However, mixtures containing wood flour, borax, sodium carbonate or other fluxing materials begin to flow plastically when a small load of a few psi. is applied. Thereafter, the load remains constant while the specimen is compressed to a shape with shorter length and increased diameter.

The load required to produce plastic flow and the extent to which the specimens can be compressed without cracking or breaking depends on the type of flux, soaking time, and temperature. If specimens made from mixtures containing 5 per cent borax or wood flour are heated to 2000° F. or above for 12 min., they can be compressed to less than three-quarters of their original lengths without cracking.

Other organic materials such as activated charcoal and sea coal were tested. Their influence on the sand mixture is similar to that of wood flour in that they make it possible for the dilatometer specimen to be compressed without complete failure, but they do not maintain smooth, uncracked surfaces. These materials were not as effective in preventing veining as those previously mentioned. The fact that sea coal imparts plasticity to the sand at elevated temperatures may be a clue to its usefulness in iron molding sands.

Washes With Low Fusion Points. Experience in casting steel has shown that a thin layer of the mold directly in contact with the metal

becomes viscous within a few seconds after pouring and, upon cooling, forms a glaze. When properly controlled, this glaze may be easily peeled from the casting and is instrumental in producing a smooth surface. This condition is not achieved in non-ferrous castings because of the wide difference between the pouring temperature of the metal and the fusion point of the sand.

As a possible means of preventing the formation of veins, such glazes were produced with bronze castings by the use of low-fusion-point washes. These washes were made in three ways; frits of ceramic materials, combined in such a way to have a fusion point 1800° F., were employed as the base material in a wash; silica flour wash was adulterated with fluxes such as sodium carbonate, borax, and potassium nitrate in order to lower its fusion point; and plain water solutions of such fluxes were applied to the surface of the mold.

Eliminating Veining

Ceramic frits were only slightly beneficial. Adulterated silica washes did not stop veining entirely, but were superior to the washes which contained ceramic frits. Only the method of applying water solutions of fluxes to the mold was completely effective in eliminating veining.

The reason for the superiority of this type of wash over the adulterated silica wash is its ability to penetrate further into the mold. If the water to silica ratio of the adulterated silica wash is increased it is more effective for preventing veins since it penetrates more deeply into the sand.

In Fig. 18, the left and center cores were molding sand, painted with a 12 per cent water solution of sodium carbonate. The center cores were of molding sand, painted with a 12 per cent water solution of sodium carbonate. The center core had, in addition, a coating of silica wash applied after the sodium carbonate solution was dried. On the right is the same sand covered only with silica wash.

No veins were formed in the core washed with a plain sodium carbonate solution. On the core which was coated with silica was in addition to a solution of sodium carbonate, a smoother surface was pro-

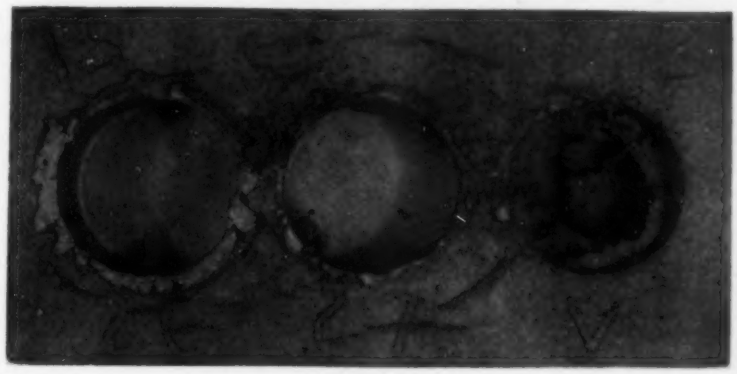


Fig. 17—Effect of sands with wood flour and borax additions compared with regular molding sand.



Fig. 18—A comparison of washes of a solution of sodium carbonate, silica flour over sodium carbonate, and regular silica flour.

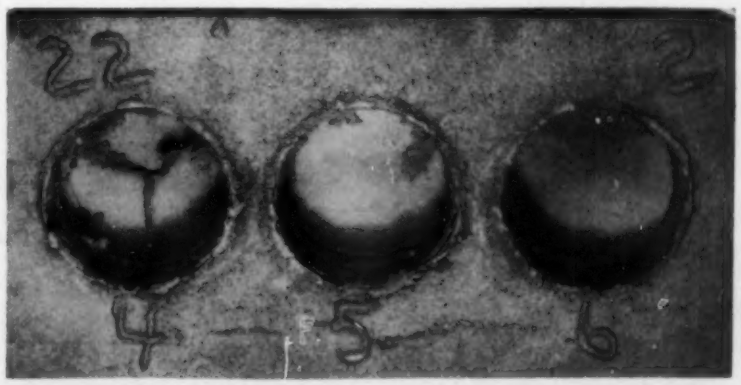


Fig. 19—Casting shows the effect of using molding sands containing 1 per cent, 3 per cent, and 10 per cent bentonite with no cereal binders.

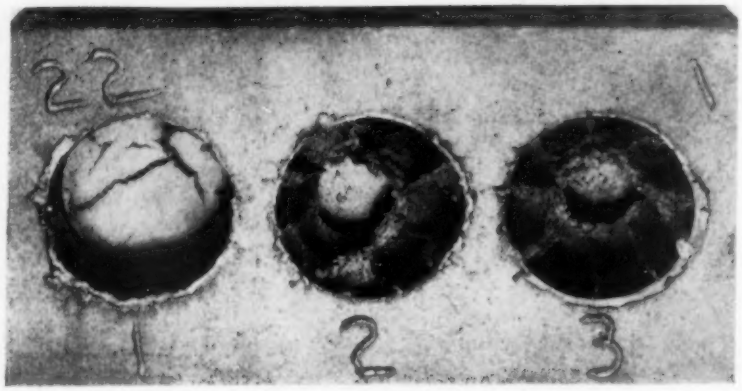


Fig. 20—Casting shows the effect of core sands with increasing amounts of core oil and corn flour.

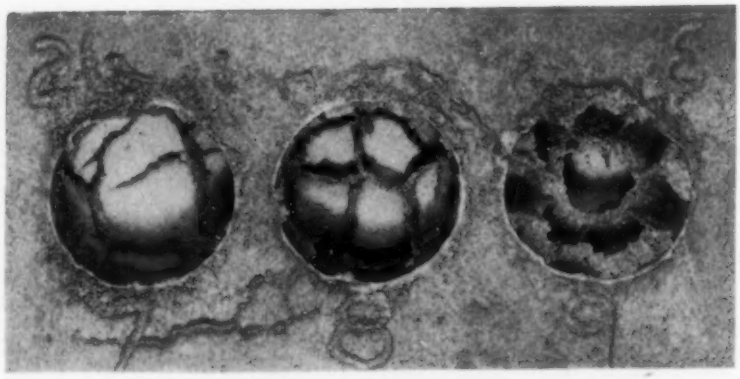


Fig. 21—Photo shows effect of cores containing a resin binder as compared with cores made with regular core sand.

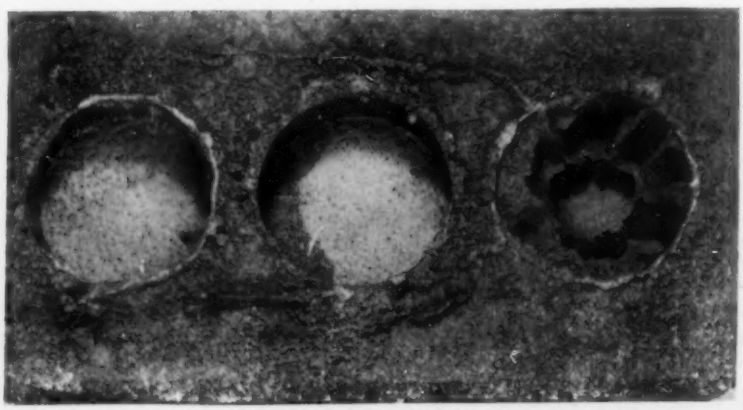


Fig. 22—A comparison of cores of the same composition mixed by different methods.

duced but a small vein was formed. The core at the right shows the normal condition of veining for the sand.

All of the low-fusion-point washes formed glazed coatings on the casting; but, rather then peeling off easily as with steel castings, they adhered tightly to the castings and were difficult to remove.

Clay Content of Molding Sands. Veining occurs more readily in sands bonded with both cereal and clay binders than it does in sands bonded with clay alone. For this reason and because of the fineness, wide grain distribution, and low fusing point of natural sands, the occurence of veins in such sand is rare.

Bentonite Addition

Fig. 19 shows a casting in which regular 180-micron base sand was bonded only with water plus one per cent, 3 per cent and 10 per cent bentonite, respectively, from left to right. The veining is decreased as the bentonite is increased.

Core Sands. The information obtained on core sands used in these experiments shows that a minimum of veining occurs when the more collapsible cores are used. It is believed that veining will not occur in cores which collapse before sweating begins because cracks in such cores are filled with loose sand. Evidence to support this theory is shown in Figs. 20, 21, and 22. In Fig. 20, a photograph of a casting is shown in which all cores were bonded with core oil and corn flour: No. 1 contained 0.5 per cent of each, No. 2 was the regular core sand with 1.0 per cent of each, and No. 3 had 2.5 per cent of each binder. Core No. 1 apparently had collapsed before sweating commenced, while the other two had not. Two cores bonded with a collapsible resin binder are compared with one containing the regular binders in Fig. 21. Core No. 7 contained 1.0 per cent resin; core No. 8, 1.0 per cent resin plus 0.5 per cent dextrine; and core No. 9 was composed of regular core sand. It is well known that quick collapsibility is a characteristic of all cores bonded with resin binders. This property probably is responsible for the appreciable reduction in veining which results from the employment of resinous core binders.

A final illustration of the effect of collapsibility is shown in Fig. 22. All

the cores used in this casting contained the ingredients of regular core sand, but they were mixed differently. The one on the left was mixed in 200-gram batches by hand and the one in the center by means of a mortar and pestle, while the righthand core was made from a batch mixed in the production muller. In both the poorly mulled cores veining was eliminated, probably because they were loosely bonded and collapsed soon after pouring.

Conclusions

Severe cases of veining in bronzes are caused by cracking of the mold due to thermal stresses and subsequent sweating of the metal. The elimination of either of the two causes is sufficient to stop or greatly reduce the occurrence of veining. Sweating is caused by gas evolution from the solidifying metal and may be eliminated by proper melting practice. The best methods found for reducing cracking of the mold are the following:

a. The use of the fine sand with a wide grain-size distribution.

b. The addition of wood flour to the molding sands.

c. Washing of molds and cores with a water solution of a flux such as sodium carbonate.

d. The use of as little organic binder as is practical.

e. The use of as collapsible cores as is possible.

Veins on gray iron castings are believed to form in a manner similar to that of vein formation on bronze castings except that the internal pressure is produced by the growth of graphite flakes instead of gases evolved from the metal.

Veining defects on steel castings are less prevalent than on bronze castings. They occur as a result of hot tears which are caused by stress conditions within the mold. Elimination of veining under these conditions is primarily a problem of preventing the formation of hot tears.

Prevention of penetration can be accomplished simply by applying a silica flour mold wash to those areas in which the penetration usually occurs.

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DISCUSSION

Chairman: B. H. Booth, Carpenter Bros., Inc., Milwaukee, Wis.

Co-Chairman: R. E. Morey, Naval Research Laboratory, Washington, D. C.

H. W. Dietert[1]: I would like to know whether the wood flour contained iron oxide. Also, I was interested in the photographs. What type of camera did you use to make those photographs?

Mr. Morey: The wood flour, as far as I know, did not contain any iron oxide. Two cameras were used. One was a Speedgraphic and the other a view camera. Some specimens were lighted on both sides. These were taken with the view camera. Others photographed with the Speedgraphic were illuminated with just one flash bulb.

Chairman Booth: Which lighting produced the best results?

Mr. Morey: I believe we got as good results with the view camera and the two lights, although that gives you a peculiar effect. It is hard to illuminate the interior of the holes to get the defects to show up well.

T. A. Tarquinio[2]: Was the test casting shown in Fig. 1 used for all tests?

Mr. Morey: Yes.

Mr. Tarquinio: How was the casting made?

Mr. Morey: The 2x2-in. core specimens were inserted in the core print, with paste to hold them in. The subsequent pictures were direct view photographs.

Mr. Tarquinio: Would the center core be subject to different heat stresses than the two on the outside?

Mr. Morey: It might have a minor effect. However, the metal thickness over all the test specimens was at least an inch, so I think they had sufficient heat on all of them and that position would have a minor effect. It is possible that there would be slightly more heat on the center specimen.

O. J. Myers[3]: Was the temperature control the same for all tests?

Mr. Kattus: Yes, the pouring temperature was controlled with an optical pyrometer at 2000° F.

There is one thing I would like to add. Mr. Morey did not mention how we came to the conclusion that gas was responsible for the sweating. We ran density determinations on six castings. The pouring temperatures, the chemical compositions, and the position of specimens for the density specimens were the same for each one; however, three of the castings had veining and sweating and three did not. The castings which sweated and veined, had lower densities than those which did not sweat and vein. Therefore, we concluded that there was gas in the castings that sweated and veined.

Member: In regard to silica wash for steel castings, what specific gravity do you recommend?

Mr. Morey: We very seldom use any specific gravity determinations for mold washes. We mix it to the consistency that seems to be most satisfactory, and as we need more, make it up to the same formula. The wash for steel castings, has five parts of total solid materials to three parts of water.

Member: Were tests made on the expansion characteristics of any of the test samples?

Mr. Kattus: Yes, we performed a lot of expansion tests but could find no correlation between expansion and the veining.

Mr. Dietert: Were any hot strength tests made?

Mr. Kattus: We made many hot strength tests, but could find no correlation, except for the plasticity. The veining did not seem to depend much on hot strength. However, we produced sands that were plastic at the pouring temperature; in other words, sands which compress without breaking. These stopped the veining. We concluded that the reason for this is because the stresses in the sand are released by plastic flow, rather than by cracking of the mold wall.

Member: I understood Mr. Morey to say that as you increase cereal binder content you increase the veining. That is a vegetable binder. So is wood flour.

When wood flour is used, veining decreases. How do you account for the different effects of two types of vegetable binders?

MR. KATTUS: I cannot account for it in any way theoretically, but we found that the cereal corn flour binders made the sand specimens more rigid, while large amounts of wood flour made them plastic. We tried commercial wood flour and went into the pattern shop and picked a handful of sawdust off the floor. Both produced the same effect.

MEMBER: You would be guided by the fluxing action of the ash.

MR. KATTUS: The wood flour does no good if it is used in a sand without clay. There seems to be some fluxing action between the clay and the wood flour. If you use it with core binders, oil or cereal binders, it will not work unless you use a clay with it.

CHAIRMAN BOOTH: How much wood flour did you use?

MR. KATTUS: About five per cent.

J. A. RASSENFOSS[4]: What was the percentage of clay and the green strength for the mixture that had 5 per cent wood flour? Is the percentage by weight or volume?

MR. KATTUS: The mixture contained 5 per cent wood flour and 3 per cent clay by weight. The green strength was around 4 or 5 psi.

MR. RASSENFOSS: What was the moisture content?

MR. KATTUS: With wood flour, 5 per cent. We performed tests with both green and dry sands, and found no difference in the penetration and veining.

MEMBER: Would wood flour make any difference in flowability in your molding sand? You say that the particular sand you chose had low flowability. Would that make any difference in ramming up?

MR. KATTUS: No, it made no difference, but it does reduce the dry strength of the sand. It does not have much influence on flowability.

D. C. WILLIAMS[5]: On gray iron castings we took the chemical analysis of the body of the casting and the chemical analysis of the vein which we found. The body had about 4 per cent carbon and the vein had 1 per cent carbon. The silicon increased from about 2 to 4 per cent. Both increments occurred when we had corn flour in the mixture. We have not eliminated them yet.

I do know that corn flour will form a vein in gray iron. I think that the vein forms about 10 min. after the casting is poured, rather than immediately after pouring.

J. L. HALL[6]: I believe the author said that fins in steel castings were caused by hot tears. Have you any data to support that, or is that just theory?

MR. KATTUS: There is a paper by Bishop, Taylor and Johnson, from the Naval Research Laboratory, that explains that. They actually have data and photographs to show it. There has also been previous work by some German investigators that shows the same thing.

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DUCTILITY AND ELASTICITY OF WHITE AND GRAY IRONS

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American Brake Shoe Co.
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ADDITIONAL TESTS to indicate serviceability of irons of different types has been a long expressed need of the foundry industry. For example, in 1933[1] it was recognized by the ASTM that an impact specification might be effective when tensile, hardness and other existing tests were inconclusive in distinguishing among various analyses, but no completely satisfactory solution was reached.

In this paper a different measurement to indicate relative serviceability of cast irons under severe conditions is discussed: the determination of the plastic and elastic deformation accompanying stress, that is, the complete stress-strain diagram.

Although popularly considered brittle materials, white, mottled and gray irons can be manufactured with definite elasticity and ductility as well as reproducible tensile strength and other properties. As the matrix structure is changed from pearlitic to austenitic, the plastic elongation changes from 0.0000 to 0.0500 per cent in white and mottled irons and from 0.05 to 3.0 in gray iron. Ferritic-pearlitic gray irons exhibit up to 1.0 per cent elongation. Modulus of elasticity varies from 26.0 to 10.4 x 10⁶ psi. depending upon structure.

Preliminary service tests show that in many instances a given amount of *elongation* rather than merely a definite tensile strength is required. Furthermore, the *type* of elongation (whether plastic or elastic) can effect service life pronouncedly.

It is the purpose of this paper to point out these differences in strain behavior among irons and to indicate in a general manner the applications of this characteristic.

Since strain is a fundamental mechanical property it is dependent upon structure just as are tensile strength and hardness[4,5]. The following structures have therefore been surveyed to determine some of the variations:

White Irons

Carbon mostly as: Iron Carbide
Matrix:
a. pearlite
b. martensite + austenite
c. austenite + martensite
d. austenite (?)

Gray Irons

Carbon mostly as: Flake Graphite
Matrix:
a. pearlite
b. pearlite and ferrite
c. martensite (tempered)
d. acicular ferrite
e. austenite

Part II. Definition of Terms

Because of the special application of certain testing terms in the case of cast iron, definitions are given here for reference before reviewing the literature. The complete stress-strain curve, Fig. 1, has been only occasionally determined for cast iron because of the instrumental difficulty in obtaining strain at the breaking load. Recent development of the SR-4 Gage by Ruge, DeForest and others[9] has permitted accurate determination of elongation to the breaking load and in some instances after fracture. Autographic devices of somewhat lower precision are also available.

The following terms used in describing the mechanical characteristics of the various structures are shown graphically in Fig. 1, the stress-strain curve of a soft gray iron.

A. Tensile Strength (T.S.) (pounds per square inch). This may also be termed the ultimate strength. The value is slightly lower than the true tensile strength since the original diameter, rather than that at rupture, is used in the standard calculation.

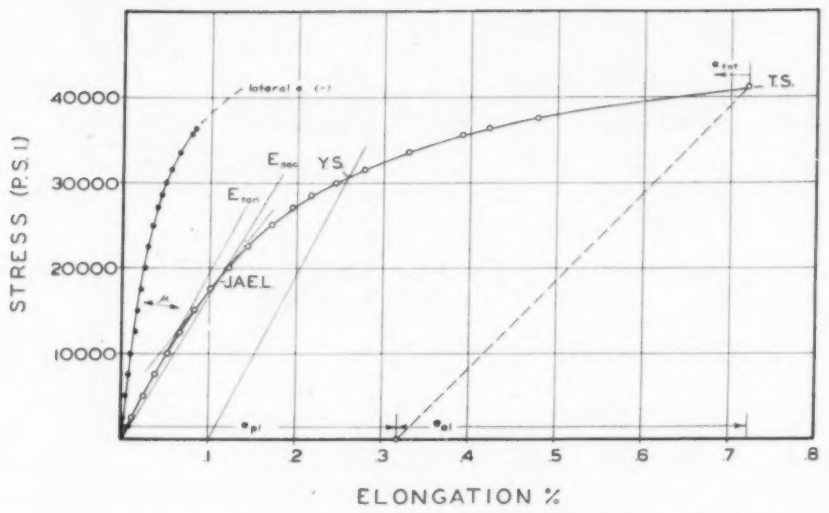
B. Elongation (e_{tot}) (inches per inch or per cent is the change in length (strain) caused by stress. At higher stresses both plastic and elastic elongation occur.

1. *Plastic Elongation* (e_{pl}) (inches per inch or per cent), Ductility. The commonest quantitative measure of ductility is plastic strain, also called permanent set. The terms will therefore be used synonymously in this discussion to describe the deformation that is not removed by unloading the specimen.

2. *Elastic Elongation* (e_{el}) (inches per inch or per cent) is the portion of the total strain which disappears if stress is removed.

C. (Tangent) Modulus of Elasticity (E_{tan}) psi. is the ratio of stress to strain at the (lower) stress values

Presented at a Gray Iron Symposium Session, Fiftieth Annual Meeting, American Foundrymen's Association, at Cleveland, May 8, 1946.



Chemical Analysis; TC, 3.48; CC, 0.84; GC, 2.64; Mu, 0.54; P, 0.27; S, 0.10; Si, 0.77 per cent.
T.S.: Tensile Strength, 41,100 psi.
e_{tot}: Total Elongation, 0.722 per cent.
e_{pl}: Plastic Elongation, 0.318 per cent.
e_{el}: Elastic Elongation, 0.404 per cent.
E_{tan}: Modulus of Elasticity, 19.5×10^6 psi.
Y.S.: Yield Strength, 0.1 per cent offset, 30,600 psi.
J.A.E.L.: Johnson's Apparent Elastic Limit: 18,000 psi.
E_{sec}: Secant Modulus: 16.8×10^6 psi.
μ: Poisson's Ratio: Lateral elastic strain/longitudinal elastic strain = 0.23.

Fig. 1—Tensile stress-strain curve of soft gray iron (AP2, Table 2) showing mechanical constants. White circles show longitudinal strain (tension). Black circles show accompanying compressive strain across specimen.

through which a straight line may be drawn. As with many other materials, e.g., brass, aluminum, the modulus changes with increasing stress. In other words, the curvature of the graph is partly due to a change in the stress to elastic strain ratio and partly to plastic strain. To average the modulus over a portion of the curve, the secant modulus (E_{sec}) is sometimes used. This modulus line may be drawn from the origin to a given percentage of the tensile strength or to "Johnson's Apparent Elastic Limit."

D. Yield Strength (Y.S.) psi. is the stress value intersected by a line parallel to the tangent modulus through 0.1 per cent or 0.2 per cent elongation and zero load. In steel this would indicate the stress with 0.1 per cent permanent deformation, but for irons less than 0.1 per cent set will have taken place at this point because of the changing modulus referred to in *C*. In some white irons with very low plastic elongation yield strength is not determinable.

E. Johnson's Apparent Elastic Limit (J.A.E.L.) psi. is the stress at which the strain rate is 50 per cent greater than at the beginning of the curve. It is determined by drawing a line through the origin with 50 per cent greater slope than the modulus. A line parallel to this but tangent to the stress strain curve locates the desired point. This value is more reproducible than the elastic limit or proportional limit.

F. Poisson's Ratio (μ) *Ratio* inches per inch. As a specimen elongates in the direction of tensile stress, compressive *strain* occurs across the specimen as shown by the reduction of cross section. The ratio of elastic compressive strain across the bar to longitudinal tensile strain is of importance in many calculations involving residual stresses and is included for reference.

G. Relative Effects of Elastic and Plastic Strain. The effects of elastic and plastic deformation upon casting performance may be mentioned here as a corollary to the definitions just discussed.

If (Fig. 1) the specimen is stressed the first time to just below the tensile strength· 41,000 psi. for example, approximately 0.7 per cent total elongation is obtained. Of this 0.3 per cent is plastic and 0.4 per cent elastic elongation. If the load is removed the elastic strain disappears and the plastic strain, or permanent set, remains.

If the specimen is again stressed to 41,000 psi. only 0.4 per cent elongation (elastic) will be obtained during loading in contrast to the 0.7 per cent observed during the first test. This difference is important when the relative uses of structures with high elastic or high plastic elongation, as in Fig. 12, are being compared.

The more elastic materials are capable of withstanding higher *repeated* deflection of a given amount without fracture. On the other hand, it must not be assumed that plastic elongation is of significance only during the first cycle of use. For example when a casting is externally heated the inner layers may be deformed plastically in tension.

During the cooling cycle, then, as the outside tends to restore the original dimensions the interior builds up compressive strain. Upon *reheating* in the next cycle, no tensile strain is encountered until this compressive strain is *removed*. The plastic deformation of the first cycle has therefore permitted the casting to adapt itself to greater elastic strain than indicated by the tensile elastic strain portion of the stress strain curve.

Literature Review

Thum, MacKenzie, Massari, Draffin and others[1, 2, 3, 6, 7, 8] have reported the plastic and elastic deformation of various irons for transverse, tensile and compressive tests. Most of the determinations have been up to 80 per cent of the breaking load. The maximum set in tension reported by MacKenzie from Thum's article was 0.05 per cent while Massari indicated values of 0.9 per cent elongation at 20,000 psi. tension and 0.6 per cent at 50,000 psi. compression for other analyses.

Thum and Massari both indicated a correlation of set with combined carbon. It was further developed by Thum that graphite flakes acted to reduce modulus of elasticity by diminishing the section of matrix under test. Ziegler and Northrup[8] also measured deformation as part of an investigation on superheating and found no correlation with pouring temperature or carbon content. They believed that variable structural effects, not discussed, were operative.

Since the completion of the majority of the above research, new engineering irons with greatly improved tensile strength, and resistant to abrasion, growth, corrosion and

general wear have been developed. The useful characteristics of these new materials were only partially indicated by standard tensile, transverse and hardness tests; an attempt to obtain a greater knowledge of the mechanical properties based upon stress-strain relationships was therefore indicated.

Part III. Experimental Procedure

A. Preparation of Specimens. Test specimens were obtained from a variety of sources, since a large share of the work was performed to correlate mechanical properties with service performance. A number of the specimens were obtained from commercially produced castings (cupola melted) as well as from induction melted experimental test bars. The details of preparation of each sample are given in Tables 1 and 2.

B. Tensile Tests. Tensile specimens, 0.505 in. in diameter, were used in all cases with a 2-in. gage length except for some instances when a 1½-in. gage length with long fillets was employed. Since strain was measured over the central inch of the specimen no difference was encountered. Type A-3, SR-4 resistance strain gages[9] were cemented to both sides of the specimen and measured in series.

Tests were made with a hydraulic testing machine, 60,000-lb. capacity, at a crosshead speed of 0.02-0.05 in. per min. The precision of strain measurement was better than 0.0003 per cent up to 0.2 per cent and better than 0.001 per cent to beyond 1 per cent. Huggenberger extensometers were used simultaneously during some of the tests and agreed with the SR-4 gage readings within the limit of error of the extensometer.

Small amounts of shrinkage were encountered in most specimens from standard ASTM arbitration bars. This shrinkage was not observable in the tensile fracture under low magnification, but was easily seen without magnification in microspecimens cut at right angles to the fracture. The seemingly slight defect causes marked reduction of both tensile strength and strain.

In Fig. 2-A the stress strain curves of tensile specimens cut from a 1.2 in. diameter arbitration bar and from an adequately headed tensile specimen of similar section (Fig. 2-B) show that total elongation is increased 100 per cent (0.42 per cent vs. 0.83 per cent) by elimination of shrinkage. The iron, furthermore, is not considered a high shrinkage material.

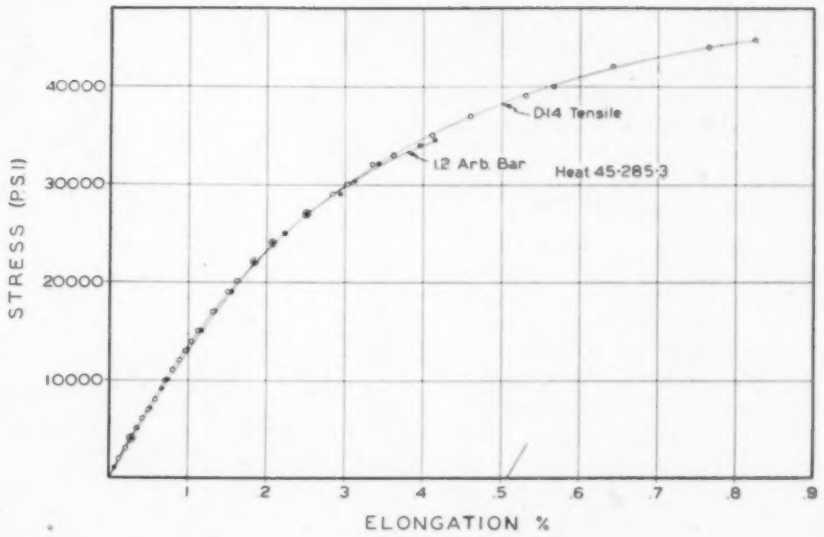


Fig. 2 (A)—Tensile stress-strain curves of 0.505 in. diameter tensile specimens for 1.2 in. diameter arbitration bar (solid circles) and properly risered D-14 specimen (open circles). High carbon nickel-molybdenum iron. Chemical analysis; TC, 3.66; Mn, 0.94; P, 0.10; S, 0.06; Si, 1.82; Ni, 1.52 Mo, 1.06 per cent.

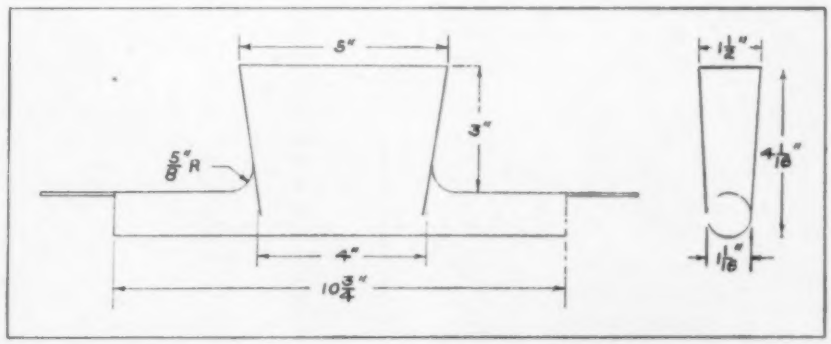


Fig. 2 (B)—Sketch of casting D-14 for obtaining single cast tensile bar. A 1 1/16 in. diameter x 10¾ in. long test bar is provided from which the tensile specimen is machined with the gage length located in the sound metal directly under the head.

Irregularity of results from the high strength nickel - molybdenum iron and the austenitic iron analyses are also caused by this variable. Curves obtained from these bars containing shrinkage are included since they are from standard specimens and indicate the general form of the first part of the curve for perfect samples (Fig. 2-A).

C. Chemical Analyses. All analyses were made from ⅛-1/16 in. cubes rather than upon drillings to prevent loss of graphitic carbon. Combined carbon was determined by the difference between the total and graphitic carbon analyses.

D. Microhardness. These determinations were made with a Tukon microhardness tester under a 25-gram load and a Vickers 136° pyramidal diamond indenter. A sliding stage was used which allowed location under the microscope of the desired area for testing.

E. Concentrated Compression (Mushroom) Tests. Behavior under heavy concentrated compressive loading was investigated by using conical specimens, as illustrated in Fig. 3. The small flat area (0.060-in. diameter) at the end of the reduced section is pushed against a large surface ground martensiti-

white iron block in the same testing machine as used in the tensile tests. The test affords an opportunity to observe marked plastic deformation and the details of structural damage in compression and shear.

F. Creep Tests. Creep tests were conducted as described by Fellows, Cook and Avery[10] and are included to illustrate the high temperature ductility of gray iron.

G. Impact Tests. Unnotched impact specimens, 4 × 0.707 × 0.707 in., were broken transversely as Charpy specimens, in a pendulum type impact testing machine. Effective span of the specimen was 3.25 in. Data are also given for Izod tests upon 1.2 in. arbitration bars, unnotched, struck 3 in. above supports.

Part IV. Discussion of Results

The stress-strain curves and other results will be discussed in the following order progressing from the white irons of less than 0.0005 per cent plastic strain to gray irons of over 3.0000 per cent plastic elongation.

A. White and Mottled Irons
1. Effect of Matrix
2. Effect of small amounts of graphite

B. Gray Irons
1. Effect of Matrix
   a. Pearlite
   b. Pearlite-Ferrite
   c. Tempered Martensite
   d. Acicular Structure
   e. Austenite

C. Miscellaneous Tests of Plastic Deformation
   a. Microhardness
   b. Mushroom (concentrated stress) Tests
   c. Creep Tests

IV-A. White Irons—Table 1, Figs. 4, 5, 6, 7, 8.

1. Effect of Matrix. In progressing from pearlitic white iron Heats 44-434, 44-440, to austenitic-martensitic irons, Heat 44-441· the plastic deformation increases from less than 0.0005 per cent to 0.0083 per cent despite the increase in hardness from 450 to 650 BHN.

Since the precision of measurement is closer than 0.0003 per cent, this represents a significant change. Modulus of elasticity exhibits a slight decrease from 25.5-26.5 × 10^6 to 24.2-24.8 × 10^6. Charpy impact strength is improved.

Table 1

STRESS STRAIN RELATIONSHIPS IN VARIOUS WHITE AND MOTTLED IRONS

Reference No.	General Description	Structure: Carbon	Structure: Matrix	Chemical Analysis, per cent: TC	CC	GC	Mn	P	S	Si	Ni	Cr	Casting	Tensile Strain, psi.	Total Strain, per cent	Elastic Strain, per cent	Plastic Strain, per cent	Tangent Modulus of Elasticity, $P_{in.}$ x10^6	BHN., As Cast: Surface	BHN., As Cast: Center	Impact, ft.lb. Charpy .707#
44-434	1.5% Cr White Iron	Carbide	Medium Pearlite	3.40	3.37	0.03	0.62	0.05	0.07	0.68	—	1.55	[1]1-in. Y	46,000	0.1780	0.1780	0.0000	26.0	495	460	6.5
													[2]3-in. Y	34,350	0.1300	0.1300	0.0000	26.5	468	444	6.8
													[1]1-in. Y	40,250	0.1580	0.1580	0.0000	25.5	—	—	5.5
				Tempered 500° F.—8 hr.																	
44-440	1.5% Ni 1.5% Cr White Iron	Carbide	Fine Pearlite	3.43	3.39	0.04	0.72	0.05	0.05	0.62	1.58	1.54	[1]1-in. Y	35,750	0.1380	0.1380	0.0000	26.0	555	555	5.5
													[2]3-in. Y	36,750	0.1450	0.1445	0.0005	25.5	555	512	4.8
				Tempered 500° F.—8 hr.																	
44-441	4.0% Ni 2% Cr White Iron	Carbide	85% Aust. 15% Mart.	3.41	3.36	0.05	0.74	0.05	0.05	0.66	4.02	2.02	[1]1-in. Y	31,500	0.1330	0.1250	0.0080	24.2	652	600	7.5
				Tempered 500° F.—8 hr.									[1]1-in. Y	51,250	0.2050	0.2002	0.0048	25.8	—	—	7.3
		Carbide	70% Aust. 30% Mart.	3.41	3.36	0.05	0.74	0.05	0.05	0.66	4.02	2.02	[2]3-in. Y	32,650	0.1430	0.1354	0.0076	24.7	652	600	6.8
				Tempered 500° F.—8 hr.									[2]3-in. Y	50,000	0.2020	0.1976	0.0044	25.3	—	—	7.5
44-435	4.5% Ni 1.5% Cr Mottled Iron	Carbide and Graphite	15% Aust. 85% Mart.	3.39	3.06	0.33	0.66	0.05	0.05	0.98	4.43	1.54	[1]1-in. Y	30,000	0.1310	0.1268	0.0042	24.2	600	532	9.0
				Tempered 500° F.—8 hr.									[1]1-in. Y	51,500	0.2150	0.2126	0.0024	24.8	—	—	10.5
		Carbide and Graphite	5% Aust. 95% Mart.	3.39	1.82	1.57	0.66	0.05	0.05	0.98	4.43	1.54	[2]3-in. Y	43,250	0.3629	0.3124	0.0505	18.7	555	504	11.0
				Tempered 500° F.—8 hr.									[2]3-in. Y	50,000	0.3134	0.2921	0.0213	20.5	—	—	10.3

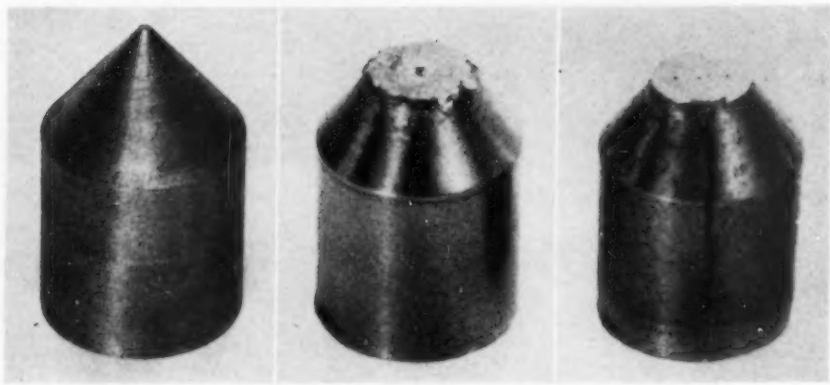
[1] 1-in. Y Block: 1 in. wide, 2¾ in. high, 6 in. long heavily headed along top. Total Weight, 12 lb. High frequency induction furnace melted.
[2] 3-in. Y Block: 3 in. wide, 2¾ in. high, 6 in. long heavily headed along top. Total Weight, 50 lb. High frequency induction furnace melted.
[3] 0.707 in. sq. x 4 in. long. Charpy unnotched. 3.25 in. between supports, struck centrally, 120-ft. lb. blow.

Table 2

STRESS STRAIN RELATIONSHIPS IN VARIOUS GRAY IRONS

Reference No.	General Description	Structure: Carbon	Structure: Matrix	Chemical Analysis, per cent: TC	CC	Mn	P	S	Si	Ni	Cr	Other	Casting	Tensile Strength, psi.	Total Strain, per cent	Elastic Strain, per cent	Plastic Strain, per cent	Tangent Modulus of Elasticity, psi. x10⁶	Yield Strength, 0.1 per cent offset, psi.	Poisson Ratio	BHN.
44-079-17	3%Si, 2%Cr Iron	Carbide + Graphite	Pearlite	3.10	1.20	0.98	0.10	0.10	3.57	—	1.94		D14[1]	42,500	0.360	0.293	0.067	17.3	42,000		293
44-448-17				3.54	1.15	0.99	0.13	0.11	3.38	—	1.95			28,600	0.423	0.353	0.070	10.7	26,500		235
AP1	Gray Iron, A.C.		Fine	3.48	0.84	0.54	0.27	0.10	0.77	—	—			42,800	—	—	—	20.2	31,000	0.24	194
AP2	from 1600° F.		Pearlite											41,100	0.722	0.404	0.318	19.5	30,600	0.23	194
BP2	Mold cooled		Med. Pearlite	3.46	0.82	0.49	0.27	0.10	0.60	—	—			40,300	0.916	0.429	0.487	19.3		0.23	187
EP1	Mold cooled,		Medium Coarse	3.48	0.81	0.58	0.29	0.11	0.87	—	—			37,700	—	—	—	19.2	24,000		165
EP2	Higher Si		Pearlite										Wheel[2]	42,800	—	—	—	17.1	24,500	0.24	165
X1	Cooled 10° F./hr.		Coarse Pearlite + Ferrite	3.49	0.60	0.47	0.29	0.11	0.60	—	—			31,000	—	—	—	17.5	19,000	0.22	130
X2	from 1600° F.													29,700	—	—	—	17.3	18,000	0.23	130
CP1	Same, higher Si			3.48	0.44	0.55	0.30	0.13	0.72	—	—			28,300	—	—	—	15.0	18,000	0.22	124
45-005B				3.51	0.40	0.61	0.07	0.08	1.05	—	—			17,750[3]	1.359	0.279	1.180	12.0	13,000		103
FP1	Heat treated[5]		Tempered Martensite and Acicular	3.51	1.03	0.55	0.28	0.10	0.72	—	—			64,000	0.530	0.416	0.104	16.9	58,000	0.22	321
FP2	Gray Iron													59,000	0.462	0.392	0.070	17.1	55,800	0.24	321
M405	Acicular Ni-Mo[6]	Graphite	Acicular	2.34	0.65	0.88	0.02	0.02	2.64	2.93	0.25	Mo 0.97	1.2 in.[4]	78,250	0.643	0.455	0.188	17.7	65,000		375
406	Iron, Low C													71,500	0.559	0.434	0.125	17.7	65,000		418
403	Same tempered		Acicular Tempered											84,250	0.584	0.491	0.093	17.6	84,000		402
404	500° F.-15 hr.													93,125	0.721	0.570	0.151	17.6	83,000		430
45-285-3	High C Ni-Mo Iron		Acicular	3.66	—	0.94	0.10	0.06	1.82	1.52	0.03	Mo 1.06	D14[1]	44,700	0.824	0.510	0.314	14.0	32,000		235
407	14%Ni, 6%Cu, 2%Cr[6]	(Some Carbide)	Austenite	2.93	0.97	1.52	0.17	0.02	2.03	14.28	1.82	Cu 6.04	1.2 in.[4]	21,400	1.980	0.350	1.630	12.2	14,300		101
408	Austenitic Iron													22,700	2.332	0.362	1.970	12.2	14,300		112
409	20%Ni, 2%Cr			2.91	0.79	1.51	0.11	0.03	1.68	20.45	2.11			23,250	2.326	0.296	2.030	12.9	15,200		116
410	Austenitic Iron													25,200	3.096	0.336	2.760	12.9	15,200		114
411	18%Ni, 3%Cu, 2%Cr			2.89	0.85	1.21	0.09	0.03	1.81	18.38	2.28	Cu 2.74		31,750	1.692	0.341	1.351	14.9	20,500		140
412	50% Steel Charge													33,900	2.215	0.361	1.854	14.9	20,500		143
413	30%Ni, 3%Cr			2.49	0.48	1.08	0.21	0.13	1.61	30.34	3.04			25,200	1.202	0.270	0.932	12.3	18,300		118
414	Austenitic Iron													27,250	1.615	0.295	1.320	12.3	18,300		137

[1] D14 Bar, 1⅛ in. dia. by 10¾ in. long, poured horizontally through heavy head over central 4 in. of bar. High frequency induction furnace melt—as cast.
[2] Specimens cut from plate section (1½ in. thick) of experimental 750-lb. wheel. Cupola melted.
[3] Same as (2) but induction furnace melt.
[4] 1.2 in. arbitration bar cast vertically in dry sand. Specimens contained microshrinkage.
[5] Water quenched 1500° F. to 700° F., air cooled.
[6] Izod impact over 120 ft./lb. (1.2 in. arbitration bar struck 3 in. above supports).



1209 988 988

Left to right: Specimen before testing; gray iron test specimen after loading to 56,000 lb.; spheroidized mottled iron specimen after loading to 42,000 lb.

Dimensions of original specimen: Angle of truncated cone, 60°; diameter of flat at small end, 0.060-0.62 in.; diameter of base of cone and cylindrical section, 1.000 in.; length of cylindrical section 1.00 in.

Fig. 3—Concentrated compression (mushroom) test specimens.

Tempering of the austenitic-martensitic structures at 500° F. for 8 hr. increases tensile strength from 35,000 to 50,000 psi., and decreases ductility 30-50 per cent. Pearlitic irons are not significantly affected since no change in structure occurs.

Impact strength of the austenitic-martensitic structure shows only slight improvement with tempering —an indication that the decreased plastic deformability diminishes the effect of the increased elastic elongation and tensile strength.

2. Effect of Graphite (Mottled Iron)—Table I, Heat 44-434. Increase of silicon and nickel and decrease in chromium causes the expected partial graphitization of the carbide structures, Fig. 8 (in the one-in. section, 0.33 per cent graphitic carbon, and in the 3-in. section, 1.57 per cent average graphitic carbon are obtained (Heat 44-434, Table 1)).

The effect of graphite in the small section in increasing ductility is more than counterbalanced by a decrease in austenite from 85 per cent to 15 per cent (estimated plastic elongation is lowered from 0.0080 to 0.0042 per cent).

However, the greater graphitic carbon in the heavy section causes 0.0505 per cent plastic elongation despite the largely martensitic matrix. A decrease in modulus from 24.2-25.8 $\times$ 10^6 to 18.7-24.2 $\times$ 10^6 psi, is also apparent and is attributed to the presence of graphite.

Tempering of these graphitic structures at 500° F. shows the same effects as in the graphite-free specimens: increased tensile strength but diminished ductility (by 50 per cent).

IV-B. Gray Irons—Effect of Matrix

1. Pearlite—Table 2, Figs. 9, 10. The substitution of graphite for the massive carbide of pearlitic irons previously discussed (Fig. 9) increases the plastic deformation (from 0.0000-0.0005 per cent to 0.0500-0.5000 per cent), decreases the tangent modulus (from 25.5-26.5 $\times$ 10^6 psi. to 12.0-19.3 $\times$ 10^6 psi.), and may increase tensile strength under proper conditions of graphite distribution. Hardness is, of course, reduced. Irons with hypereutectoid combined carbon content, Fig. 9, are less ductile (0.1 per cent set) than eutectoid (0.5 per cent set) compositions.

2. Pearlite-Ferrite—Table 2, Fig.

Fig. 4—Photomicrograph of pearlitic white iron (L 91303).

Chemical analysis: TC, 3.40; CC, 3.37; GC, 0.03; Mn, 0.62; P, 0.05; S, 0.07; Si, 0.68; Ni,—; Cr, 1.55 per cent.

Microhardness readings (left to right): Vickers hardness 248,355 (pearlite); (855,730) 1080, 1080, 990 (carbide); 295, 268 (pearlite) (855) carbide; 320, MnS and carbide.

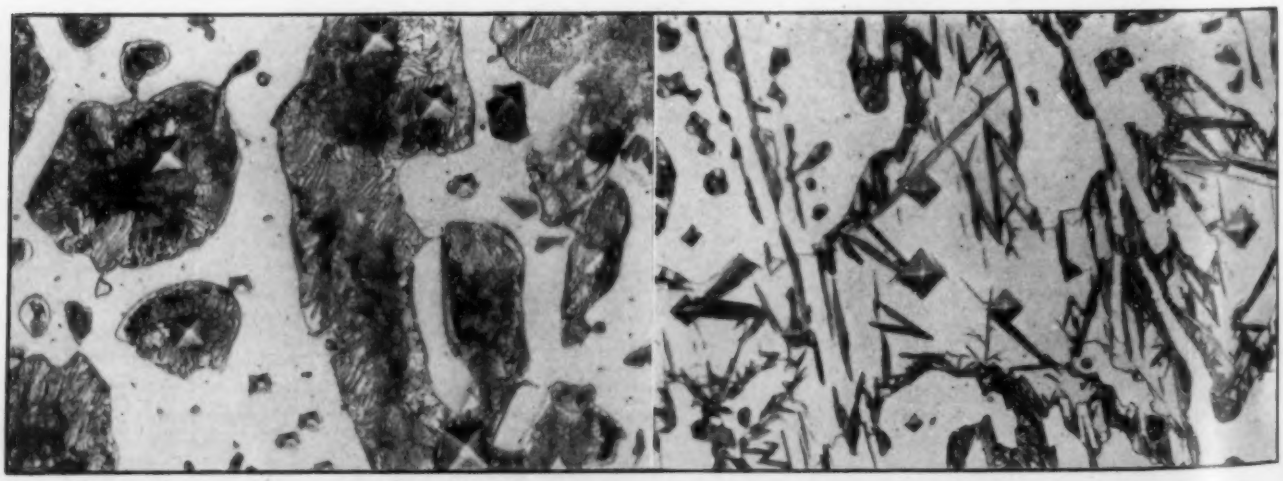
Section of 1-in. Y block. Heat No. 44-434. Etchant; 3 per cent Nital, 8 sec. 500X. BHN. —495.

Fig. 5—Photomicrograph of martensitic white iron (martensite, austenite, carbide) L 91603.

Chemical analysis: TC, 3.41; CC, 3.36; GC, 0.05; Mn, 0.74; P, 0.05; S, 0.05; Si, 0.66; Ni, 4.02; Cr, 2.02 per cent.

Microhardness (Vickers hardness Nos.): austenite-martensite, 595-390; Carbide, 1320-1130.

Section of 1-in. Y block. Heat No. 44-441. Etchant; 3 per cent Nital, 8 sec. 500X. BHN. —652.



10. As combined carbon is decreased (from 0.34 to 0.40 per cent) and corresponding amounts of ferrite appear, tensile strength declines (from 42,800 to 17,750 psi.) and plastic deformation increases (from 0.318 to 1.180 per cent).

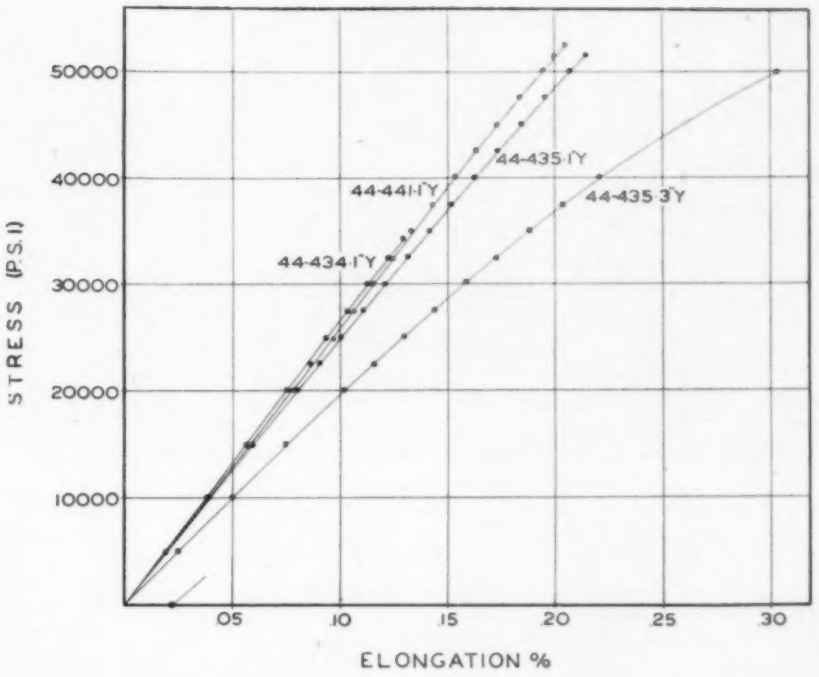
Total strain is greater than for higher strength irons, but the elastic component of the strain is less (1.359 vs. 0.722 per cent total, 0.404 vs. 0.279 per cent elastic strain, respectively). The coarser pearlite in the lower combined carbon irons may also have contributed to the above effect.

3. Tempered Martensitic—Acicular Structures—Table 2, Fig. 12. In contrast to the ferritic pearlitic structures, 45-005B, tempered martensite, FP-2, shows high strength (60,000 psi. in the case selected) but less than 10 per cent of the ductility. However, if plastic deformation is to be avoided, higher stresses may be sustained with the latter iron without encountering appreciable set. For example, yield strength at 0.1 per cent offset is 45,000 psi. for the martensitic material vs. 13,000-19,000 psi. for pearlitic (Table 2).

4. Acicular Structure—Figs. 11, 12, Table 2. Greater strength and ductility were obtained in the low carbon acicular structures, Fig. 11, than in the high carbon tempered martensites, despite greater hardness (93,125 psi. maximum tensile and 0.151 per cent permanent set vs. 64,000 psi. and 0.104 per cent permanent set). The effect of tempering upon the plastic properties of the acicular structure is not evident, probably because of the effect of shrinkage in the test bar discussed in Part II.

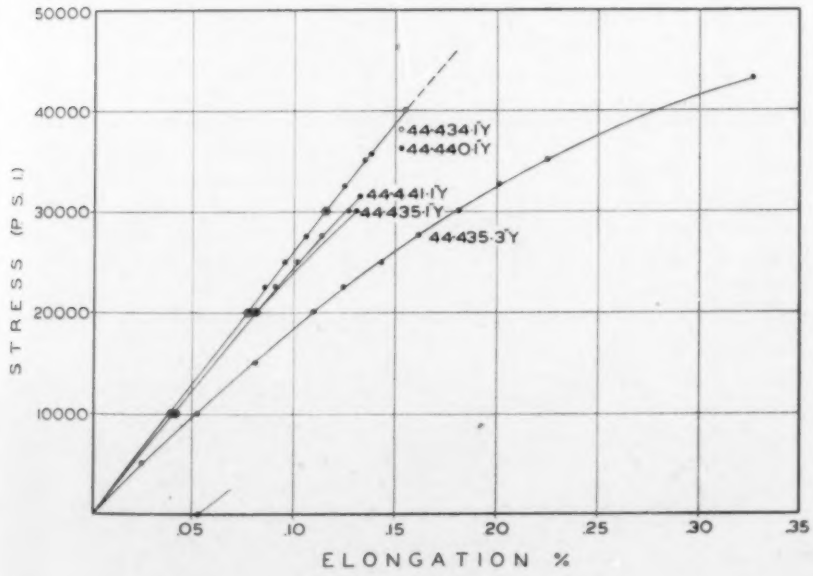
The high carbon, acicular, nickel-molybdenum iron (Fig. 12) indicates that strength and modulus are materially decreased by the increased carbon content.

5. Austenite. The austenitic gray irons (M-407-414, inclusive) containing 14-30 per cent nickel with 2-3 per cent Cr are the most ductile tested, possessing permanent elongation of 1-3 per cent. This is parallel to the characteristics of austenite in white irons previously mentioned and in austenitic steels. The variation between specimens is attributable to pronounced microshrinkage in the test bars.



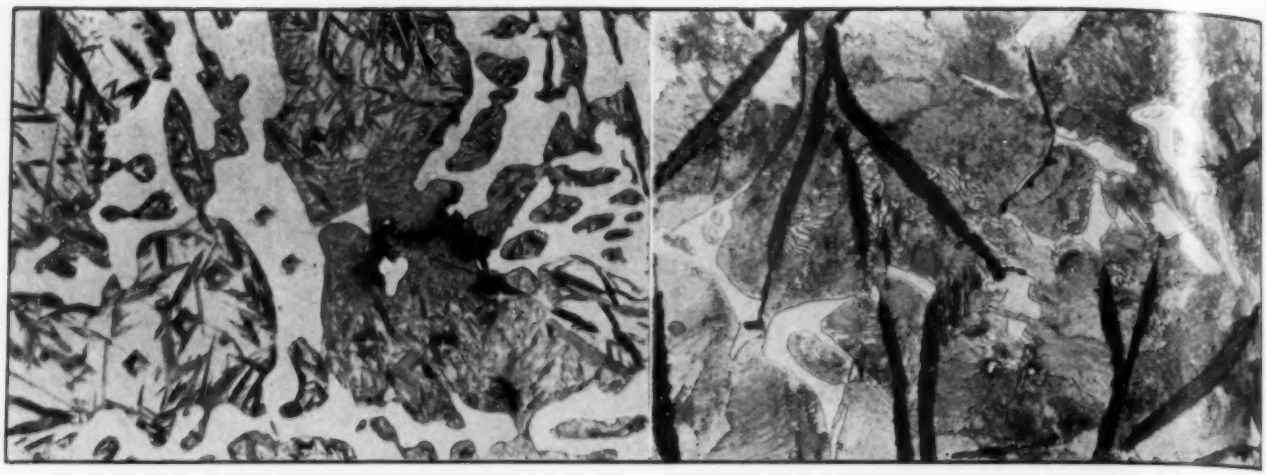
Reference No.	Chemical Analysis, per cent TC	CC	GC	Mn	Si	Ni	Cr
44-434 1-in. Y	3.40	3.37	0.03	0.62	0.68	—	1.55
44-441 1-in. Y	3.41	3.36	0.05	0.74	0.66	4.02	2.02
44-435 1-in. Y	3.39	3.06	0.33	0.66	0.98	4.43	1.54
44-435 3-in. Y	3.39	1.82	1.57	0.66	0.98	4.43	1.54

Fig. 7—Tensile stress-strain curves for white and mottled irons (tempered, 500° F., 8 hr.) of different matrices (0.505 in. diameter specimen).



Reference No.	Chemical Analysis, per cent TC	CC	GC	Mn	Si	Ni	Cr	Matrix
44-434 1-in. Y	3.40	3.37	0.03	0.62	0.68	—	1.55	Pearlite
44-440 1-in. Y	3.43	3.39	0.04	0.72	0.62	1.58	1.54	Pearlite
44-441 1-in. Y	3.41	3.36	0.05	0.74	0.66	4.02	2.02	Austenite
44-435 1-in. Y	3.39	3.06	0.33	0.66	0.98	4.43	1.54	Austenite
44-435 3-in. Y	3.39	1.82	1.54	0.66	0.98	4.43	1.54	Martensite

Fig. 6—Tensile stress-strain curves for white and mottled irons (as cast) of different matrices (0.505 in. diameter specimens).



Chemical analysis: TC, 3.39; CC, 3.06; GC, 0.33; Mn, 0.66; P, 0.05, S, 0.05; Si, 0.98; Ni, 4.43; Cr, 1.54 per cent.

Microhardness (Vickers hardness Nos.) readings (left to right): 572, 520, 555 (martensite-austenite); 1600, 1500 (carbide); 635, 640, 555 (martensite).

Section of 1-in. Y block. Heat No. 44-35. Etchant; 3 per cent Nital, 8 sec. 500X. BHN.—600.

Fig. 8—Photomicrograph of martensitic white iron (slight mottle). Martensite, austenite, carbide, graphite (L 91403).

Chemical analysis: TC, 3.54; CC, 1.15; GC, 2.39; Mn, 0.99; P, 0.13; S, 0.11; Si, 3.38; Ni,—; Cr, 1.95 per cent.

Section of D-14 (1-1/16 in. dia.) tensile specimen. Heat 44-448. Etchant.; 3 per cent nital, 8 sec. 500X.

Greater elongation and lower modulus than pearlite-carbide structure (Fig. 4).

Fig. 9—Photomicrograph of pearlitic iron with free carbides (M 64101).

Tensile strengths are 20,000-34,000 psi. and generally lower than previously reported values obtained from larger test bars in which the microshrinkage at the center of the bar has less effect. The inoculated iron with a steel base charge (M411-412) possesses characteristically higher tensile strength and intermediate ductility.

IV-C. Miscellaneous Tests of Plastic Deformation

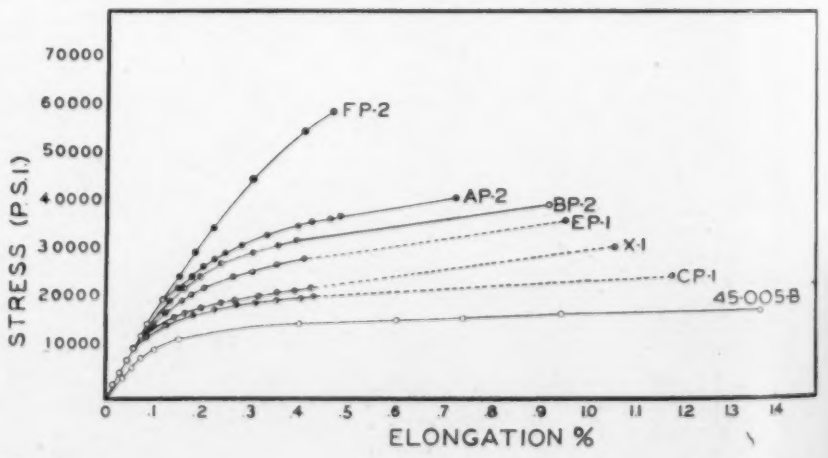
1. Microhardness. A duplex structure rarely behaves in a manner which averages the properties of the individual constituents. For example, the pearlitic white iron (Heat 44-434, Table 1) showed less than 0.0005 per cent plastic elongation, yet over 50 per cent of the structure is composed of pearlite, which if tested alone would exhibit over 10 per cent elongation.

Therefore, to supplement the standard hardness readings in which the indenter presses simultaneously upon large amounts of all constituents, the separate behavior of each phase or aggregate (such as pearlite) has been studied by use of the microhardness technique. Photomicrographs of the impressions, Figs. 4, 5, 8, 14, 15, 16, have been included to show the effects of the impression upon the metal structure, as well as the relative size of the indentations.

The data of Table 3 may be discussed as follows:

a. Carbide Hardness
b. Austenite : Austenite Martensite
c. Other Matrices : Pearlite Acicular Ferrite

a. Carbide Hardness—Figs. 4, 5, 8, 14, 15, 16. Contrary to the general impression that alloyed carbides are pronouncedly harder than iron carbide, no consistent differences in penetration hardness are evident as the total alloy content is changed from zero to 26 per cent Cr or to 4.5 per cent Ni, 4 per cent Mn, 2 per cent Cr. Hardness as measured by this method remains in the range 990-1600 VHN. (VHN. or Vickers Hardness Number is equal to BHN.,



Reference No.	Chemical Analysis, per cent TC	CC	GC	Mn	P	S	Si	BHN.	Heat Treatment
FP2	3.51	1.03	2.48	0.55	0.28	0.10	0.72	321	Quenched and tempered
AP2	3.48	0.84	2.64	0.54	0.27	0.10	0.77	194	Air cool
BP2	3.46	0.82	2.64	0.49	0.27	0.10	0.60	187	Mold cool
EP1	3.48	0.81	2.67	0.58	0.29	0.11	0.87	165	Mold cool, higher Si
X1	3.49	0.60	2.89	0.47	0.29	0.11	0.60	130	Cooled 10° F./hr.
CP1	3.48	0.44	3.04	0.55	0.30	0.13	0.72	124	Cooled 10° F./hr.
45-005B	3.51	0.40	3.11	0.61	0.07	0.08	1.05	103	Cooled 10° F./hr.

Fig. 10—Effect combined carbon upon tensile stress-strain curves of gray iron.

Brinell Hardness Number up to 320; above this value VHN. increases more rapidly than BHN.; example, 500 VHN. =461 BHN., 1021 VHN. =712 BHN.).

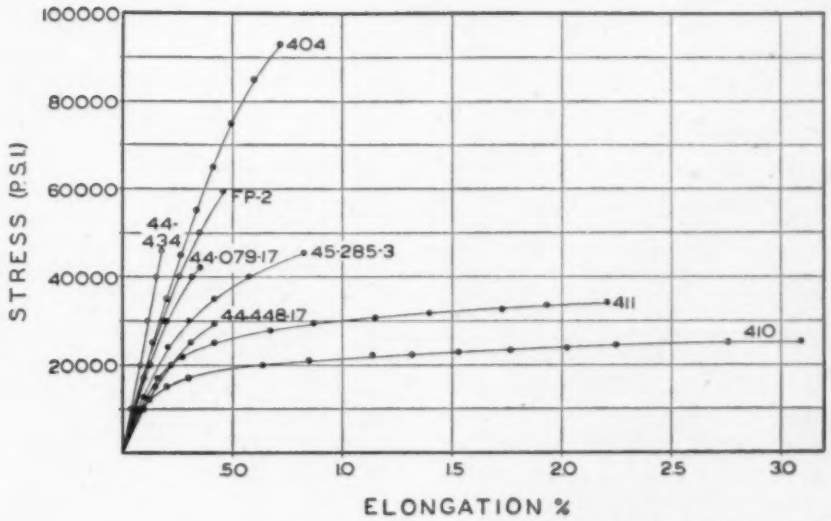
It is of interest that the impressions upon the carbides of the austenitic gray irons (Fig. 14) exhibit nearly the same hardness as those of the white irons (Figs. 4, 5, 8) despite the small areas available for measurement and the soft matrix.

The broad range of carbide hardness may be attributed to the brittleness of this constituent, together with the lowered precision of measurement with small impressions (± 100 VHN. at 1500 VHN.), the directional properties of the carbide crystals and the unknown depth of the cementite at the location of the impression permitting possible penetration into the underlying softer structure.

b. Austenite, Austenite-Martensite—Figs. 5, 8, 14, 15. The hardness of austenite in the gray irons M407, 409, 411, 413, Table 3, varies from 163 to 243 VHN. (BHN.). Matrix microhardness in gray iron is higher than macrohardness because the graphite flakes are eliminated.

In white iron the converse is true, because carbides increase the macrohardness. The change in the matrix surrounding the impression (Figs. 14, 15, either strain hardening or transformation) indicates that even under light loads the material around the penetrator is affected.

The austenite of the highly alloyed nickel-manganese white iron, XE15D (Table 3), exhibits hardness partly characteristic of the austenitic gray irons, but the effect of the traces of martensite is quite evident in the range of readings obtained (160-348 VHN.). As increased amounts of martensite are encountered, L914, L916, the characteristic hardness of austenite is no longer evident and readings from 390 to 695 VHN. are obtained.



Heat No.	Material	Chemical analysis, per cent TC	Si	Ni	Cr	Mo	Cu
44-434	1.5 per cent Cr white cast iron	3.40	0.68	—	1.55	—	—
404	High strength Ni-Mo gray iron	2.34	2.64	2.93	0.25	0.97	—
FP2	Quenched and tempered gray iron	3.51	0.72	—	—	—	—
44-079-17	3.10 per cent carbon Si-Cr heat resistant irons	3.10	3.57	—	1.94	—	—
45-285-3	High carbon Ni-Mo gray iron	3.66	1.82	1.52	—	1.06	—
44-448-17	3.54 per cent carbon Si-Mo heat resistant iron	3.54	3.38	—	1.95	—	—
411	{Austenitic} 18 Ni, 3 Cu, 2 Cr	2.89	1.81	18.38	2.28	—	2.74
410	{Gray Iron} 20 Ni, 2 Cr	2.91	1.68	20.45	2.11	—	—

Fig. 12—Tensile stress-strain curves for gray irons of different matrices.

Hardness Variance

This effect is caused either by underlying crystals of martensite or greater instability of the austenite which causes pronounced transformation as the penetrator is applied. However, the hardness range is still related to the austenite percentage as estimated microscopically. L916: 85 per cent austenite, 15 per cent martensite, 390-595 VHN.; L914: 15 per cent austenite, 85 per cent martensite, 500-695 VHN.

c. Other Matrices—Pearlite, Acicular, Ferritic. Pearlite hardness, as in the case of steels, is related to fineness. In L913, Fig. 4, the range of 212 to 380 VHN. is quite apparently caused by spacing and orientation of the cementite plates. Nickel content produces more uniform pearlite with increased fineness and hardness, 399 to 440 VHN. This is because of the well-known effect upon austenite transformation[4].

Behavior of ferrite alone contrasts sharply with ferrite interspersed with carbide spheroids, Fig. 16, Table 3. The scattered carbides raise the hardness from 140 to 208 VHN.

Impressions upon acicular structures, Fig. 11, confirm previous conclusions that the light etching areas contain a harder austenite-martensite mixture (455-635 VHN.), and the dark etching needlelike areas have a coarser ferrite-carbide aggregate (378-555 VHN.).

Microhardness readings upon a



Fig. 11—Photomicrograph showing low carbon acicular structure (M 48601).

Chemical analysis: TC, 2.34; CC, 0.65; GC, 1.69; Mn, 0.88; P, 0.02; S, 0.02; Si, 2.64; Ni, 2.93; Cr, 0.25; Mo, 0.97 per cent.

Section of 1.2 in dia. arbitration bar from M 403. 500X. BHN. 402.

Acicular ferrite, martensite, retained austenite, graphite (tempered at 550° F.—15 hr.). Highest strength and elastic elongation obtained in this structure.

Table 3

MICROHARDNESS OF VARIOUS WHITE AND GRAY IRON STRUCTURES

Reference No.	Micro No.	General Description	Matrix Hardness — Type	Matrix Hardness — Range Vickers	Matrix Hardness — Av. Vickers	Carbide Hardness — Range	Carbide Hardness — Av.	Micro Hardness — BHN.	Micro Hardness — VHN.[1]
L475	L475	White Iron "0" Alloy	Ferrite alone	150-137	140	1270-1200	1220	—	—
			Spheroidized structure	217-200	208	—	—	—	—
44-434	L913	White Iron, 1.5%Cr	Pearlite	380-212	295	1400-990	1130	460	502
44-440	L915	White Iron, 1.5%Ni, 1.5%Cr	Pearlite	440-339	380	1400-1130	1270	555	649
44-441	L916	White Iron, 4.0%Ni, 2.0%Cr	15% martensite, 85% austenite	595-390	440	1320-1130	1250	600	746
44-435	L914	White Iron, 4.5%Ni, 1.5%Cr	85% martensite, 15% austenite	695-500	595	1600-1500	1500	532	608[2]
XE15D	E231	White Iron, 4.5%Ni, 4.2%Mn, 2%Cr	Austenite, trace martensite	348-160	256	1080-990	1030	482	540
L888	L888	2.5%TC, 26%Cr	Austenite ?	415-348	365	1600-1270	1400	555	649
FPI	L473	Gray Iron, quenched and tempered	Acicular, martensitic	400-355	378	—	—	—	—
		As above, cold worked	Acicular, martensitic	574-536	555	—	—	—	—
M403	M486	Gray Iron, Ni, Mo	Dark etching acicular	555-378	455	—	—	—	—
			Light etching acicular	635-455	555	—	—	—	—
M407	M488	Gray Iron, 14%Ni, 6%Cu, 2%Cr	Austenite	208-163	180	940	940	107	—
M409	M489	Gray Iron, 20%Ni, 2% Cr	Austenite	183-170	180	1030	1030	115	—
M411	M490	Gray Iron, 18%Ni, 3%Cu, 2%Cr	Austenite	243-212	227	—	—	141	—
M413	M491	Gray Iron, 30%Ni, 3%Cr	Austenite	234-217	227	—	—	128	—

[1] Converted from 3000-kg. Brinell impressions.

[2] Presence of graphite (Table 2) lower macrohardness.

quenched and tempered acicular martensitic structure, FP-1, before and after concentrated compression testing show an increase from 355-400 VHN. to 536-574 VHN. This is a further indication that appreciable work hardening occurs in cast iron.

2. *Concentrated Compression (Mushroom) Tests*—Fig. 3. In many applications the structure is subjected to concentrated compressive loading as in line or point contact (e.g., car wheel bearing upon a rail). Because of the difficulty of predicting the effect of these conditions from standard tests, a "mushroom" test has been devised and is being studied (Fig. 3).

Figure 13 and Table 4 indicate that plastic flow under these circumstances is far greater than the amount measured in the conventional tensile test.

Some flaking of both specimens takes place as maximum load is approached. Figure 13 illustrates that the graphite flakes are turned perpendicular to the direction of loading as the matrix flows. Cracking of the carbide-phosphide (steadite) is apparent, but the cracks do not extend into the surrounding metal.

From the final areas of the deformed top surfaces and the maximum load, surface pressures of the order of 200,000-300,000 psi. are obtained. At the edges of the top surface, severe stress concentration is evident and higher pressures are probable.

3. Elevated Temperature Tests. Elevated temperature tests were conducted upon the 3.10 per cent carbon silicon-chromium iron previously described (Table 2). Plastic deformation increased from 0.06 per cent (room temperature test) to 15 per cent when tested at 1400° F. The enhanced ductility is accompanied by marked decrease in strength, but illustrates the marked change in properties at elevated temperatures.

Part V. Review of Data and Application to Practice

Data of sections IV-A,-B,-C indicate that irons of a wide range of elastic and plastic properties are available. Referring again to Figs. 6, 9, 10, these alloys vary from the stiff pearlitic white irons which may be stressed to 35-45,000 psi. with 0.13-0.18 per cent elastic elongation and less than 0.0001 per cent plastic strain, to the comparatively ductile austenitic gray irons with 0.270-0.362 per cent elastic and over 3.0000 per cent plastic elongation at 21-31,000 psi. The high strength acicular gray irons withstand higher stress and more *elastic* deformation (0.434-0.570 per cent) than any of the other materials.

The stress-strain curves are therefore functions of the structure and provide a better evaluation than tensile strength alone. The compressive

Table 4

MUSHROOM TEST OF GRAY AND MOTTLED IRONS

	L473 Gray Iron Martensitic Acicular	L475 Mottled Iron Spheroidized-Ferritic
Longitudinal Plastic Deformation under load, in.	0.204	0.142
Maximum Load, lb.	56,000 (no rupture)	42,000 (rupture)

Table 5

ELEVATED TEMPERATURE TESTS OF SI-CR IRON

Chemical Analysis, per cent — TC	CC	Mn	P	S	Si	Cr
3.10	1.20	0.98	0.10	0.10	3.75	1.94

Temp., °F.	Stress Psi.	Duration of Test, hr.	Plastic Elongation During Test per cent	Remarks
70	42,500	Tensile Test	0.067	
1000	6,000	1,000	0.7	Not carried to fracture
1200	2,000	1,000	1.3	Not carried to fracture
1200	6,000	35	6.0	Fracture
1400	1,500	93.0	15.0	Fracture
1400	2,000	81.7	12.0	Fracture



Severely worked martensitic-acicular structure containing elongated graphite flakes. (Compressive stress in vertical direction.)

Vickers hardness: Before compression—355-400; after compression—536-574 (microhardness readings).

Section of mushroom specimen from wheel FP2. Etchant—3 per cent Nital, 8 sec. 500X.

Fig. 13—Martensitic-acicular iron after compression test (L 47303).

and elevated temperature tests show that greater ductility may be obtained than indicated by room temperature tensile tests. Since the specimens melted in widely different furnaces form an orderly pattern unrelated to melting practice, the foregoing conclusions may be taken as generally applicable.

It remains now to justify the somewhat more careful technique required for determination of the complete stress-strain diagram by a review of the possible practical applications of the data. The uses may be divided into two fields:

I. Quantitative correlation of present uses of irons with different strain characteristics but similar strength.
II. Future Development.

I. Quantitative Correlation of Present Applications

Foundrymen and engineers have learned through long experience that differences in rigidity and toughness exist between white and gray irons; many proper applications have been developed on this *qualitative basis.* A brief review of the proven uses of the different structures on a quantitative basis permits a better correlation and understanding of the applications.

First, it should be decided how much plastic or elastic deformation is of importance. The *white and mottled irons* with low plasticity may therefore be considered first.

In practice, a mottled iron generally is conceded to be tougher than a white iron. In parts subject to wear, mottled irons are used successfully in the critical sections where white alloy iron would break. Figure 6 indicates that although a mottled iron (44-435-3-in. Y) is practically the same strength as a white iron, the total elongation at rupture is twice as great and plastic deformation is markedly higher (0.0505 vs. 0.0076 per cent).

Other applications where elongation rather than strength may be of paramount importance are numerous, for example, the general case of a part operating with a thermal gradient. If castings of the aforementioned irons are seated rigidly in place, elongation of 0.038 per cent is required per 100° F. temperature difference per inch.

Mottled iron will withstand over 0.3 per cent in tension or 800° F. differential, while white iron fractures at 0.15 per cent or 400° F. differential. From another point of view, white iron of 90,000-psi. tensile strength would be needed in this application to withstand the same deflection as the mottled iron.

High strength irons (60-90,000 psi.) with high elastic elongation (0.4-0.5 per cent) and some plastic deformation (0.1-0.2 per cent) have well established service records as crankshaft and camshaft materials. In this field, where the majority of failures occur in fatigue, very great ductility is of little importance. A steel with 20 per cent elongation in a static test obviously will exhibit no greater deformation than gray iron in a fatigue failure, and possesses the disadvantages of greater notch sensitivity and lower damping capacity.

Intermediate strength irons of machine tool type provide an interesting range from the fully pearlitic—tensile strength 50,000 psi., 0.70 per cent combined carbon, to the ferritic pearlitic—30,000 psi., 0.40 per cent combined carbon. As the combined carbon is decreased, softer and higher plastic elongation materials are provided with a decrease in strength and modulus.

Thermal shock resistant, ductile pearlitic-ferritic irons with below 0.20 per cent combined carbon are a logical progression from the pearlitic irons. For glass molds and ingot molds a given strain rather than strength is necessary, as explained in the section under white irons. Brake drum materials also require higher deformation and lower modulus, although combined carbon is higher (similar to 45-283-3, Table 2 and Fig. 2).

High elongation, low modulus austenitic irons offer an interesting combination of properties.

II. Future Development

A. Effect of Structure upon Stress-Strain. Since strain evidently is complementary to stress in determining serviceability, it may be profitable to evaluate further the effects

Fig. 14—Photomicrograph showing austenitic gray iron (austenite, graphite, carbide) (L 48801).

Microhardness readings—upper left, lower right: Vickers hardness numbers, 165, 165 (austenite), 940 (carbide). Other two impressions show effects of duplex structure upon penetration.

Chemical analysis: TC, 2.93; CC, 0.97; GC, 1.96; Mn, 1.52; P, 0.175; S, 0.02; Si, 2:03; Ni, 14.28; Cr, 1.82; Cu, 6.04 per cent.

Section of tensile specimen M407. Etchant—3 per cent Nital, 8 sec. 500X. BHN. 101.



of structure upon both variables. The series of irons presented in this paper obviously is incomplete; the quantitative effects of graphite amount and distribution and of other matrices may well receive attention. The effects of phosphorus and sulphur should also be carefully defined.

Other mechanical properties of cast iron may be correlated and understood by stress-strain observations. For example, the impact strength of both the austenitic and the high strength acicular irons is in excess of 120 ft. lb. (Izod 1.2 in cast bar), yet the reasons underlying the value may be different, such as high plastic elongation in the former case, high elastic elongation and high strength in the latter.

However, from the stress-strain curve excellent strength under *repeated* impact can be predicted for the more elastic material and diminishing resistance for the less elastic. The higher damping capacity of the more ductile irons compared with the less ductile materials may also be explainable upon the basis of the greater plastic deformation at a given stress.

The loosely used term "toughness," which impact tests are intended to evaluate, may be better defined by further work as consisting of a strain and a strength component with perhaps additional consideration of notch sensitivity.

Conclusion

A survey of the different types of stress-strain curves encountered in white and gray irons indicates that strain behavior may be complementary to strength in determining the serviceability of castings. Because of the severe effects of shrinkage encountered in certain specimens prepared from standard 1.2 in. arbitration bars, it is suggested that future work of this type be conducted upon adequately fed test bars.

Acknowledgment

The authors are greatly indebted to the entire staff of the Metallurgical Department of the American Brake Shoe Co., especially to R. H. Schaefer, Earnshaw Cook and H. S. Avery for helpful criticisms and suggestions. They also appreciate greatly the metallographic work of Mrs. Eva Rule and the careful mechanical testing conducted by Misses M. A. Moran, Elaine Reuhl and S. A. Decker. The assistance of D. J. Reese of the International Nickel Co. in providing certain specimens and in criticising the work is also gratefully acknowledged.

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DISCUSSION

Presiding: R. G. McELWEE, Vanadium Corp. of America, Detroit, Mich.

Fig. 15—Photomicrograph showing austenitic gray iron (austenite, graphite, carbide L 49001).

Microhardness readings—left to right: Vickers hardness numbers, 390, 465 (carbide + austenite), 243, 222, 212 (austenite). Note slip lines in austenite adjoining impressions.
Chemical analysis: TC, 2.89; CC, 0.85; GC, 2.84; Mn, 1.21; P, 0.09; S, 0.03; Si, 1.81; Ni, 18.38; Cr, 2.28; Cu, 2.74 per cent.
Section of tensile specimen M411. Etchant—3 per cent Nital, 8 sec. 500X. BHN. 140.



Fig. 16—Photomicrograph, mottled iron no alloy (carbide, graphite, spheroidized pearlite (L 47501).

Microhardness readings—left to right: Vickers hardness numbers, 1200, 1270, 1270 (carbide), 200, 217 (ferrite + carbide), 137, 150 (ferrite).
Chemical analysis (approx.): TC, 3.50; Mn, 0.55; Si, 0.55 per cent.
Section of spheroidized chilled iron L475. Etchant—3 per cent Nital, 8 sec. 500X.



Co-Chairman: W. E. MAHIN, Armour Research Foundation, Illinois Institute of Technology, Chicago, Ill.

T. E. EAGAN[1] *(written discussion):* The authors of this paper have offered a considerable amount of fundamental data on cast iron. It is this type of information that is of great value to the design engineer in that he is given the physical constants that he can use effectively.

In addition to this, the authors in their study of the micro-hardness of the various micro-constituents have given a wealth of information to the metallurgist who is interested in cast iron. I believe that this is the first time these hardnesses have been presented in a concise form.

The modulus of elasticity is one of the physical constants that are most important to a design engineer. He uses this in his calculation of stiffness and stress. As the authors point out, this so-called constant is not constant in gray iron but changes with increasing stress.

The authors by determining the tangent modulus have given a value which is perhaps the best one to use. It no doubt is a safe one. It may also be pointed out that the actual value obtained is influenced considerably by the accuracy of the strain measurements.

The authors by the use of SR4 strain gages have taken very accurate measurements; hence the values they show are relatively low when compared to some of the results given elsewhere. The results given compare favorably with those obtained in our laboratory on comparable material.

Cast iron is usually considered a brittle material. This has come about because it does not have a great deal of plastic or elastic elongation under stress and should it break, the fracture is flat with no visible signs of any elongation or deformation. The degree of brittleness is one that needs consideration. Almost all impact tests such as Izod, Charpy, or modifications of these tests measure the amount of energy required for fracture.

No great amount of study has been made of the amount of deformation before fracture. Any great amount of plastic movement of the material will in general make the piece useless in that this will most likely affect mating parts, etc. Hence, the impact breaking strength does not have significance to the designer except in a relative way.

The important thing to the designer is the total amount of movement the material will have under a given stress. Knowing these values for various materials will help the designer choose the right material for the particular stress conditions. Brittleness will only come into the picture under very sudden impact stresses. Such impact stresses are not normal in most design considerations. Hence, the information on plastic and elastic strains given by the authors is valuable.

Chemical analyses were made using ⅛ to 1/16-in. cubes instead of drillings. Were these cubes machined from the test bars, or were they cast from the metal?

The authors should be congratulated on presenting an important subject in a clear and concise form.

H. W. LOWNIE, JR.[2] *(written discussion):* The authors are congratulated for what is, in my opinion, the most constructive and informative paper presented at the current convention. More frequent papers containing factual information of this type are sorely needed by the industry. The cast irons selected for testing have been so well chosen that the data are unusually comprehensive.

In Fig. 1, it has been indicated that unloading of the tension-test specimen from a stress just below the tensile strength will result in a reduction of strain along the dotted line until 0.318 per cent plastic strain remains at zero load. I should like to inquire how this unloading curve was determined. The confusion of plastic elongation with elastic elongation in the caption of Fig. 1 has probably been noticed by the authors.

In Part III and Fig. 2A, the difference obtained between the stress-strain curve for specimens cut from standard arbitration test bars and the risered test bars should not be overlooked by the reader. Independent data also have indicated that the traditional form of test bar leaves much to be desired.

The spread of data obtained in many investigations may well be caused simply by small amounts of unsoundness in the test bars employed and not to certain "mysterious" influences often blamed for the spread. Investigations conducted with a large number of standard arbitration test bars cast by various methods have indicated that the majority of such bars cast in commercial foundries by the recognized standard methods are probably unsound.

This unsoundness is a matter of degree, but is usually sufficient to yield erratic test results and often is enough to mask the effect of various factors that are being investigated. Unsound test bars also penalize the foundry by forcing the production of a higher strength iron in order to meet minimum tensile strength specifications.

Foundrymen who consider that their own test bars are sound might well cast ten or fifteen of their standard test bars from the same ladle of iron and determine the spread in tensile strength obtained by pulling one tension specimen from each bar. The results will probably be disconcerting.

R. W. LINDSAY[3]: I would like to compliment the authors with regard to the data that have been collected, and the presentation of it by Dr. Flinn. The subject of elastic and plastic deformation of cast iron is of definite importance. The behavior of different cast irons toward sudden thermal changes is an example of this importance.

I would like to make just a brief, additional comment. Some time ago, Massari[3] studied the stability of combined carbon in low silicon iron of the type used in the manufacture of chilled-iron car wheels when such iron was annealed for increasing lengths of time at various temperatures.

His investigation pointed out the effects of increasing amounts of ferrite (hence decreasing amounts of combined carbon) in the matrix in relation to the stress-strain behavior. This is another illustration of the importance of elastic and plastic deformation of gray iron where it relates to the mounting of railroad car wheels on axles. An iron of low combined carbon content and high plastic deformation (or set) is to be avoided in this case.

The present paper makes an important contribution by extending these studies to differences in occurrence of carbon and differences in matrices.

MESSRS. FLINN AND CHAPIN *(authors' closure):* We wish to thank Messrs. Egan, Lindsay, and Lownie for their thoughtful discussion.

In answer to the sixth paragraph of Mr. Eagan's comment, it should be noticed that while cast iron does not have as much *plastic* elongation as steel, it has greater *elastic* elongation than steel of the same strength. For example in Fig. 1, the elastic elongation for gray iron is 0.404 per cent at 41,100 psi. while a steel specimen under this load would have a maximum elastic elongation of: 41,100 psi. × 1/30,000,000 psi. = 0.137 per cent.

This greater elastic deflection may be favorable or not, depending on whether the application requires a given strain or a given stress. In reply to the ninth paragraph of the same discussion, the chemical analyses were made using ⅛ to 1/16-in. fragments machined *from the gage length* of the tensile specimen. This procedure avoids loss of graphitic carbon and will be discussed fully in a later publication.

Regarding Mr. Lownie's comment, we are inclined to agree with his stand that the present form of test bar is inadequate in providing reliable data especially for irons which require careful control of directional solidification such as white iron, and the higher strength and alloyed gray irons.

In our own experimental work we have practically abandoned the standard test bar in favor of the properly risered specimens described in the paper. We have also found it necessary to adopt non-standard test specimens to determine the effect of different section sizes encountered in the foundry which possess cooling rates far beyond those of the standard 0.875, 1.2 and 2.0-in. arbitration bars.

Concerning the determination of the unloading curve, two methods may be used:

1. The load is removed every 2,000 psi. as the breaking load is approached and the resulting points plotted on the X-axis. When the final elongation (ϵ_{tot}, Fig.1) is obtained, a line is drawn parallel to the last unloading line and the X-intercept taken as plastic elongation.

2. When the specimen breaks just beyond the gage, a final reading is obtained at zero load to give plastic elongation. At the low elongations dealt with, elongation is quite uniform over the entire gage length.

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HOMOGENIZATION HEAT TREATMENTS FOR CAST STEEL

Part I

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IT IS COMMON PRACTICE to "homogenize" the heterogeneous, coarse-grained structure developed during the solidification of a steel casting. It is generally assumed that the treatment refines the grain, diffuses segregated areas resulting from the dendritic mode of freezing, and improves the mechanical properties, particularly ductility and toughness.

Homogenizing heat treatment is generally carried out at temperatures considerably higher than those used for normal heat-treating practice, but there is no consistency among producers in regard to the temperature or the holding time at temperature. Thus, it is not unusual to find one steel foundry homogenizing at 1700° F. for 4 hr. and another at 2000° F. for 10 hr. for castings of the same section size.

While composition of the steel and heat-treating facilities govern to some extent the choice of homogenizing heat treatment, there can be no doubt that variations in time and temperature as great as those mentioned are much beyond the normal range imposed by these factors. The implication to be derived from this decided lack of uniformity in homogenizing heat-treating practice is that there is no unanimity of opinion in the cast-steel industry regarding the value of homogenizing; yet the advantages to be gained if such opinion could be brought to closer agreement are perfectly obvious.

If it is true that a drastic heat treatment of many hours at high temperatures is necessary, then certainly those who now employ the lower temperatures and shorter holding times are not realizing the maximum properties from their steels. On the other hand, if there is no particular advantage in going to higher temperatures and longer times, or if the advantage is so slight that a minor adjustment in composition would compensate for it, then the drastic heat treatments could well be curtailed to save time and heat-treating capacity.

Survey Made

To obtain a better understanding of the worth of homogenizing heat treatments and to accumulate a set of data that would be useful to the cast-steel producers in governing their heat-treating practice, an extensive survey of the homogenizing heat treatment was undertaken at Battelle Memorial Institute on 36x36x2-in. cast-steel plates. These plates, which were submitted by five producers, were sectioned into coupons and each coupon given a different homogenizing heat treatment.

Range of temperature and holding time covered the extremes currently existing in actual practice, and in one instance exceeded the upper commercial homogenizing temperature. After quenching and tempering, each coupon was cut into tensile and notched-bar specimens and into specimens suitable for macro- and micro-examination.

Hardenability tests were made on standard end-quench specimens machined from the coupons after they were homogenized. Correlative studies were concerned with temper-brittleness susceptibility and low-temperature toughness. The final result was a comprehensive survey of the hardenability and mechanical properties of the five commercial plates after a variety of homogenizing heat treatments.

Procedure. The five plates used

Table 1

CHEMICAL ANALYSIS OF THE FIVE TEST PLATES

Plate No.	Components, per cent											
	C	*Mn*	*Si*	*P*	*S*	*Cr*	*Ni*	*Mo*	*Sn*	*Cu*	*B*	*Acid Soluble Al*
1	0.27	1.50	0.44	0.020	0.023	0.32	0.05	0.16	0.014	0.17	<0.0005	0.02
2	0.23	1.24	0.36	0.033	0.028	0.29	0.08	0.41	0.015	0.13	<0.0005	0.05
3	0.25	0.96	0.55	0.030	0.046	0.47	0.44	0.41	0.016	0.18	<0.0005	0.05
4	0.40	1.69	0.47	0.022	0.022	0.10	0.04	0.41	0.027	0.25	<0.0005	0.09
5	0.32	0.86	0.34	0.046	0.038	0.54	0.44	0.37	0.008	0.18	<0.0005	0.01

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This is Part I of an investigation conducted at Battelle Memorial Institute. The second and concluding installment follows on page 171.

for this study and their analyses are listed in Table 1. The general procedure in handling these plates was to cut coupons from each of the five 36x36x2-in. plates and to homogenize them at temperatures of 1650, 1750, 1850, 1950, or 2050° F. for 2, 6, or 12 hr., and at 2250° F. for 12 hr. The latter heat treatment was given after the tests on all the other heat-treated bars had been completed. Because of its severity, it cannot be classed as a commercial heat treatment, but was included for comparison purposes.

In the discussion, the results obtained from this homogenizing heat treatment are generally considered separately from those obtained after the other homogenizing heat treatments. After an end-quench hardenability specimen and specimens for micro- and macro-examination were removed from each coupon, the remainder of the coupon was reheated to the austenitizing temperature and water quenched and tempered to 240 to 260 Brinell.

Cooling Rate

Part of each coupon was water quenched from the tempering temperature and the other part was furnace cooled to determine the effect of cooling rate from the tempering temperature on the notched-bar toughness (temper-brittleness susceptibility). Charpy-type V-notch specimens were removed from the furnace-cooled part of the coupon, while Charpy-type V-notch, Izod, and tensile specimens were taken from the water-quenched portion.

Each of the original plates was cut into at least 19 coupons, measuring 5¾x7 in. A sketch of a sample layout is given in Fig. 1. All of the plates were sectioned in essentially the same manner. The numbering system in the figures identifies the position of each coupon in the original plate.

Location of the test specimens in each coupon is shown in Fig. 2. Beginning at the left side of the sketch, the test specimens are located as tabulated below the sketch. It should be noted that as far as possible the specimens were taken from material near the surface of the coupons, reducing the probability of having porous areas in the specimens.

The coupons received the homogenizing heat treatments in an electric resistance furnace with no attempt made to control the atmosphere. They were then air cooled. After homogenizing, the coupons had a portion removed which was large enough for the machining of an end-quench bar, two fracture grain-size bars and a macro- and micro-specimen (right-hand portion shown in Fig. 2).

The remainder of each coupon was held for 4 hr. at 1600° F. in a metal box sealed with a graphite lid, removed, and then water quenched, using a submerged spray. The time in the quenching medium was 1¾ to 2 min. After this time interval, the coupon was cold enough to be removed from the fixture by hand.

One coupon from each plate was quenched without receiving previous homogenizing treatment, thus providing test specimens for unhomogenized material. Each of the as-quenched coupons was cut into two portions, and after tempering both portions of each coupon to the 240 to 260 Brinell range, one portion was water quenched after the draw (left-hand portion in Fig. 2) and the other was furnace cooled (center portion in Fig. 2).

To obtain accurate hardness measurements during the heat-treating operations, as-quenched hardness readings were taken on a surface ground below the decarburized layer. This surface was also used for measuring the hardness in tempering the coupons to the 240 to 260 Brinell range.

At the completion of the tempering operations, a minimum of 1/16 in. was shaped off the cope and drag

The effects of homogenizing heat treatments of 2, 6, or 12 hr. at 1650, 1750, 1850, 1950, or 2050° F. and of 12 hr. at 2250° F. on the hardenability and mechanical properties of five hardened and tempered commercial 2-in. cast plates from five different producers were determined. The plates varied in hardenability from one air-hardening type to two plates which did not harden fully through the 2-in. section. No significant effects or trends from the homogenizing treatments were observed on the austenitic (fracture) grain size, hardenability, notch-bar toughness on V-notch Charpy specimens at room and sub-atmospheric temperatures, temper-brittleness susceptibility, or tensile properties. The properties of unhomogenized specimens were generally equivalent to those from homogenized material. Any deviations that arose could be more closely connected to the position of the test pieces in the original plate than with any influence of the homogenizing heat treatment. The position of the test pieces in the original plate was important because variations in soundness of the castings from one part to another were reflected in the mechanical properties. Successive increases in temperature or holding time brought about a gradual diffusion of the dendritic pattern on etched macrospecimens; however, even the most drastic heat treatment did not obliterate the structure. Segregated areas in the microstructure, represented by nonuniform etching rates and nonuniform distribution of the microconstituents, were also broken up by the more drastic heat treatments in three of the steels. The effect was noticeable to a lesser degree in the fourth steel and completely absent in the fifth. Despite these changes in the appearance of the homogenized structure, after quenching and tempering no difference could be observed in the microstructure in even the extreme conditions of no homogenization whatsoever and a treatment of 12 hr. at 2250° F. It was generally concluded that, in the case of these five low-alloy steels, the various homogenizing treatments produced no appreciable change in the characteristics of the hardened and tempered specimens.

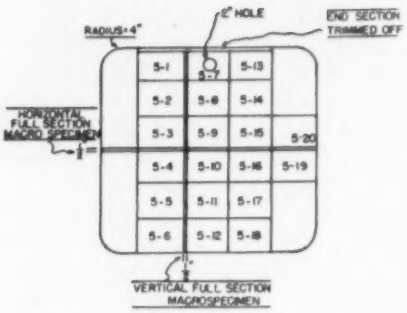


Fig. 1—Typical layout of test coupons on the cast plates. Cope side of Plate 5 is illustrated.

surfaces of the coupons to remove the decarburized layer before cutting out any of the specimens.

Some difficulty was experienced with quenching cracks in coupons from Plate 4. An interrupted quench was tried as a means of eliminating the cracks, but this treatment was not entirely successful. No further cracking was encountered when the coupons were quenched 1½ min., followed immediately by a draw at 600° F. Table 2 gives the complete heat-treating data.

A V-notch Charpy-type bar was chosen for the notched-bar tests over a temperature range of —60 to +75° F. The bars were notched on the side facing the center of the plate to include as much of the sound steel as possible within the cross section below the V.

Testing temperatures from 0° F. to —60° F. were obtained with a mixture of commercial acetone and dry ice. Specimens were held at the testing temperature for a minimum of 15 min. An interval of 5 sec. was used between the time the specimen was removed from the bath and the time the hammer was permitted to fall. To allow for heat absorption by the specimen during this 5-sec. interval, the bath was under-cooled for the various temperatures as follows:

Undercool, °F.	*Testing Temperature, °F.*
2	0
3	—20
3	—40
4	—60

Duplicate test specimens were broken at each of the testing temperatures. These duplicate specimens represented the cope and drag side of the plate.

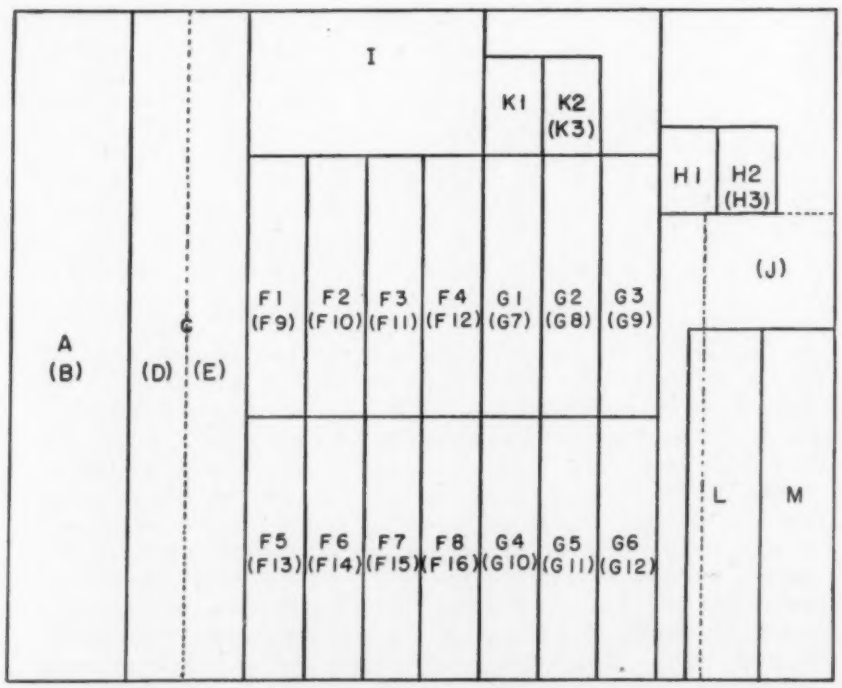
Besides the tests on the Charpy-type bars, two standard triple-notch Izod bars were tested at room temperature. These bars were cut from sections D and E shown in Fig. 2.

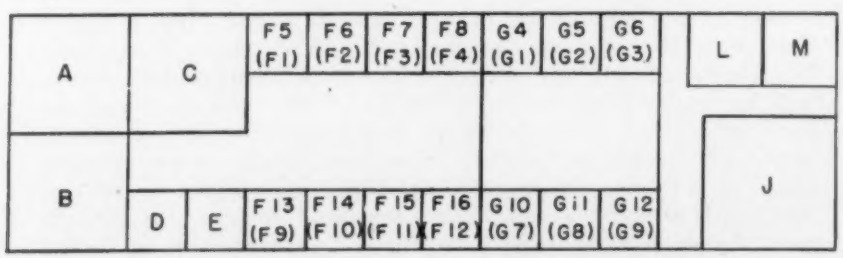
Sections A, B, and C in Fig. 2 were machined into standard 0.505-in. diameter, 2-in. gage length, tensile bars. These bars were tested with a load application rate of 5,000 lb. per min. A Baldwin-Southwark, Templin type, automatic-recording extensometer was used to obtain a stress-strain curve for determining the yield load at 0.2 per cent strain.

Cross-section macrospecimens, measuring the full length and width of the 36x36-in. plates, were taken from each of the five plates in the as-received condition. Smaller cross-section macrospecimens were taken from each coupon after it had received its particular homogenizing treatment, and also after it had been quenched and tempered (K1 and H1 bars in Fig. 2).

All of the macrospecimens were etched in a solution of 50 per cent water, 38 per cent hydrochloric acid and 12 per cent sulphuric acid, which was heated to 170° F. The time required to produce a satisfactory etch was about 30 min.

Examination of the macrospeci-





Portion of Coupon Water Quenched from Tempering Temperature	
Tensile bars	A, B, and C
Izod bars	D, E
Charpy bars	F1 to F16
Hardness-survey block	I

Portion Furnace Cooled from Tempering Temperature	
Charpy bars	G1 to G12
Macrospecimen	K1
Microspecimens	K2 and K3
Fracture grain-size bar	M

Specimens Homogenized Only	
Macrospecimen	H1
Microspecimens	H2 and H3
Fracture grain-size bar	L
Jominy bar	J

Fig. 2—Detail of sectioning of each 5¾x7-in. coupon. Numbers in parenthesis are numbers of specimens directly below or behind the specimens shown in the drawing.

Table 2

TEMPERING TREATMENTS AND HARDNESSES OBTAINED ON ALL COUPONS

Coupon* Number	As-Quenched Bhn.	Final Draw hr.	Final Draw ° F.	Brinell Hardness No. Water Quenched from Draw	Brinell Hardness No. Furnace Cooled from Draw
HOMOGENIZING TREATMENT—NONE					
1-9	448	1	1200	239	245
1-13	467	1	1200	250	230
2-13	441	¼	1250	252	240
3-9	477	½	1250	256	246
4-9	514	¾	1250	255	248
5-19	507	1⅓	1250	256	242
HOMOGENIZE, 2 HR. AT 1650° F.					
1-7	415	2	1150	250	241
2-1	451	3	1200	259	241
3-1	451	½	1250	248	248
4-21**	485	1	1250	257	253
5-13	495	1	1250	259	246
HOMOGENIZE, 6 HR. AT 1650° F.					
1-14	457	1	1200	253	240
2-14	391	2	1150	259	256
3-14	451	½	1250	253	245
4-22**	471	1	1250	257	255
5-14	477	1	1250	256	253
HOMOGENIZE, 12 HR. AT 1650° F.					
1-2	415	2	1150	250	241
2-2	444	2	1200	259	241
3-2	454	½	1250	253	248
4-19**	477	¾	1250	257	246
5-2	485	1	1250	246	241
HOMOGENIZE, 2 HR. AT 1750° F.					
1-18	461	1	1200	252	252
2-18	444	2	1200	241	230
3-18	451	¾	1250	250	241
4-18	485	¾	1250	255	245
5-18	488	¾	1250	241	242
HOMOGENIZE, 6 HR. AT 1750° F.					
1-12	477	1	1200	259	256
2-12	444	2¾	1200	241	248
3-12	457	½	1250	256	246
4-12	471	½	1250	259	260
5-12	477	1	1250	255	242
HOMOGENIZE, 12 HR. AT 1750° F.					
1-3	444	1	1200	244	255
2-3	432	2	1200	256	244
3-3	474	½	1250	255	242
4-3	526	¾	1250	257	246
5-3	477	1	1250	255	252
HOMOGENIZE, 2 HR. AT 1850° F.					
1-17	451	2	1150	256	249
2-17	448	3	1200	257	246
3-17	474	½	1250	255	244
4-24**	477	1	1250	255	244
5-17	481	½	1250	257	250
HOMOGENIZE, 6 HR. AT 1850° F.					
1-11	461	1	1200	255	250
2-11	454	3½	1200	259	245
3-11	444	½	1250	250	245
4-11	492	¾	1250	248	249
5-11	488	1	1250	256	245
HOMOGENIZE, 12 HR. AT 1850° F.					
1-4	435	1	1200	252	245
2-4	415	2	1200	241	241
3-4	451	½	1250	255	255
4-4	514	¾	1250	240	241
5-4	451	1	1250	248	241
HOMOGENIZE, 2 HR. AT 1950° F.					
1-16	467	2	1150	256	245
2-16	451	2	1200	252	245
3-16	454	½	1250	255	260
4-23**	485	1	1250	253	253
5-16	477	1	1250	245	240
HOMOGENIZE, 6 HR. AT 1950° F.					
1-10	448	1	1200	255	249
2-10	406	1¾	1200	240	248
3-10	454	½	1250	255	242
4-10	503	¾	1250	245	249
5-10	467	1	1250	253	255
HOMOGENIZE, 12 HR. AT 1950° F.					
1-5	444	1	1200	250	248
2-5	426	2	1200	255	244
3-5	448	½	1250	257	250
4-5	454	½	1250	244	255
5-5	415	¾	1250	244	245
HOMOGENIZE, 2 HR. AT 2050° F.					
1-15	481	1¾	1200	252	249
2-15	385	2	1200	250	244
3-15	464	½	1250	253	255
4-15	551	¾	1250	248	242
5-15	474	1	1250	255	241
HOMOGENIZE, 6 HR. AT 2050° F.					
1-8	406	¾	1200	241	243
2-8	406	2	1200	253	242
3-8	417	½	1250	255	255
4-20**	467	¾	1250	256	255
5-8	467	1	1250	249	242
HOMOGENIZE, 12 HR. AT 2050° F.					
1-6	415	1	1200	242	244
2-6	411	2	1200	250	236
3-6	420	½	1250	249	255
4-6	448	½	1250	248	255
5-6	444	¾	1250	241	242

**Six coupons from No. 4 Plate are replacements for those which were badly cracked after being quenched. These six replacements were given a 1½-min. water quench and an immediate draw of one hour at 600° F. The value in the "As-Quenched Bhn." column was obtained after this preliminary draw.

*The first number is the plate number. The second number is the coupon number.

mens taken from the material that had been quenched and tempered did not show anything that could not be seen in the as-homogenized macrospecimens, hence the quenched and tempered specimens were not photographed.

Microspecimens were taken from each plate in the as-received condition, and from each coupon after the homogenizing treatment and after the quenching and tempering treatment. The specimens were selected to present a cross-section surface extending from the center to the cope side of the plates.

Specimens taken from the various coupons after homogenizing were photographed at ×100 in the etched condition, at a spot ½ in. below the cope surface. Only representative photographs of these specimens are illustrated in this report. Examination of the microspecimens of the coupons after quenching and tempering showed so little variation that only the specimens that had received the most drastic homogenizing treatment and those that had received no homogenizing treatment were photographed.

Grain Size Determinations. The fracture grain-size bars ("M" bar in

Table 3

FRACTURE GRAIN SIZES OBTAINED ON WATER-QUENCHED BARS HEATED WITH THE COUPONS TO 1600° F. AND HELD 4 HR.

Homogenizing Time, hr.	*Homogenizing Temperature, ° F. 1650*	*1750*	*1850*	*1950*	*2050*	*2250*
PLATE NO. 1—GRAIN SIZE BEFORE HOMOGENIZING—6						
2	6½	6½	6½	6½	6½	—
6	6½	6½	6½	6½	6½	—
12	6½	6½	6½	6½	6½	7
PLATE NO. 2—GRAIN SIZE BEFORE HOMOGENIZING—7						
2	6½	6½	6½	6½	6½	—
6	6½	6½	6½	5½	6½	—
12	6½	5½	6½	6½	6½	6
PLATE NO. 3—GRAIN SIZE BEFORE HOMOGENIZING—5½						
2	5-7 Mixed	5-7 Mixed	5-7 Mixed	5-7 Mixed	5-7 Mixed	—
6	5-7 Mixed	5-7 Mixed	5-7 Mixed	5-7 Mixed	5-7 Mixed	—
12	5-7 Mixed	5-7 Mixed	5-7 Mixed	5-7 Mixed	5-7 Mixed	5½
PLATE NO. 4—GRAIN SIZE BEFORE HOMOGENIZING—7						
2	6½	6	6	5½	6	—
6	6½	6	5½-6½	5½	5½	—
12	6½	6	5½	6	6	7-0 Mixed
PLATE NO. 5—GRAIN SIZE BEFORE HOMOGENIZING—4						
2	2½	3	2½	2	2½	—
6	2½	2½	2 & 7	2 & 7	2½	—
12	2	1½ & 8	2	2	2½	4½

Table 4

FRACTURE GRAIN SIZES OBTAINED ON END-QUENCH BARS

(All bars quenched from 1600° F. after holding 20 min.)

Homogenizing Time, hr.	*Homogenizing Temperature, ° F. 1650*	*1750*	*1850*	*1950*	*2050*	*2250*
PLATE NO. 1—GRAIN SIZE BEFORE HOMOGENIZING—6						
2	7	6½	6	5½	5½	—
6	7	6	6½	6	6	—
12	6	7	5½	6	6	6
PLATE NO. 2—GRAIN SIZE BEFORE HOMOGENIZING—6						
2	6½	6	6	6	6	—
6	6	6½	6	5½	5½	—
12	6½	5½	5½	5	7	6
PLATE NO. 3—GRAIN SIZE BEFORE HOMOGENIZING—7						
2	6	5½	5	5	5	—
6	5½	5	5	5	5	—
12	5½	5	5	5½	5	6
PLATE NO. 4—GRAIN SIZE BEFORE HOMOGENIZING—7						
2	7	7½	7	6	6½	—
6	7½	7½	7½	7	6	—
12	7	7	7	7	7½	7-0 Mixed
PLATE NO. 5—GRAIN SIZE BEFORE HOMOGENIZING—5						
2	5	5	4	5½	5	—
6	4½	4	4	4	4½	—
12	4½	4	4	4½	4½	5

Fig. 2) cut from each coupon were given the same treatment as the coupons, namely, 4 hr. at 1600° F., followed by a water quench. The fractures of the broken bars were compared with a set of Shepherd standard fractures and the grain size rated.

Exclusive of the data for the 2250° F. homogenize, the results in Table 3 show that Plate 5 was the only plate to have a coarse austenitic grain size after holding for 4 hr. at 1600° F. before water quenching. The steel in this plate was understood to have been deoxidized with silicon only.

Plates 1, 2, 3, and 4 had a much finer grain size, ranging from No. 5 to No. 7. Plate 3, which was understood to have been deoxidized with an intermediate amount of aluminum (under 1 lb. Al per ton), consistently showed a mixed grain size of No. 5 and No. 7, except in the coupons which received no homogenizing treatment. This coupon had a grain size of No. 5½.

Plates 1, 2, and 4 were understood to have been deoxidized with at least 2 lb. Al per ton. It will be noted that although the various homogenizing treatments established no noticeable trend in grain size, the coupon which received no homogenizing treatment had a tendency toward a finer grain size than the coupons receiving the treatments. This trend is exhibited in Plates 2, 4, and 5.

Further, steels from Plates 1, 2, and 4, deoxidized with a minimum of 2 lb. Al per ton, tended to have a finer grain size than either steel from Plate 3, deoxidized with an intermediate aluminum addition, or the silicon-deoxidized steel from Plate 5. The grain-size characteristics of the steel were in accordance with those expected from the deoxidization practices employed.

Grain Size Changes

Table 3 shows that the more noticeable changes in grain size resulting from the 2250° F. homogenizing heat treatment took place in the coupons from Plates 3, 4, and 5. The effect of this treatment on Plate 4 is the most outstanding since it produced a mixed grain of Nos. 7 and 0. The low grain-coarsening temperature indicated by this mixed grain is unusual for an aluminum-killed steel quenched from 1600° F.

Additional fracture grain-size bars ("L" bar in Fig. 2) were heated and quenched with the hardenability bars, but because of the difficulty in achieving the same rate of heating for these fracture grain-size bars that was obtained with the end-quench bars, and because of the differences in holding time, the grain size of the fracture grain-size bars did not always match the grain size found in the hardenability bars receiving a similar preliminary heat treatment.

Herty and coworkers[1] have discussed the effect of heating rate on austenitic grain size. Their work indicated that, in certain instances, heating quickly through the transformation range increased the aus-

Table 5

END-QUENCH HARDENABILITY DATA

Homogenizing Treatment—Temperature, ° F.	Time, hr.	Coupon No.	Rockwell "C" 1/16 in. from Quenched End	Distance from Quenched End (1/16 in.) for Drop of 5 Rockwell "C"	Distance from Quenched End (1/16 in.) for Drop of 10 Rockwell "C"	Rockwell "C" at 40/16 in. from Quenched End
PLATE No. 1						
None	None	1-9	50	4.4	6.1	27
1650	2	1-7	49	4.2	5.7	25
1650	6	1-14	49	4.1	5.8	25
1650	12	1-2	49½	4.3	6.1	27½
1750	2	1-18	49	4.3	5.9	24
1750	6	1-12	49½	3.9	5.6	24½
1750	12	1-3	49½	4.6	6.6	27½
1850	2	1-17	49½	4.3	6.3	25½
1850	6	1-11	49	4.0	5.6	26½
1850	12	1-4	48½	4.9	6.5	25
1950	2	1-16	49	4.2	5.9	25½
1950	6	1-10	49	3.5	4.8	24
1950	12	1-5	49	3.6	5.1	25
2050	2	1-15	50	3.6	5.2	25
2050	6	1-8	48	4.1	5.6	24½
2050	12	1-6	49	4.2	5.8	25
2250	12	1-19	48½	4.3	5.7	23
PLATE No. 2						
None	None	2-13	48½	4.0	5.6	27
1650	2	2-1	47½	3.4	4.7	24
1650	6	2-14	48	3.7	5.0	25
1650	12	2-2	46	4.3	5.8	25
1750	2	2-18	47½	4.3	5.6	24
1750	6	2-12	47½	4.5	5.9	24½
1750	12	2-3	45½	3.6	5.0	22½
1850	2	2-17	48	4.1	5.6	24
1850	6	2-11	47	4.8	6.6	24½
1850	12	2-4	45½	4.1	5.7	21
2050	2	2-15	47	4.4	5.9	23½
2050	6	2-8	47	4.4	6.1	23
2050	12	2-6	46	4.2	6.0	23½
2250	12	2-19	45	4.6	6.1	23
PLATE No. 3						
None	None	3-9	49½	9.0	14.2	31½
1650	2	3-1	48½	8.7	12.7	29
1650	6	3-14	49	9.5	14.2	31
1650	12	3-2	48½	8.5	12.6	28
1750	2	3-18	49	8.6	13.4	31
1750	6	3-12	49	7.8	11.7	30½
1750	12	3-3	49	8.9	13.5	31
1850	2	3-17	49½	7.8	11.0	29½
1850	6	3-11	49	8.8	13.0	31
1850	12	3-4	49½	8.8	13.4	31
1950	2	3-16	49	9.1	13.7	31½
1950	6	3-10	48	9.4	14.4	30
1950	12	3-5	48½	8.6	12.8	31
2050	2	3-15	49	8.3	12.5	31
2050	6	3-8	48	8.8	13.2	30½
2050	12	3-6	48	8.5	12.9	30½
2250	12	3-19	48½	7.6	10.8	26½
PLATE No. 4						
None	None	4-9	58½	22.0	28.0*	50½
1650	2	4-21	58	38.0	—	58
1650	6	4-22	58	29.0	—	51½
1650	12	4-19	58	28.0	37.0*	51
1750	2	4-18	57½	24.0	32.0*	49½
1750	6	4-12	58½	26.5	34.0*	50½
1750	12	4-3	57	30.0	—	50½
1850	2	4-24	58½	27.0	39.0*	51½
1850	6	4-11	58½	23.5	—	52
1850	12	4-4	58½	28.0	38.0*	51
1950	2	4-23	57½	—	—	53½
1950	6	4-10	58	33.5	—	51½
1950	12	4-5	57	36.0	—	52
2050	2	4-15	59	26.0	—	52½
2050	6	4-20	58½	20.0	30.5*	51
2050	12	4-6	59	26.5	—	52½
2250	12	4-7	57	20.0	—	48½
PLATE No. 5						
None	None	5-19	51½	7.7	10.0	31
1650	2	5-13	51½	7.8	11.2	31½
1650	6	5-14	51	8.2	11.5	31
1650	12	5-2	50	8.0	12.2	31
1750	2	5-18	51	8.1	11.3	31½
1750	6	5-12	50½	8.0	11.2	31½
1750	12	5-3	50½	7.8	11.2	30
1850	2	5-17	50	8.5	11.7	31½
1850	6	5-11	50	8.4	11.8	30
1850	12	5-4	50	8.0	10.8	30
1950	2	5-16	51	8.5	11.6	31½
1950	6	5-10	50½	8.9	12.7	31½
1950	12	5-5	50½	9.1	12.8	31
2050	2	5-15	50½	8.5	11.6	31½
2050	6	5-8	51½	8.7	11.8	31½
2050	12	5-6	51½	8.5	11.3	30½
2250	12	5-20	51	9.5	13.0	29½

*Drop of 7 Rockwell "C"

tenite grain size, depending upon the particular deoxidization practice which was used on the steel.

To obtain correct values for the grain size of the end-quenched hardenability bars, they were deeply cut close to the quenched end and fractured in the martensitic zone. These fractures were than rated for the grain size of the bar (Table 4).

With the exception of Plates 3 and 5, the grain sizes in Tables 3 and 4 check closely. In Plate 3 the short holding time of 20 min. for the end-quench test (Table 4) was not sufficient to develop the mixed grains found after 4 hr. at temperature (Table 3). A similar explanation can be advanced for the differences in grain size of Plate 5 after the two treatments. In neither table is there an apparent trend in grain size resulting from the homogenizing heat treatments.

No great difference exists in the effects of the 2250° F. treatment as compared with the other homogenizing treatments, except in the case of Plate 4 where the mixed grain of Nos. 7 and 0 is again evident.

Hardenability Properties. The hardenability properties were investigated by means of end-quench tests,

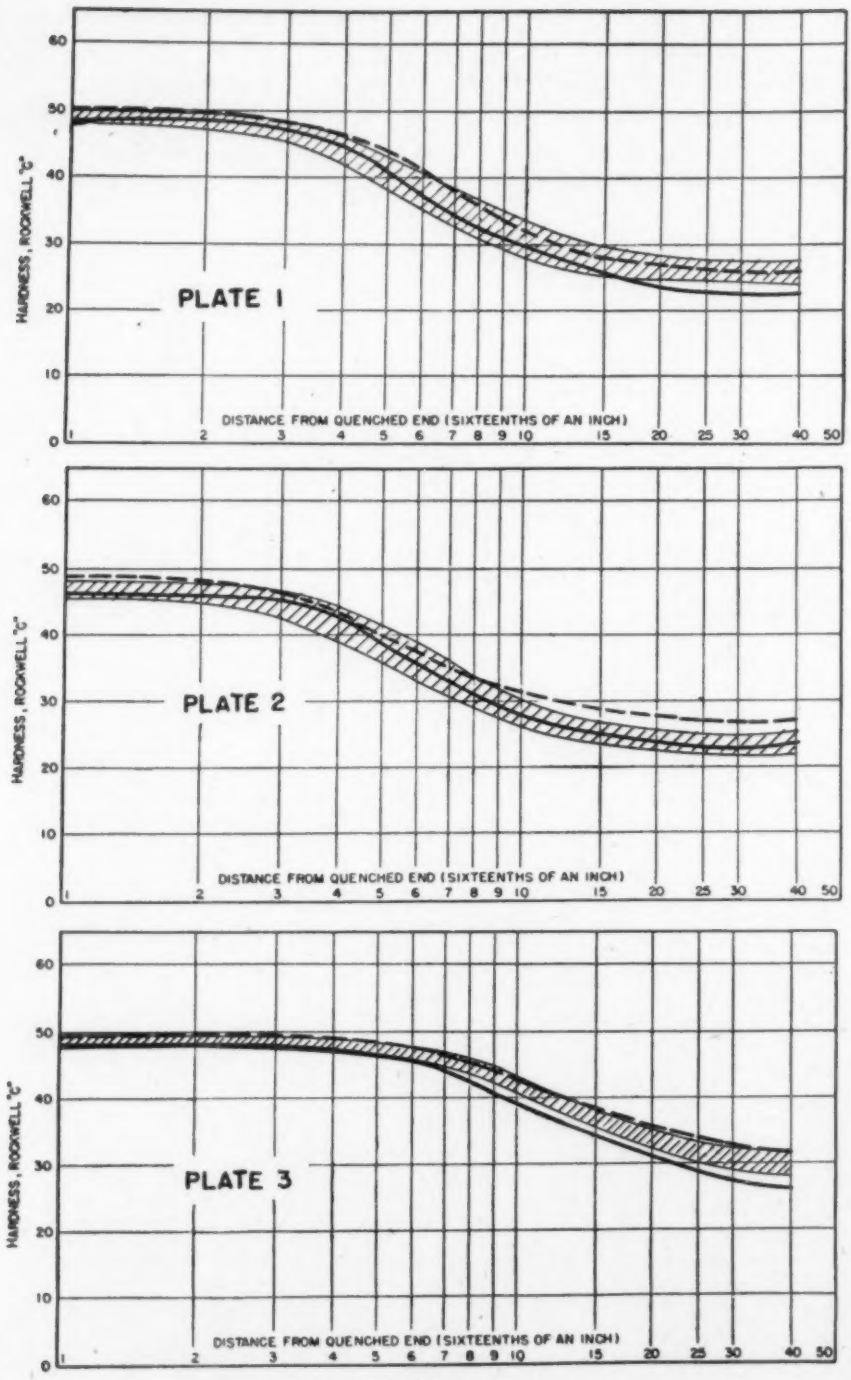


Fig. 3—(Continued on opposite page).

and cross-section hardness surveys made after quenching and tempering. The end-quench bars were the standard 3-in. long, 1-in. diameter bars. They were placed in protective graphite crucibles and heated to 1600° F. in 20 min. After holding for 20 min. at temperature, the bars were end-quenched, using the standard procedure.

The hardenability data are given in Table 5 by tabulating the hardness at 1/16 in. from the quenched end, the distances for a drop of 5 and of 10 Rockwell "C", and the hardness at 2½ in. from the quenched end. This tabulation of hardenability data in Table 5 permits a quantitative comparison of the results from each plate for each heat treatment.

It is obvious from Table 5 that homogenizing has not effected any change in the depth-hardening properties of the steels; within the limits of error, all of the bars from each steel have the same hardenability regardless of the homogenizing time or temperature.

Figure 3 shows the range of hardenability for each of the five steels after the various homogenizing heat treatments. In no case is this range broad, another indication that the homogenizing treatment has had no influence on hardenability.

The hardenability curves of the unhomogenized bars (and the bars homogenized at 2250° F. for 12 hr.) are also plotted in Fig. 3. The solid lines are the curves for the steels after homogenizing at 2250° F. for 12 hr., while the dashed lines are for the unhomogenized bars. The curves for the unhomogenized bars generally fall near the upper part of the range of the homogenized specimens.

It may be questioned whether this slight tendency toward higher hardenability on the part of the unhomogenized specimens was great enough to be significant. However, Parke and Herzig[2] have reported two instances where homogenization did result in lowered hardenability.

Hardenability Ratings

Hardenability of Plate 5 is increased somewhat by the 2250° F. homogenization heat treatment, but the hardenability ratings of the specimens from the other plates given the same treatment fall within the range estabished by the other homogenizing treatments. The hardenability curves show Plate 4 to be of the air-hardening type. The others have considerably lower hardenability properties.

The "I" block (Fig. 2) was used to reveal the total effect of quenching and tempering on the hardness distribution through the cross section of each coupon. These hardness distribution curves are plotted in Figs. 4 to 8, and the tables opposite the corresponding curves list the pertinent heat-treating data and surface hardness of each coupon.

No correlation can be made between the homogenizing treatments and the cross-section hardness surveys because the cross-sectional hardness appears to be influenced more by the position of the coupon in the original plate than by the homogenizing heat treatment. The cross-section hardness surveys instead give a picture of the hardness distribution which is characteristic of that

particular [illegible]rt of the original plate from whi[illegible] the section was removed. Th[illegible] also provide a means of evaluating more accurately the hardness o[illegible] the test bars which were removed from the coupons.

Note that Plate 1 (Fig. 4) shows a decided dip, Plate 2 (Fig. 5) a shallow dip, while Plates 3, 4, and 5 (Figs. 6, 7, and 8) have fairly flat curves. It may be deduced that Plates 1 and 2, since the hardenability curves (Fig. 3) show them to be of considerably lower hardenability than the rest of the plates, were never thoroughly hardened to the center and that the tempering treatment did not completely even out the differences in hardness from surface to center.

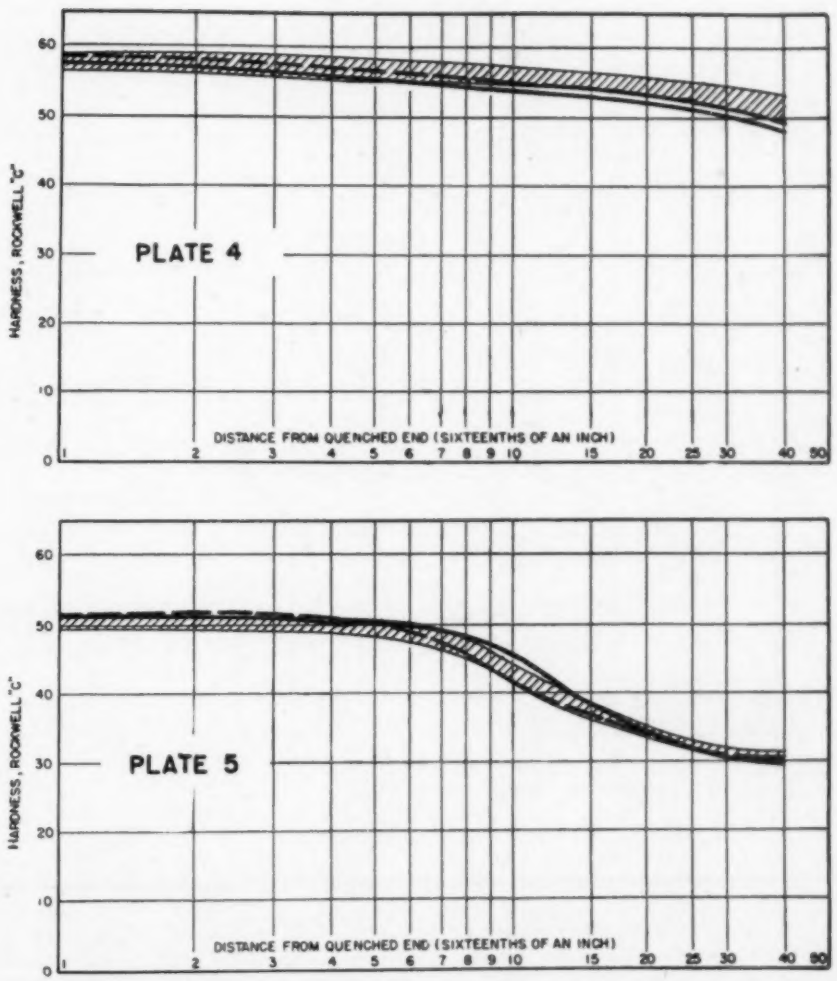
It is not unusual that Plates 1 and 2 did not harden fully since tests on submerged spray quenching used in this work have shown that the cooling rate at the center of a 2-in. plate is equivalent to that at 11/16 in. from the quenched end of an end-quench bar. Figure 3 shows that both Plates 1 and 2 had low hardnesses at this point.

Low-Temperature Notched-Bar Tests. A property required for certain applications is ability to withstand shock, not only at room temperatures but under subzero conditions as well. For this reason, a study of the toughness of the five steels as affected by the various homogenizing heat treatments was made a part of the research work.

Low-Temperature Properties

Closely allied with the study of low-temperature properties is the problem of temper-brittleness susceptibility, which is a latent factor whenever alloy steels are used after quenching and drawing. Temper brittleness is manifested by a loss in toughness after a tempering treatment which is followed by slow cooling to room temperature. It can be repressed by rapid cooling after tempering or by controlling the composition; the use of molybdenum is effective in this respect.

The study of temper-brittleness susceptibility was incorporated with the determination of low-temperature toughness by preparing duplicate sets of test bars, one set being taken from a portion of the coupon that had been water quenched after tempering and the other set from another portion that was furnace cooled at the rate of 100 to 115° F. per hr. after the final draw.



—— 12-hr. homogenize at 2250° F.
- - - No homogenizing treatment.
Quenching temperature, 1600° F. for all plates

Plate No.	C	Mn	Si	P	S	Cr	Ni	Mo	Sol. Al	Cu	Grain Size
1	.27	1.50	.44	.020	.023	.32	.05	.16	.02	.17	6
2	.23	1.24	.36	.033	.028	.29	.08	.41	.05	.13	6
3	.25	0.96	.55	.030	.046	.47	.44	.41	.05	.18	6-7
4	.40	1.69	.47	.022	.022	.10	.04	.41	.09	.25	7*
5	.32	0.86	.34	.046	.038	.54	.44	.37	.01	.18	5

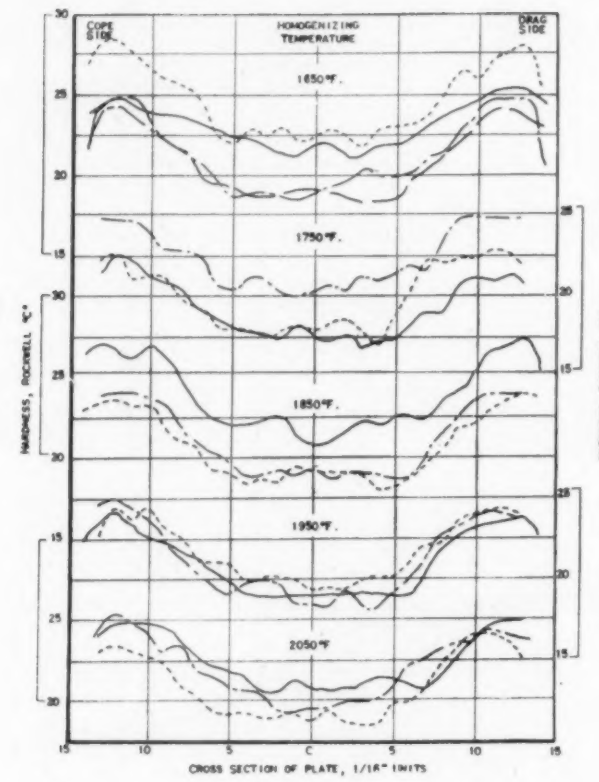
*Grain size was 7 and 0 after 2250° F. homogenize.

Fig. 3—Range of hardenability for each of the five steels after the various homogenizing heat treatments.

Originally, air cooling after tempering had been planned for use in revealing temper brittleness, because air cooling is common practice in the industry. However, a test on a pilot coupon proved that there was little difference in the toughness of the steel at low temperatures, whether water quenched or air cooled. This is shown by the curves in the left-hand graph in Fig. 9.

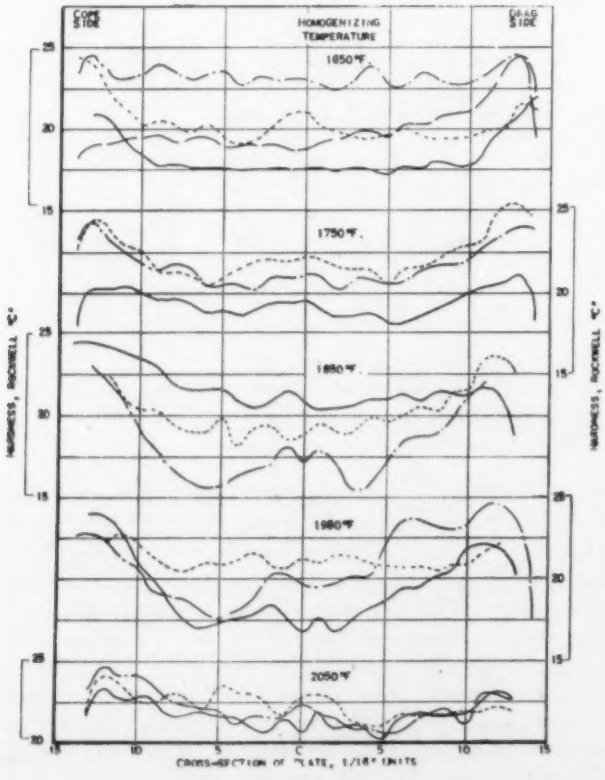
Furnace cooling rather than air cooling was therefore resorted to in an attempt to show up temper brittleness. The curves in Fig. 9 for the furnace-cooled specimens indicate that furnace cooling caused an appreciable reduction in toughness at sub-atmospheric temperatures. These results led to the adoption of furnace cooling instead of air cooling for all the coupons.

Actually, furnace cooling of the relatively small test coupons at the



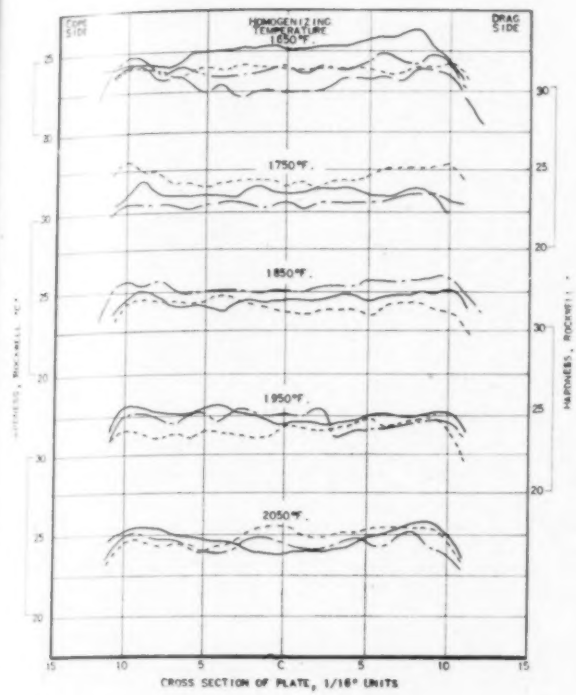
Coupon No.	Homogenizing Treatment, Hr.	° F.	As-Quenched Bhn.	Drawing Treatment, Hr.	° F.	As-Drawn Bhn. Cope	Drag	Code
1650° F. Homogenize								
1-9	None		448	1	1200	242	232	— —
1-7	2	1650	415	2	1150	250	249	——
1-14	6	1650	457	1	1200	244	231	— · —
1-2	12	1650	415	2	1150	259	256	· · · · ·
1750° F. Homogenize								
1-18	2	1750	461	1	1200	248	245	——
1-12	6	1750	477	1	1200	257	248	— · —
1-3	12	1750	444	1	1200	244	231	· · · · ·
1850° F. Homogenize								
1-17	2	1850	451	2	1150	256	253	——
1-11	6	1850	461	1	1200	245	244	— · —
1-4	12	1850	435	1	1200	224	245	· · · ·
1950° F. Homogenize								
1-16	2	1950	467	2	1150	241	245	——
1-10	6	1950	448	1	1200	250	232	— · —
1-5	12	1950	444	1	1200	239	236	· · · · ·
2050° F. Homogenize								
1-15	2	2050	481	1	1200	241	239	——
1-8	6	2050	406	¾	1200	231	234	— · —
1-6	12	2050	415	1	1200	246	229	· · · · ·

Fig. 4—Effect of homogenizing treatments on the hardness distribution within the cross section of quenched and drawn coupons from Plate No. 1



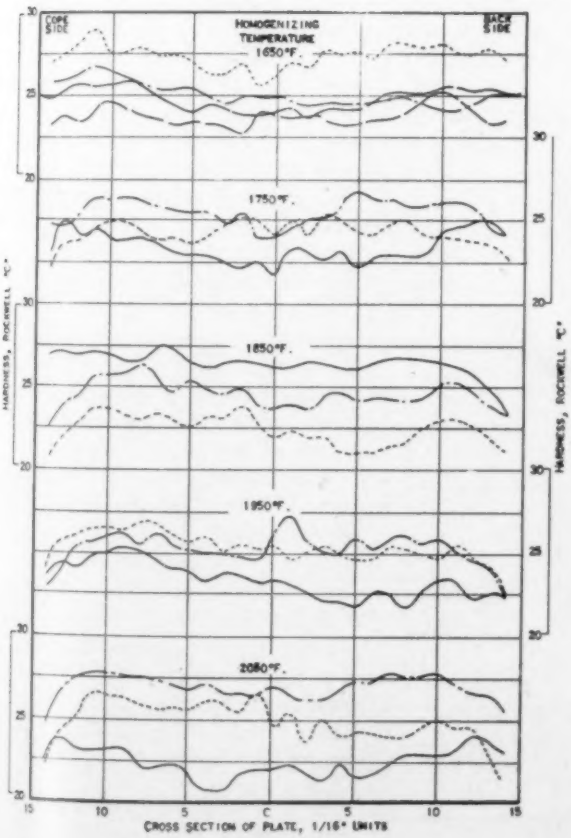
Coupon No.	Homogenizing Treatment, Hr.	° F.	As-Quenched Bhn.	Drawing Treatment, Hr.	° F.	As-Drawn Bhn. Cope	Drag	Code
1650° F. Homogenize								
2-13	None		441	¼	1250	245	253	— —
2-1	2	1650	451	3	1200	232	255	——
2-14	6	1650	391	2	1150	269	250	— · —
2-2	12	1650	444	¼	1250	252	248	· · · · ·
1750° F. Homogenize								
2-18	2	1750	444	2	1200	232	217	——
2-12	6	1750	444	2¾	1200	256	241	— · —
2-3	12	1750	432	2	1200	255	219	· · · · ·
1850° F. Homogenize								
2-17	2	1850	448	3	1200	255	246	——
2-11	6	1850	454	3½	1200	241	235	— · —
2-4	12	1850	415	2	1200	245	229	· · · · ·
1950° F. Homogenize								
2-16	2	1950	467	2	1150	253	215	——
2-10	6	1950	406	1¾	1200	255	229	— · —
2-5	12	1950	426	2	1200	242	237	· · · · ·
2050° F. Homogenize								
2-15	2	2050	385	2	1200	255	234	——
2-8	6	2050	406	2	1200	242	250	— · —
2-6	12	2050	411	2	1200	240	226	· · · · ·

Fig. 5—Effect of homogenizing treatments on the hardness distribution within the cross section of quenched and drawn coupons from Plate No. 2



Coupon No.	Homogenizing Treatment, Hr.	° F.	As-Quenched Bhn.	Drawing Treatment, Hr.	° F.	As-Drawn Bhn. Cope	Drag	Code
1650° F. Homogenize								
3-9	None		477	½	1250	248	217	— —
3-1	2	1650	451	½	1250	252	241	——
3-14	6	1650	451	½	1250	260	237	— · —
3-2	12	1650	454	½	1250	255	224	- - - - -
1750° F. Homogenize								
3-18	2	1750	451	¾	1250	241	239	——
3-12	6	1750	457	½	1250	245	242	— · —
3-3	12	1750	474	½	1250	245	234	- - - - -
1850° F. Homogenize								
3-17	2	1850	474	½	1250	241	245	——
3-11	6	1850	444	½	1250	241	250	— · —
3-4	12	1850	451	½	1250	235	245	- - - - -
1950° F. Homogenize								
3-16	2	1950	454	½	1250	255	253	——
3-10	6	1950	454	½	1250	244	232	— · —
3-5	12	1950	448	½	1250	248	235	- - - - -
2050° F. Homogenize								
3-15	2	2050	464	½	1250	259	250	——
3-8	6	2050	417	½	1250	248	248	— · —
3-6	12	2050	420	½	1250	255	239	- - - - -

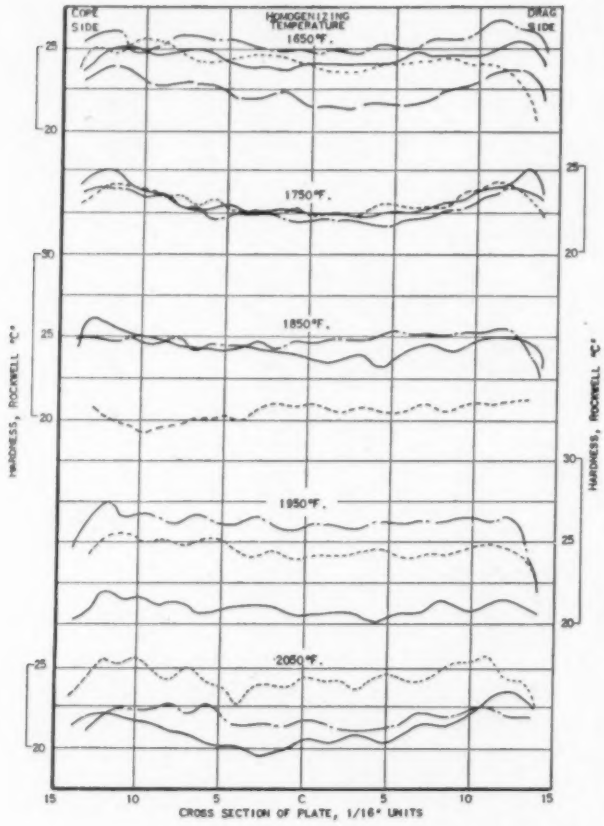
Fig. 6—Effect of homogenizing treatments on the hardness distribution within the cross section of quenched and drawn coupons from Plate No. 3.



Coupon No.	Homogenizing Treatment, Hr.	° F.	As-Quenched Bhn.	Drawing Treatment, Hr.	° F.	As-Drawn Bhn. Front	Back	Code
1650° F. Homogenize								
4-9	None		514	¾	1250	246	252	— —
4-21	2	1650	485*	1	1250	256	255	——
4-22	6	1650	471*	1	1250	269	255	— · —
4-19	12	1650	477*	¾	1250	256	248	- - - - -
1750° F. Homogenize								
4-18	2	1750	485	¾	1250	255	253	——
4-12	6	1750	471	½	1250	255	255	— · —
4-3	12	1750	526	¾	1250	255	246	- - - - -
1850° F. Homogenize								
4-24	2	1850	477*	1	1250	257	250	——
4-11	6	1850	492	¾	1250	241	255	— · —
4-4	12	1850	514	¾	1250	223	229	- - - - -
1950° F. Homogenize								
4-23	2	1950	485*	1	1250	255	248	——
4-10	6	1950	503	¾	1250	241	255	— · —
4-5	12	1950	454	½	1250	242	232	- - - - -
2050° F. Homogenize								
4-15	2	2050	551	¾	1250	240	241	——
4-20	6	2050	467*	¾	1250	246	248	— · —
4-6	12	2050	448	½	1250	241	237	- - - - -

*Hardness value of these replacement coupons were taken after a one-hour draw at 600° F.

Fig. 7—Effect of homogenizing treatments on the hardness distribution within the cross section of quenched and drawn coupons from Plate No. 4.



Coupon No.	Homogenizing Treatment, Hr.	Homogenizing Treatment, °F.	As-Quenched Bhn.	Drawing Treatment, Hr.	Drawing Treatment, °F.	As-Drawn Bhn. Cope	As-Drawn Bhn. Drag	Code
1650° F. Homogenize								
5-19	None		507	1⅛	1250	252	242	— —
5-13	2	1650	495	1	1250	255	255	——
5-14	6	1650	477	1	1250	262	255	— · —
5-2	12	1650	485	1	1250	247	237	· · · · ·
1750° F. Homogenize								
5-18	2	1750	488	¾	1250	241	230	——
5-12	6	1750	477	1	1250	244	235	— · —
5-3	12	1750	477	1	1250	246	242	· · · · ·
1850° F. Homogenize								
5-17	2	1850	481	½	1250	252	226	——
5-11	6	1850	488	1	1250	253	240	— · —
5-4	12	1850	451	1	1250	231	229	· · · · ·
1950° F. Homogenize								
5-16	2	1950	477	1	1250	241	246	— —
5-10	6	1950	467	1	1250	252	230	— · —
5-5	12	1950	415	¾	1250	244	246	· · · · ·
2050° F. Homogenize								
5-15	2	2050	474	1	1250	242	242	——
5-8	6	2050	467	1	1250	245	232	— · —
5-6	12	2050	444	¾	1250	231	242	· · · · ·

Fig. 8—Effect of homogenizing treatments on the hardness distribution within the cross section of quenched and drawn coupons from Plate No. 5

ate of 100 to 115° F. per hr. is probably a close approach to the cooling speed of full-size (2-in.) sections cooled in air under industrial conditions. Thus, it would seem that the data on the furnace-cooled specimens are more nearly representative of operating conditions than those from air-cooled coupons.

Since some coupons had already been air cooled from the tempering temperature before tests on the pilot coupon were completed, it was necessary to ascertain whether or not reheating of these air-cooled specimens to the tempering temperature followed by furnace cooling would in any way change the toughness values.

A glance at the two sets of curves in Fig. 9 discloses that reheating to the tempering temperature and then furnace cooling gives results that are in every way comparable to those obtained after furnace cooling directly from original draw.

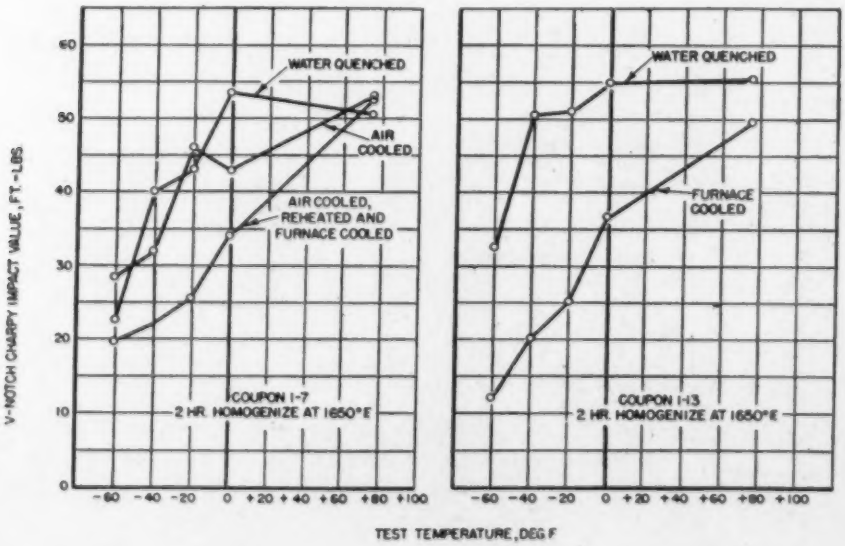


Fig. 9—Effect of cooling rate from the tempering temperature on notched-bar toughness.

The Charpy-type bar was chosen for the low-temperature notched-bar tests because it can be placed in the impact testing machine more rapidly than the Izod bar. A V-notch rather than a keyhole notch was used; however, because the V-notch, by giving a wider numerical spread in values, appears to be more sensitive to changes in the toughness of the steel, and thus would accentuate the location of a temperature-sensitive range if one were present.

Charpy results for the low-temperature tests are shown in Figs. 10 to 14, which are plotted to show the effect of the homogenizing temperature. Each point represents the average of two tests.

Considering the effects of both time and temperature on steels water quenched from the tempering temperature, comparison of the results presents no evidence that any one homogenizing heat treatment develops superior properties. Neither is there evidence of a trend in that direction. Furthermore, the position of the notch toughness-temperature curve for the unhomogenized speci-

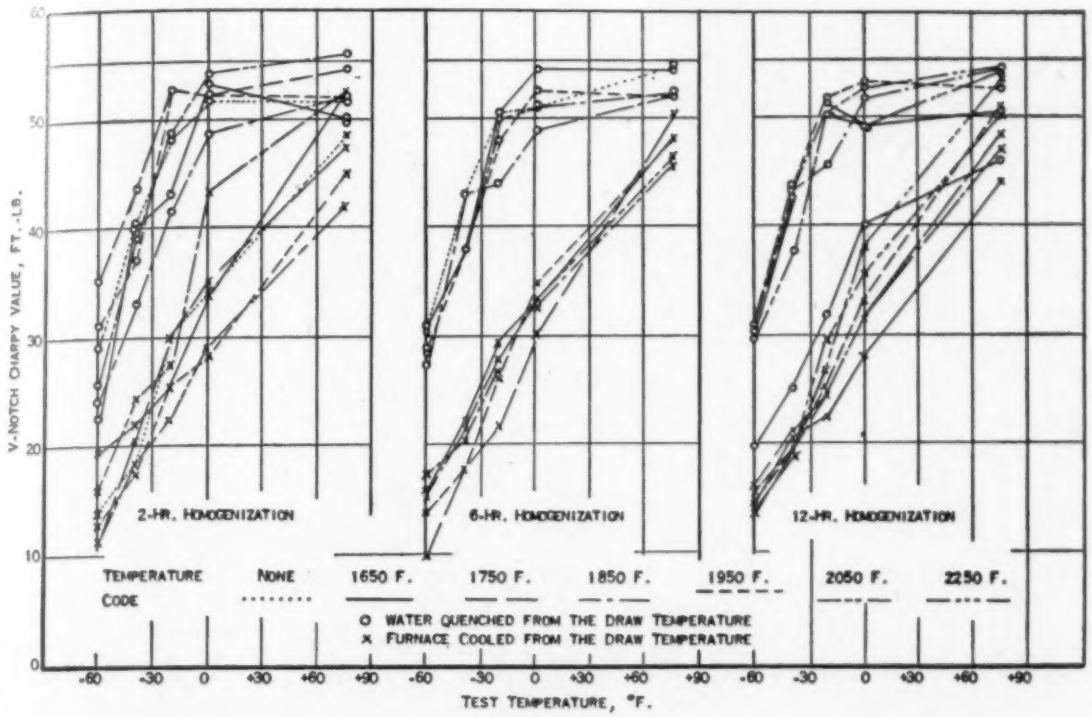
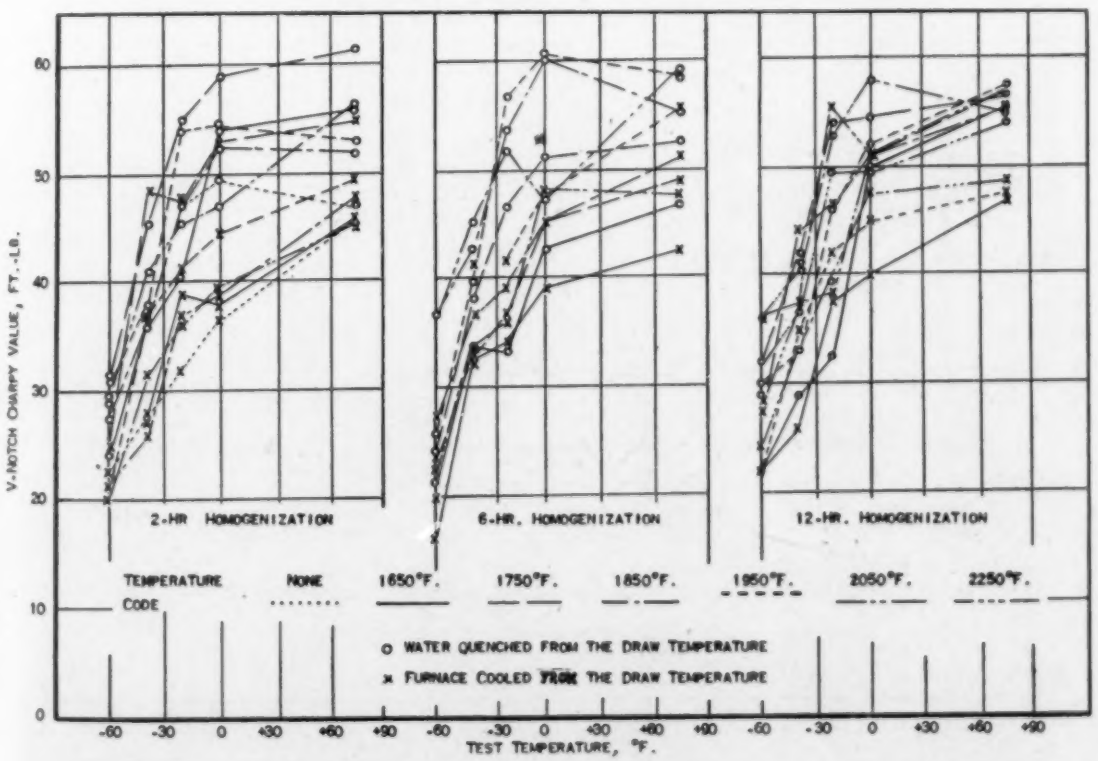


Fig. 10—Plate No. 1—Effect of homogenizing temperature on V-notch Charpy values.

Fig. 11—Plate No. 2—Effect of homogenizing temperature on V-notch Charpy values.



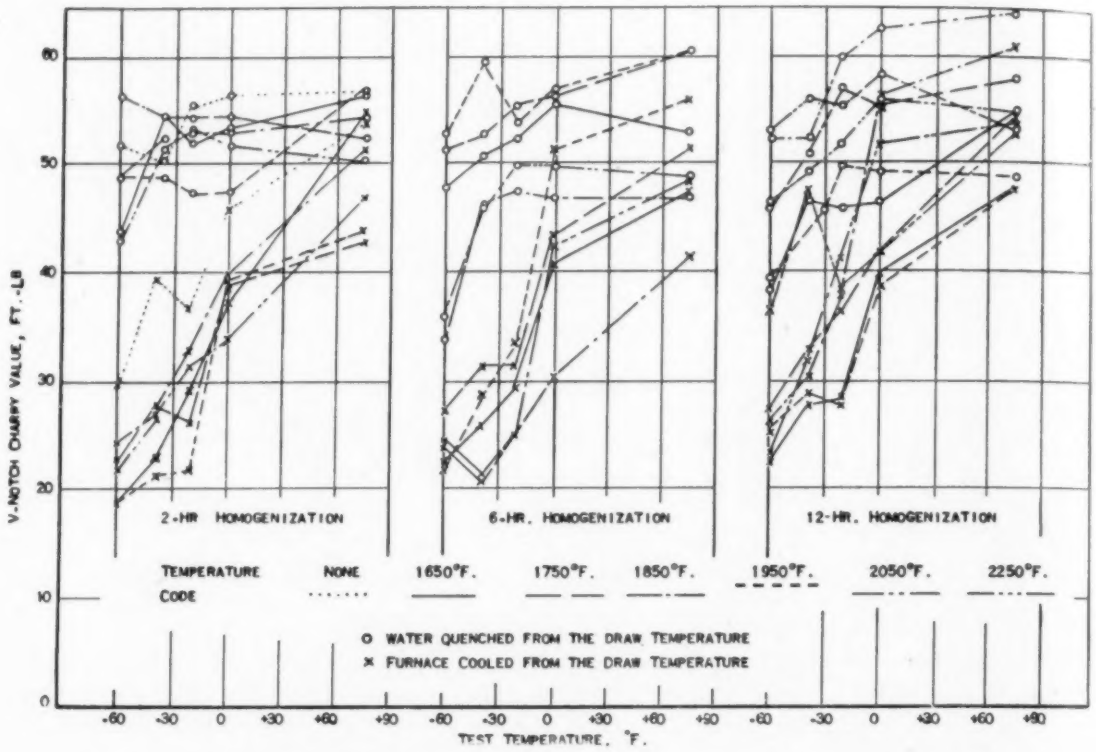
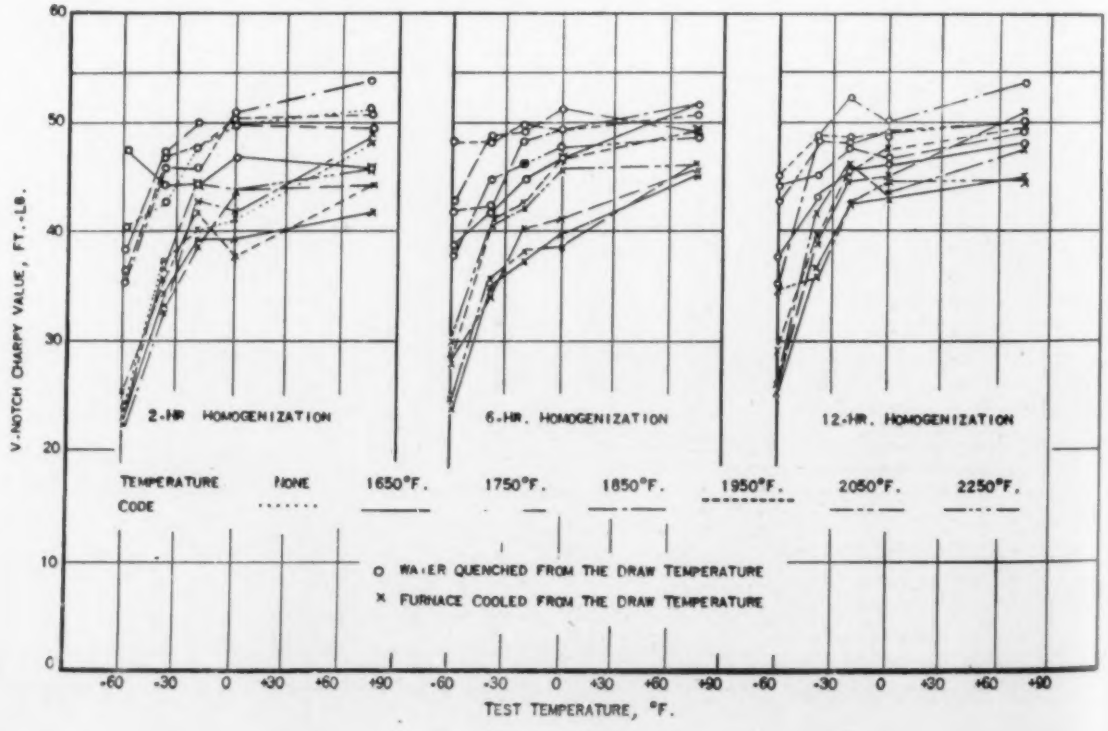


Fig. 12—Plate No. 3—Effect of homogenizing temperature on V-notch Charpy values.

Fig. 13—Plate No. 4—Effect of homogenizing temperature on V-notch Charpy values.



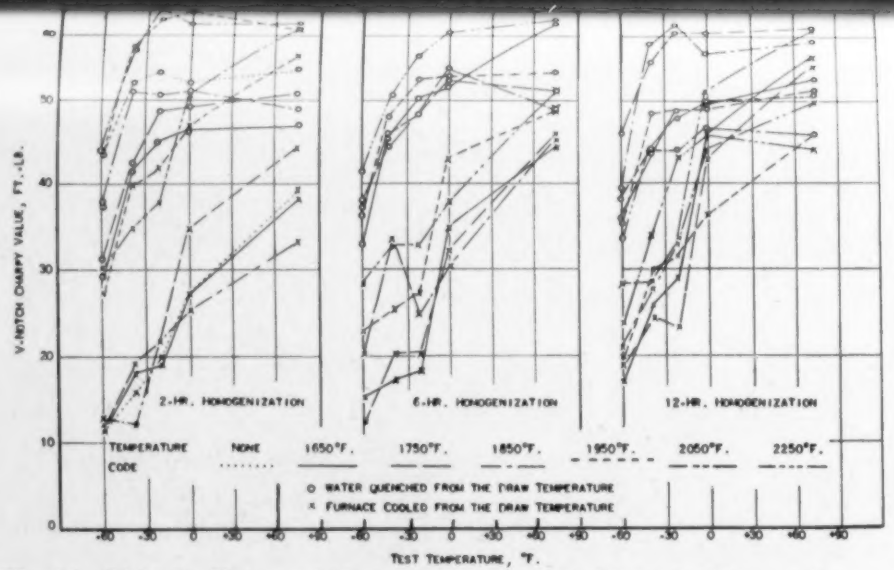


Fig. 14—Plate 5—Homogenizing temperature effect, V-notch Charpy values.

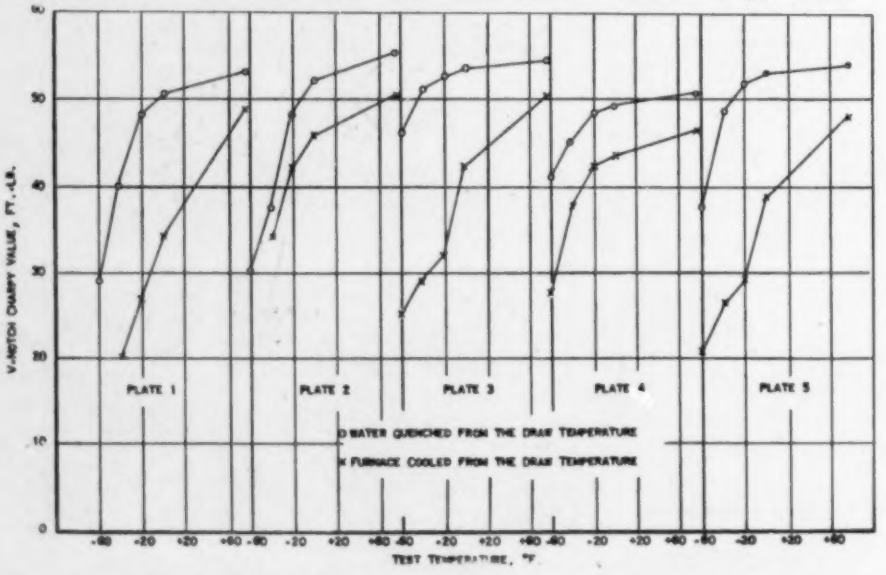


Fig. 16—Effect of testing temperature on the average Charpy values of the five steels. Each point represents the average of all specimens tested at that temperature regardless of prior heat treatment.

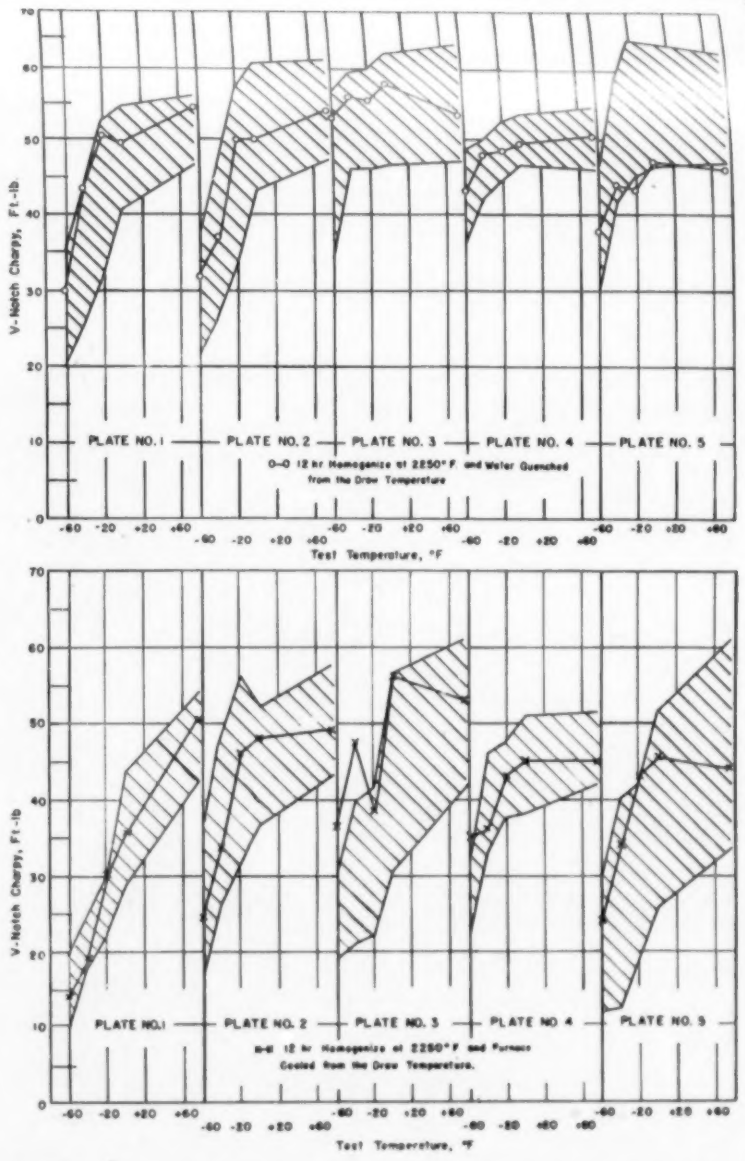


Fig. 15—Range of Charpy values after various homogenizing heat treatments and values after a 12-hr. homogenize at 2250° F.

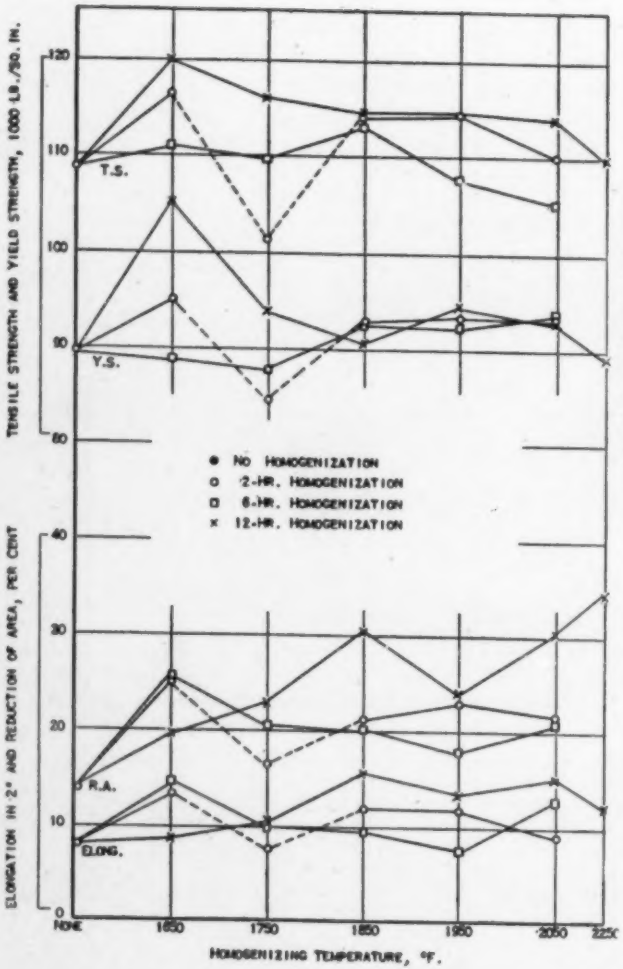


Fig. 17—Plate No. 1—Effect of homogenizing temperature and time on tensile properties. Dotted lines indicate uncertain values because of blow holes in the bar.

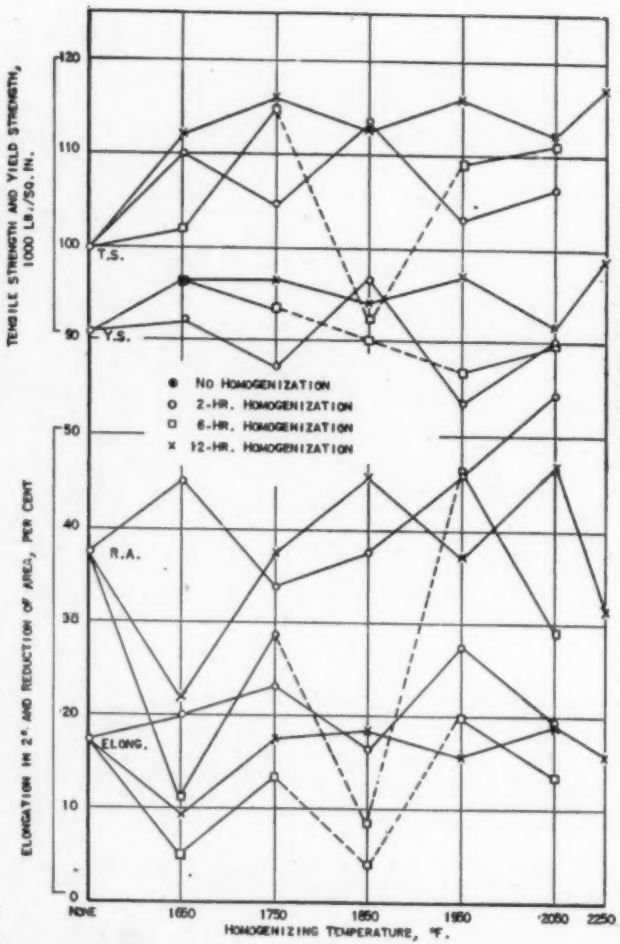


Fig. 18—Plate No. 2—Effect of homogenizing temperature and time on tensile properties. Dotted lines indicate uncertain values because of porosity in the bar.

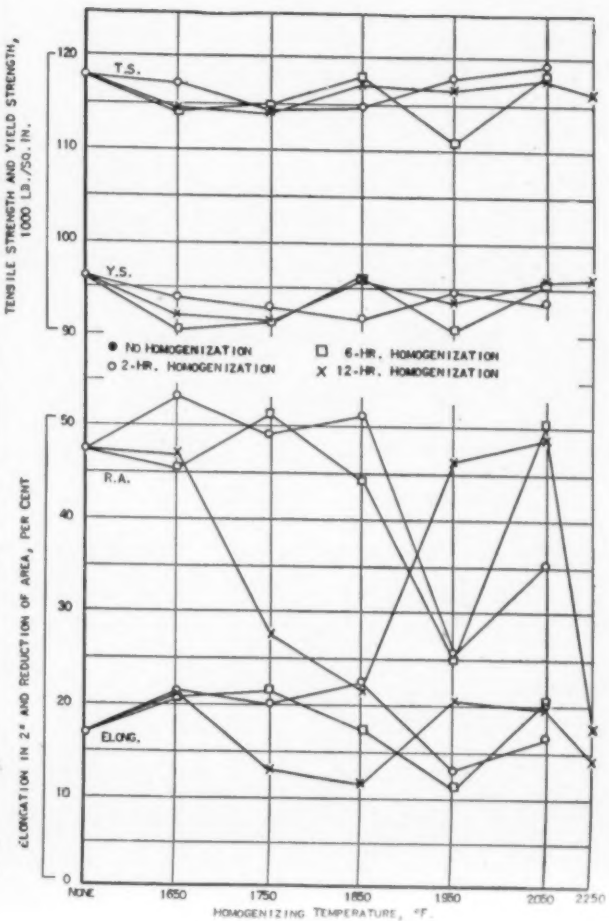


Fig. 19—Plate No. 3—Effect of homogenizing temperature and time on tensile properties.

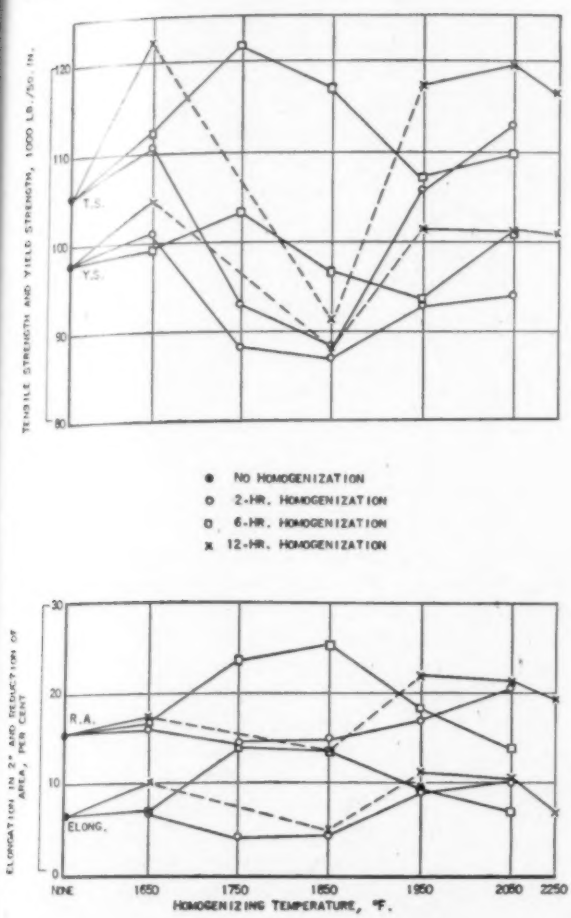


Fig. 20—Plate No. 4—Effect of homogenizing temperature and time on tensile properties. Dotted lines indicate uncertain values because of porosity in bar.

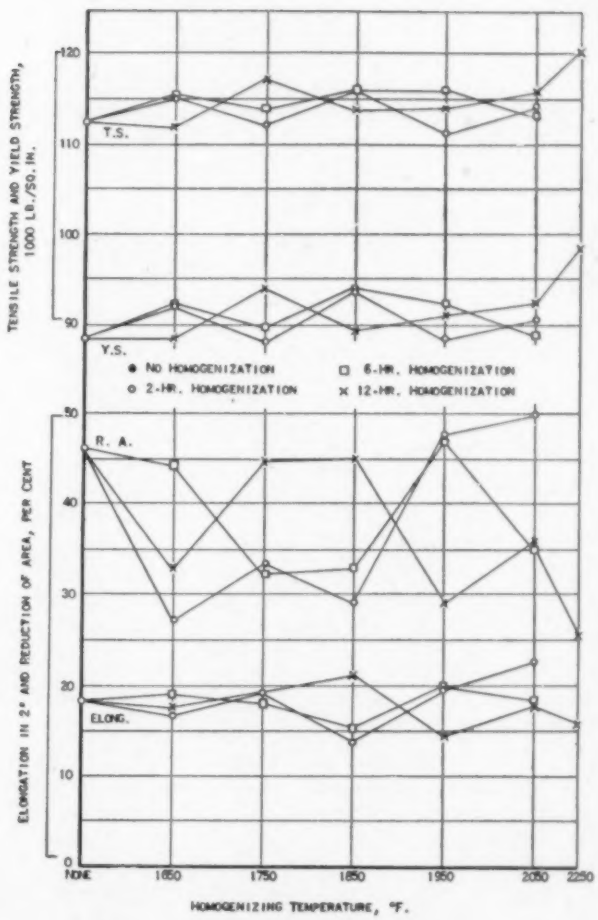


Fig. 21—Plate No. 5—Effect of homogenizing temperature and time on tensile properties.

mens in all cases falls within the range of the homogenized bars, suggesting that homogenizing has not resulted in any improvement in notch toughness.

In general, the same conclusions can be drawn with regard to steels furnace cooled after tempering. For Steels 2 and 4, the longer holding times appear to bring the room-temperature values closer to those for the water-quenched specimens, but the normal scatter in results may make this more a matter of chance than an actuality. It can be concluded that homogenizing heat treatments, in the case of the five steels examined, have not noticeably improved the notch toughness nor reduced the tendency toward temper brittleness.

It would be expected that if homogenization heat treatments affect the notched-bar properties, the most drastic treatment used, i.e., 12 hr. at 2250° F., would show a more obvious trend than the others. Thus, it might be expected that if the 2250° F. treatment improved the notched-bar properties, there would be some evidence of this effect in the data. Figure 15 was constructed to examine this point. The shaded areas in this figure represent the range of values obtained from coupons which had received the various homogenizing heat treatments excepting that at 2250° F.

The curves within these shaded areas give the test results on the bars homogenized 2250° F. for 12 hr. Since these curves generally fall within the shaded areas and are not restricted exclusively to either the top or bottom portions, it can be concluded that the homogenization at 2250° F. has not noticeably affected the notched-bar properties. This holds for both the water-quenched and furnace-cooled portions of the coupons.

Figure 16 represents the temper-brittleness susceptibility more clearly than do the preceding ten curves. In Fig. 16 the points obtained by averaging all of the values for any one testing temperature are plotted. Notable differences in temper-brittleness susceptibility and low-temperature toughness in the five steels are evident in Fig. 16. For instance, Steels 2 and 4 show the least tendency toward temper brittleness, and Steels 3, 4, and 5 have the highest Charpy values at the lower temperatures.

The ability of molybdenum to repress temper brittleness is at least partially indicated by the results of the Charpy tests. The two steels, 2 and 4, having the lowest susceptibility also have the highest molybdenum contents. The other three steels all contain molybdenum and show little sign of loss of toughness from temper brittleness at room temperature, but at lower temperatures the brittleness develops. This information suggests that molybdenum may act merely to shift temper brittleness to lower tempera-

Table 6

AVERAGES OF TRIPLE-NOTCH IZOD TESTS MADE AT ROOM TEMPERATURE

Time at Temperature, hr.	Izod Impact Values, ft.-lb. Plate No. 1	Plate No. 2	Plate No. 3	Plate No. 4	Plate No. 5
HOMOGENIZATION—NONE					
None	49½	57	58½	53	55½
HOMOGENIZATION AT 1650° F.					
2	51	58	58	51	49
6	52	53½	57	48	54
12	46½	54	50	49	52
HOMOGENIZATION AT 1750° F.					
2	52	57½	56½	51	51
6	55	54	56½	50	52½
12	53	55½	59½	—	56
HOMOGENIZATION AT 1850° F.					
2	48½	56	57	—	52
6	51	53	54	52	55
12	52	57	62½	54	60
HOMOGENIZATION AT 1950° F.					
2	49	62	57½	51	58
6	52	57	65	53	57
12	56	55	52	55	49½
HOMOGENIZATION AT 2050° F.					
2	53	56	56½	51	57½
6	53	57	51	53	57
12	51	53	59½	53	49
HOMOGENIZATION AT 2250° F.					
12	55	51	54½	49½	49½

tures rather than entire repression.

Figure 16 shows that the five steels have about the same room-temperature Charpy values. It will be remembered that Plates 1 and 2 were too low in hardenability to harden throughout the cross section. It would be expected that this lower hardenability would result in lowered notch toughness values, since best results with this property are obtained after a full quench and temper. Apparently, the specimens were taken from metal sufficiently close to the surface to avoid the semihardened areas of Plates 1-2.

Room-Temperature Izod Tests. The average of two triple-notch Izod bars for each steel and heat treatment are given in Table 6. Neither a variation in homogenizing time nor temperature has produced any significant change in room-temperature Izod values. The values closely obtained from the V-notch Charpy bars tested at room temperature.

Tensile Properties. The properties determined by the tensile test were quite frequently lowered by porosity or blowholes in the test specimens. This is apparent from the occasionally erratic curves in Fig. 17 to 21 wherein the tensile properties are plotted against homogenizing temperature. It should be recalled that the test coupons were drawn to 240 to 260 Brinell hardness, and presumably the tensile values should be quite similar in all cases.

In several of the figures, it appears that homogenizing has had some effect on the tensile properties, but it is believed that this relationship is merely a coincidence, since a thorough study revealed a closer connection between tensile properties and location of the specimens in the original plate than between tensile properties and homogenizing heat treatment. The variations in soundness throughout the castings, discussed later, certainly were large enough to account for some differences in properties.

The elongation and reduction of area data given in Figs. 17 to 21 bring out the fact that Plates 1 and 4 on the whole are less ductile than the other three plates, with Plate 4 the least ductile of the two. There are two likely causes for the deficiency in the case of Plate 4—its unsoundness, as brought out by deep-etch tests, and its higher carbon content (Table 1).

Less evidence of unsoundness in Plate 1 is shown, but closer examination revealed a finely distributed porosity or unsoundness that is likely the cause for low ductility in this plate, since its carbon content was not excessively high. It will be noted that the room-temperature notched-bar values for Plates 1 and 4 were also slightly, although insignificantly, lower than those of the other three plates.

The inclusion distribution is another factor which controls ductility in cast steels, but it is fairly evident from an examination of the inclusions that they are not involved in the low ductility of Plates 1 and 4. The results of the inclusion study are discussed more fully in a later section.

(Part 2 Begins on Next Page)

HOMOGENIZATION HEAT TREATMENTS FOR CAST STEEL

Part II

METALLOGRAPHY. The photographs in Fig. 22 show the as-received macrostructure of a full cross section, extending from edge to edge and passing through the center, of each of the five 36x36x2-in. plates investigated. Two such sections taken at right angles are shown for Plates 2, 3, and 5.

Photomacrographs of Plates 2, 3, and 5 show where the risers were located before their removal. The structure under the risers is mottled while that in other parts of the plate is dendritic. Plates 2 and 3 exhibit a "centerline weakness" in the areas not under the risers.

Photomicrographs in Fig. 23 show the type of inclusions and typical as-cast microstructures found in each of the five test plates. The nonmetallic inclusion rating for each plate, according to Sims and Dahle[3], is as follows:

Plate No.	*Inclusion Type*
1	III
2	III
3	I and II
4	III
5	I

The microstructures of the five plates are typical cast structures. A close examination of the micrographs representing the structures of Plates 3 and 4 will show a decided black line at the primary austenite grain boundaries. In some cases this type of grain boundary has been associated with steels containing boron. However, chemical analyses show that the boron content of these steels is negligible; therefore, these unusual grain boundaries must be the result of something other than boron.

Figures 24 and 25 show the macrostructure of each of the coupons from Plates 1 and 2 after they had received their particular homogenizing treatment. In both plates the dendritic structure becomes less sharp with successive increases in temperature and holding time. This effect was typical of that noted in all the plates.

Figures 26 and 27 indicate that homogenizing treatments have a tendency to make the microstructure more uniform in the as-homogenized condition. This effect was noticed in Plates 1, 2, 4, and 5; relatively little change occurred in the structure of Plate 3.

In general, the change in structure consists of changing the ferrite from equiaxed, polyhedral grains to acicular ferrite, and of making its distribution more uniform. In the case of Plate 4 the general structure is not changed appreciably; however, the areas of high alloy segregation, such as the one shown in Fig. 28, are somewhat equalized.

In this respect the improvement obtained from homogenizing at 2250° F. is illustrated by Fig. 29. These areas of high alloy segregation usually contained quenching cracks when the steel received a quenching treatment after little or no homogenization. However, prolonged homogenizing treatments seemed to break down the carbides and equalize the composition so that no cracks occurred on quenching.

Figure 30 shows the microstructure of the unhomogenized coupons of the five plates in the quenched and tempered condition, in contrast with the microstructure of coupons receiving the 12-hr. treatment at 2050° F. prior to quenching and tempering. A comparison of these photomicrographs shows that, in general, there is no improvement in microstructure.

It can be concluded that in some cases where high alloy contents are encountered, homogenizing as ordinarily practiced may improve the microstructure, but in relatively low alloy steels no benefits (to microstructure) are gained.

High Alloy Concentration

Examination of the specimens homogenized at 2250° F. for 12 hr. substantiated the conclusions reached from the other specimens; namely, that homogenizing treatments have no effect on the microstructures of quenched and tempered specimens, except in higher alloy steels containing areas of high alloy concentration. When such areas are encountered in the as-cast steel, they are effectively diffused by a prolonged high-temperature treatment such as the 12 hr. at 2250° F. used in the present case.

Quench Cracking. It was noted earlier in this paper that Steel 4 tended to crack during quenching unless special precautions were taken. The problem of quench cracking was made the subject of considerable study in another phase of the research work sponsored by the Office of Scientific Research and Development.

As a result of this work, valuable information regarding the relative importance of a number of variables influencing quench cracking was obtained. Although the subject is too voluminous to be discussed here, those results pertinent to the present investigation are worth noting.

As might be expected, carbon was found to be a potent element controlling quench cracking, the incidence of cracking increasing markedly with the carbon content. This would account for the difficulty ex-

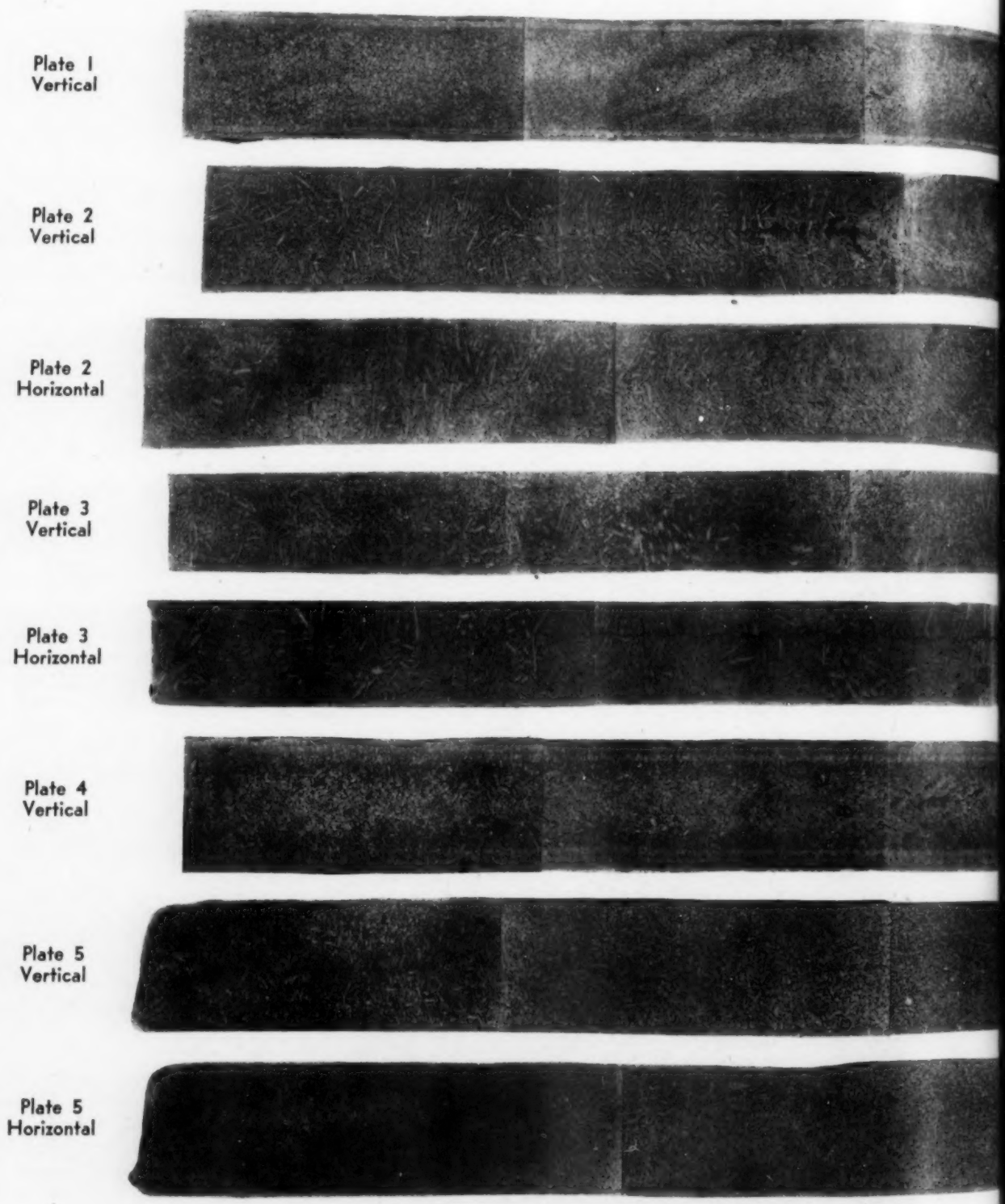


Fig. 22—Full section macro survey of each plate. See Fig. 1 for location of the sections in each coupon.

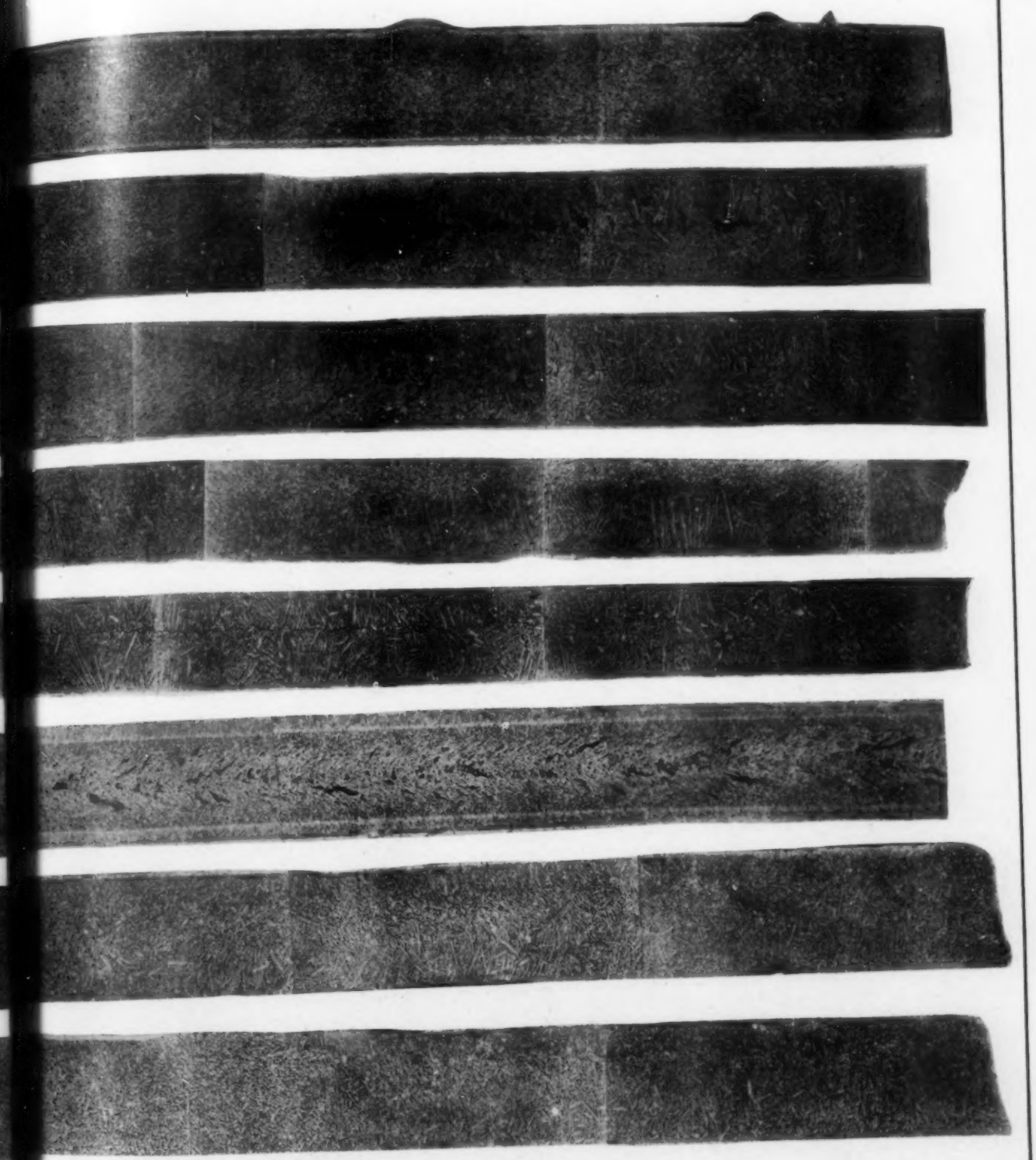
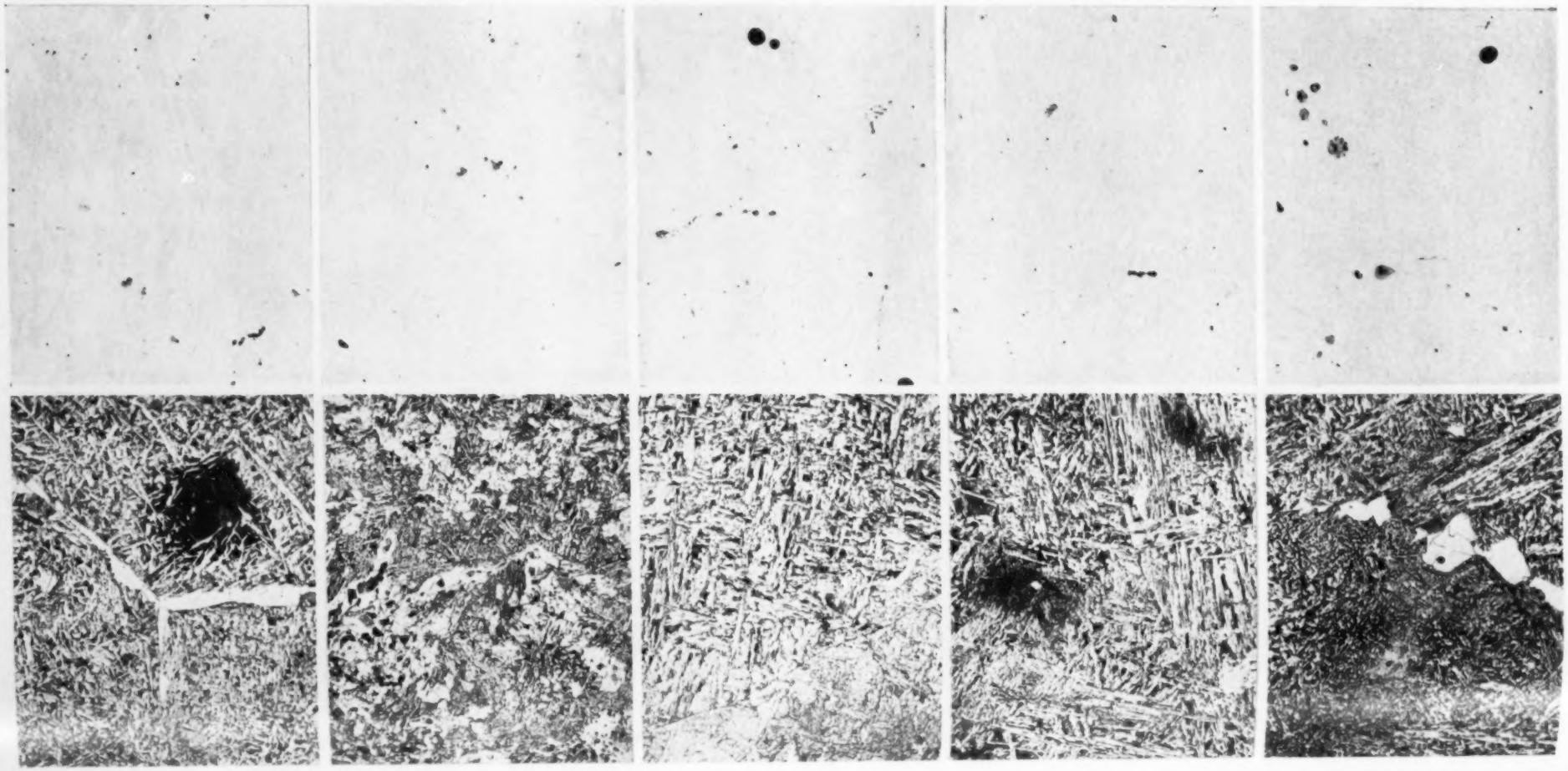


Fig. 23—Top row—Typical inclusions of each of the five cast plates. Unetched. X100. Bottom row—Typical microstructures of each of the five plates in the "as received" condition. Nital etch. X100.



perienced with Plate 4, which was higher in carbon than the rest of the steels. This work also showed that a homogenizing treatment at 2300° F. for 8 hr. was effective in reducing quench cracking.

Quenching Temperature

Furthermore, less cracking was experienced the lower the quenching temperature used during a hardening treatment. Although pretreatment in the form of a high-temperature homogenization was found to be beneficial, almost equivalent results were obtained by extending the holding time at the austenitizing temperature during the quenching treatment; consequently, the true worth of homogenization from the standpoint of reducing quench cracking is difficult to evaluate.

General Discussion. The general impression gained from a review of all the data is that the various homogenizing treatments were on a par in their effect on the properties of the five steels under consideration. Furthermore, there was no direct evidence that homogenizing benefited the mechanical properties or hardenability.

Most of the data which suggested that homogenizing was helpful were contained in the study of microstructure and quench cracking. It was shown by the microstructural study that in four out of the five steels, heterogeneity in the as-homogenized structure was broken up by using the higher temperature or longer holding times.

It was also apparent that in the case of Plate 4, which was of high hardenability, the more drastic homogenizing treatments broke up areas of alloy segregation and reduced the number of quench cracks. Despite this equalizing effect of the homogenizing heat treatment on the microstructure of the air-cooled specimens, this effect was not apparent in the quenched and tempered structures, and it was not possible to distinguish between unhomogenized or drastically homogenized quenched and tempered steel.

Segregation Type

Strictly speaking, heat treatments in the temperature range from the critical to about 2000° F. may not actually "homogenize" the structure. The segregation that occurs in cast steels is of the dendritic type, and the little evidence in the literature concerning this type of segregation leads to the conclusion that excessively high temperatures—in the neighborhood of 2300° F.—are necessary before an appreciable degree of homogeneity is attained.

Parke and Herzig[3], for instance, held a rolled steel at 2310° F. for 80 hr. to eliminate the bands originating from the original dendritic structure. Davenport[4] held various steels at 2200° F. to 2250° F. for 40 hr. to achieve homogeneity. A die block showing signs of segregation was given a fivefold increase in ductility by homogenizing for 10 hr. at 2200° F., or for 7 hr. at 2300° F.[5]. Segregation was still evident after these heat treatments. A treatment of 16 hr. at 2000° F. was completely ineffective in improving properties.

From the limited amount of information available, it appears that a homogenizing temperature of about 2200° F. is the minimum which can be expected to be effective in dissipating dendritic segregation, and that, even then, many hours at temperature are necessary.

Conclusions

In the five steels used for this investigation, it was found that:

1. Homogenizing treatments exerted no obvious influence on the austenitic (fracture) grain size. In general, a finer grain size was established in the coupons receiving no homogenization than in those which were homogenized.

2. Within the limits of error of the test, the hardenability of each steel remained the same regardless of the homogenizing time or temperature. The hardenability of the unhomogenized specimens generally tended toward the highest hardenability displayed by any of the specimens. These results were obtained on steels that varied in hardenability from air-hardening to fairly shallow-hardening properties.

3. The two steels with the lowest hardenability rating did not harden fully across the 2-in. section.

4. Considering the effects of both time and temperature during the homogenizing treatment, no difference in low-temperature notched-bar properties could be discerned. Unhomogenized steel was not inferior to homogenized steels in notch toughness.

5. Temper brittleness exists to a characteristic degree in each of the steels and is not affected by homogenization.

6. The data suggests that the function of molybdenum may be to move the range of temper brittleness to lower temperatures rather than to repress the brittleness entirely.

7. The notched-bar values for all of the steels were about the same at room temperature.

8. Homogenizing had no effect on tensile properties. The ductility of two of the steels was inferior, but this appeared to be caused by unsoundness in the casting rather than by any other factor.

9. No effect of homogenizing was observed in the microstructure of quenched and tempered specimens; however, the more drastic homogenizing treatments reduced the heterogeneity in the as-homogenized microstructures of four out of five steels.

Dendritic Pattern

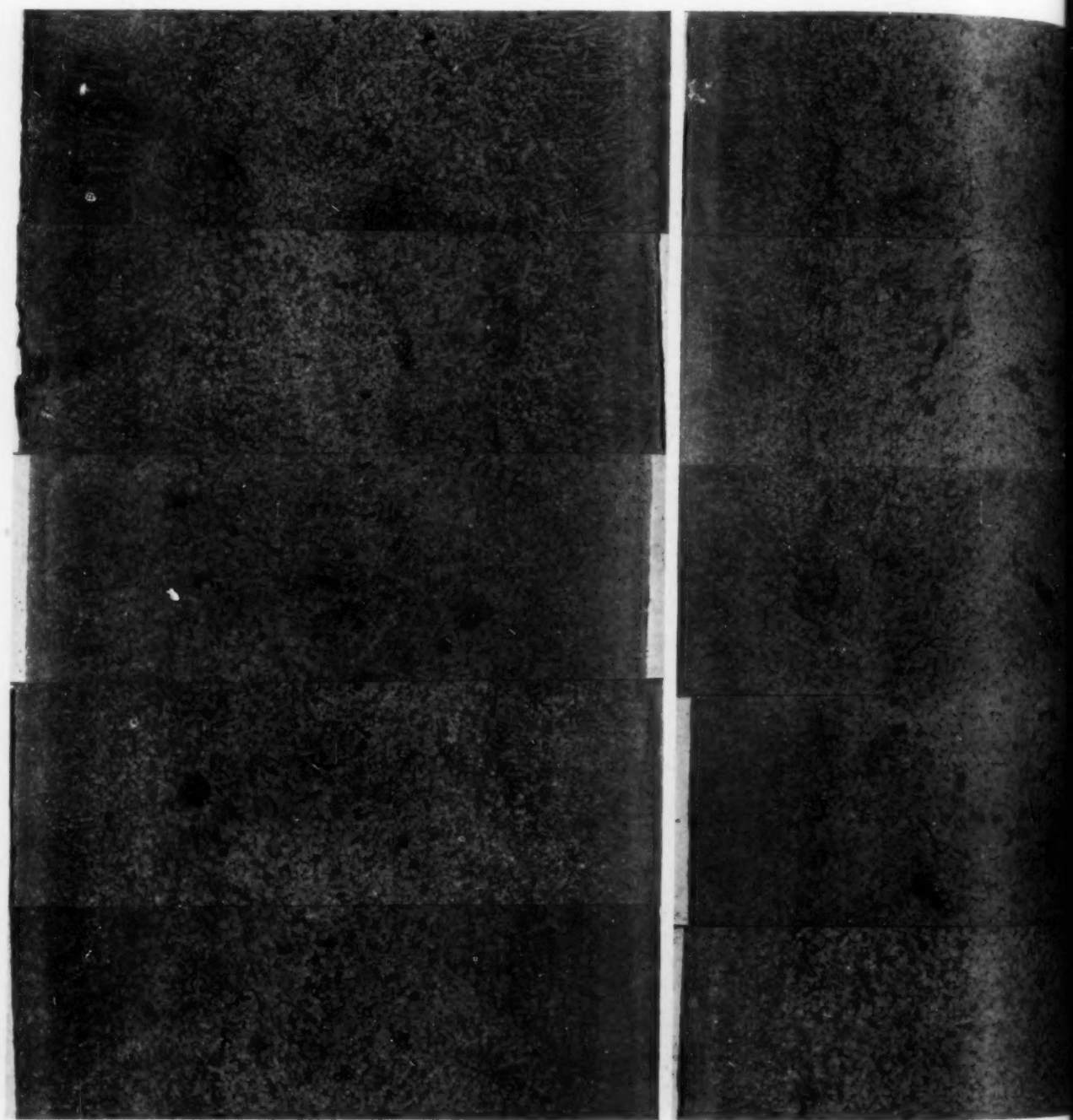
These treatments also reduced the sharpness of the dendritic pattern on macro-specimens. All but the last traces of the dendritic structure were removed by the 2250° F. treatment, and segregated areas in the higher alloy steels were successfully diffused. However, no change was apparent in the quenched and tempered microstructure.

10. A tendency toward quench cracks in the air-hardening steel was reduced by the more drastic homogenizing treatments. This improvement was believed to be the result of partial diffusion of areas of high alloy concentration. Other work has indicated that quench cracking may be reduced by a high-temperature (2300° F.) homogenizing heat treatment.

11. The benefits to properties to be derived from a homogenizing heat treatment even as high as 2250° F. are completely overshadowed by the effect of the steel quality (porosity, nonmetallic inclusions, etc.) on properties.

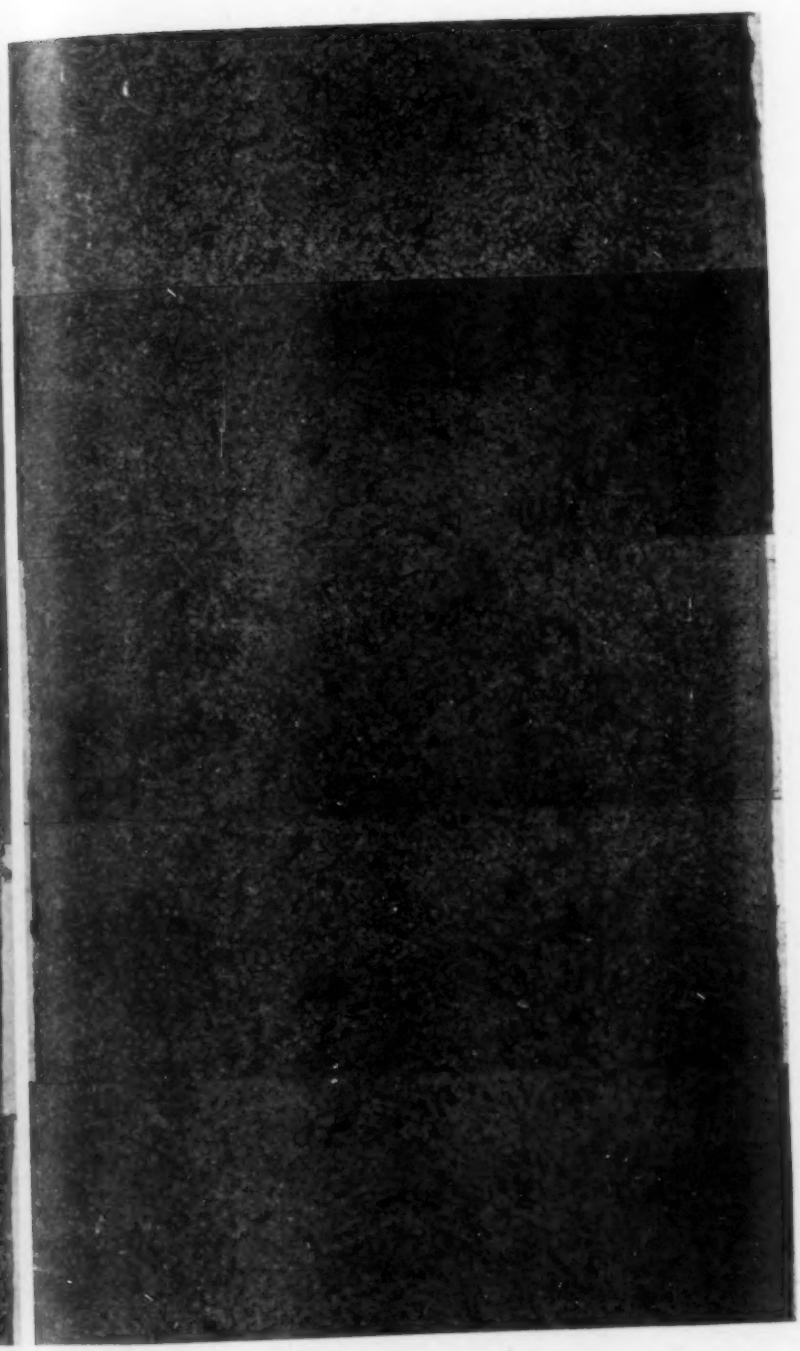
12. Inclusion distribution was not a factor in affecting mechanical properties of these five steels.

13. True diffusion of the dendritic segregation in cast steels, apparently, is realized only by heating for an extended time at temperatures higher than those currently used.



2 Hours 6 Hours

Fig. 24—Macrostructures of cross-section specimens from Plate 1 that have received the various homogenizing heat treatments. Temperatures are given at right and times below each column. Note the gradual decrease in sharpness of the dentrites with increase in time and temperature. All specimens photographed at 2 diameters.



1650° F.

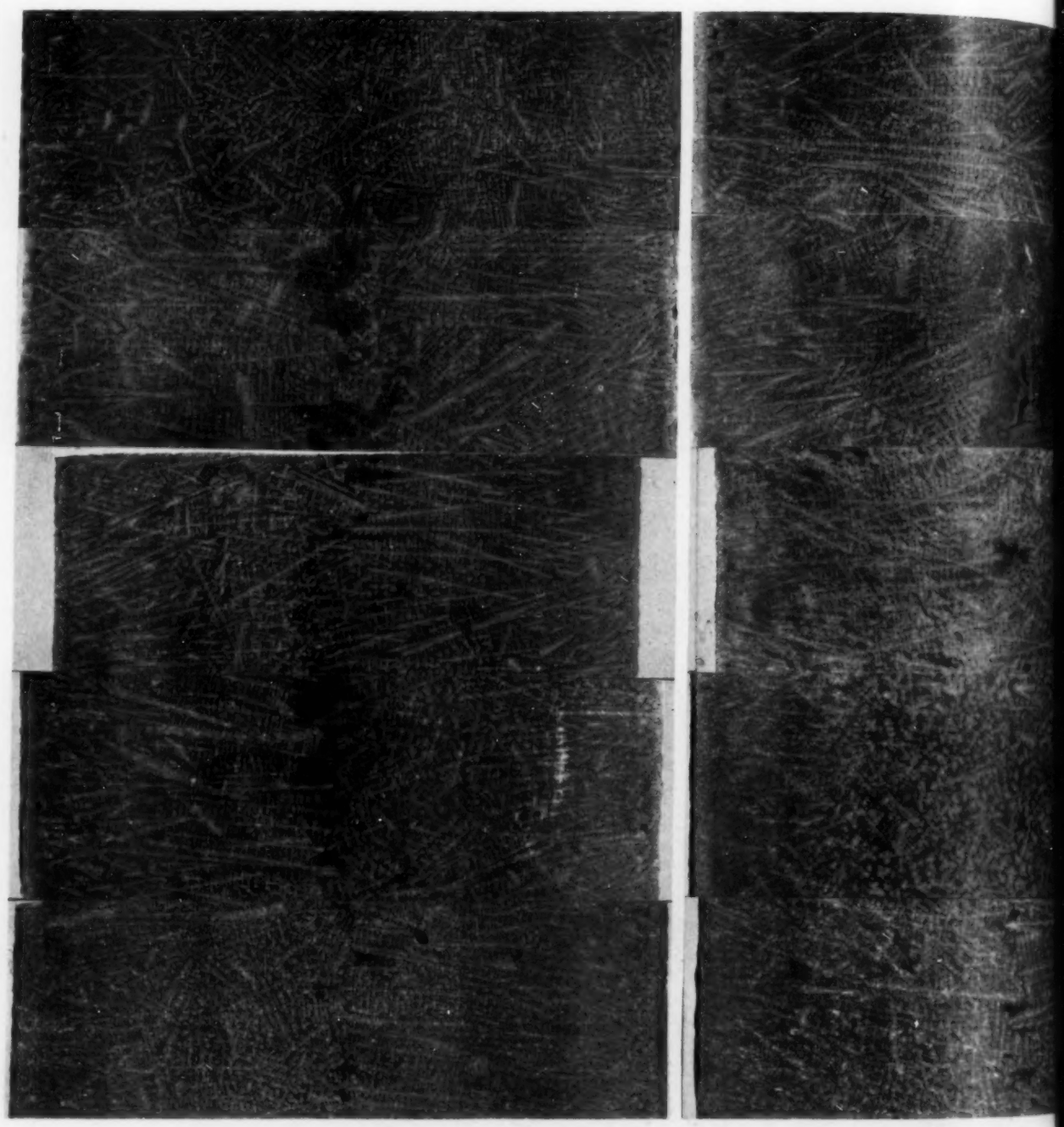
1750° F.

1850° F.

1950° F.

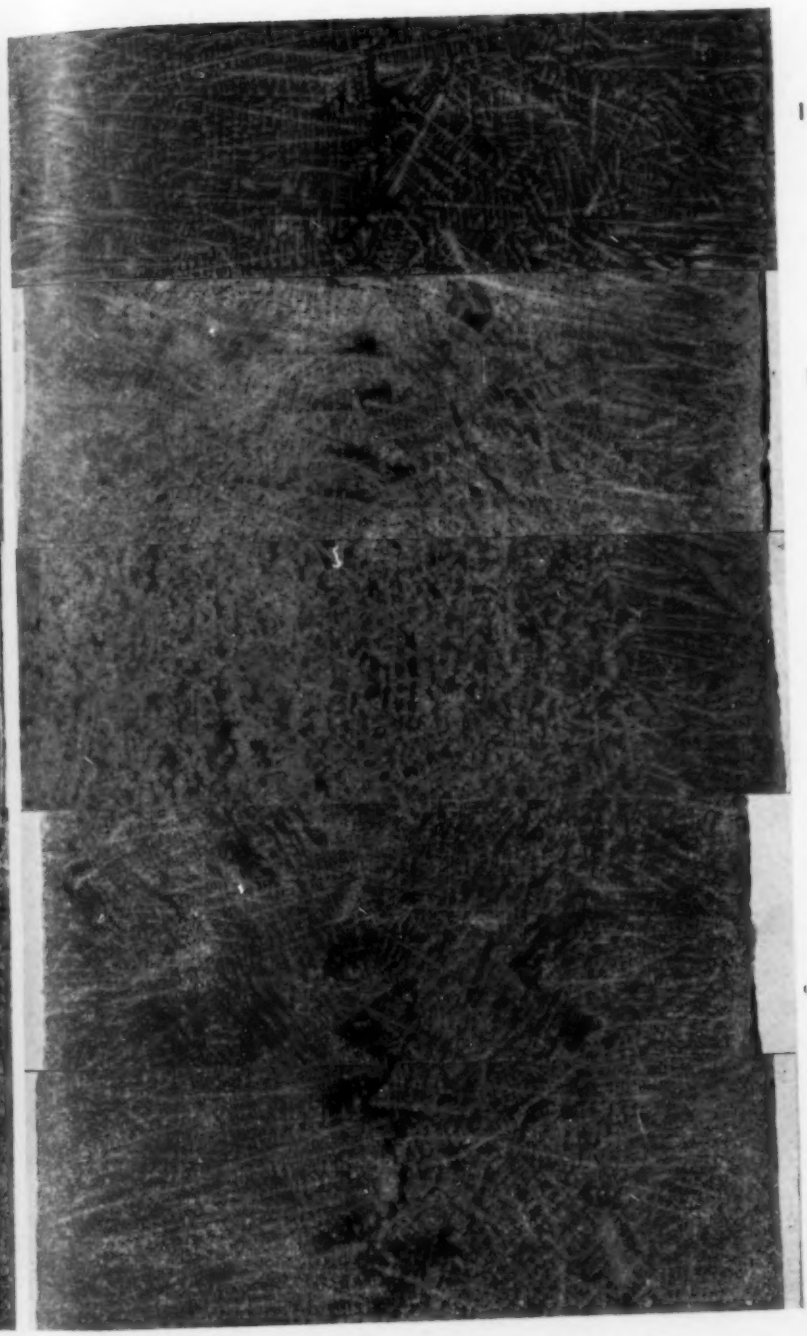
2050° F.

12 Hours



2 Hours 6 Hours

Fig. 25—Macrostructures of cross-section specimens from Plate 2 that have received the various homogenizing heat treatments. Temperatures are given at right and times below each column. Note the gradual decrease in sharpness of the dendritic structure with increase in time and temperature. All specimens at 2 diameters.



12 Hours

Fig. 26—Photomicrographs show the microstructure of Plate 1 after receiving homogenizing treatments indicated by temperatures given at top of columns and times at right. All specimens nital etched. X100.

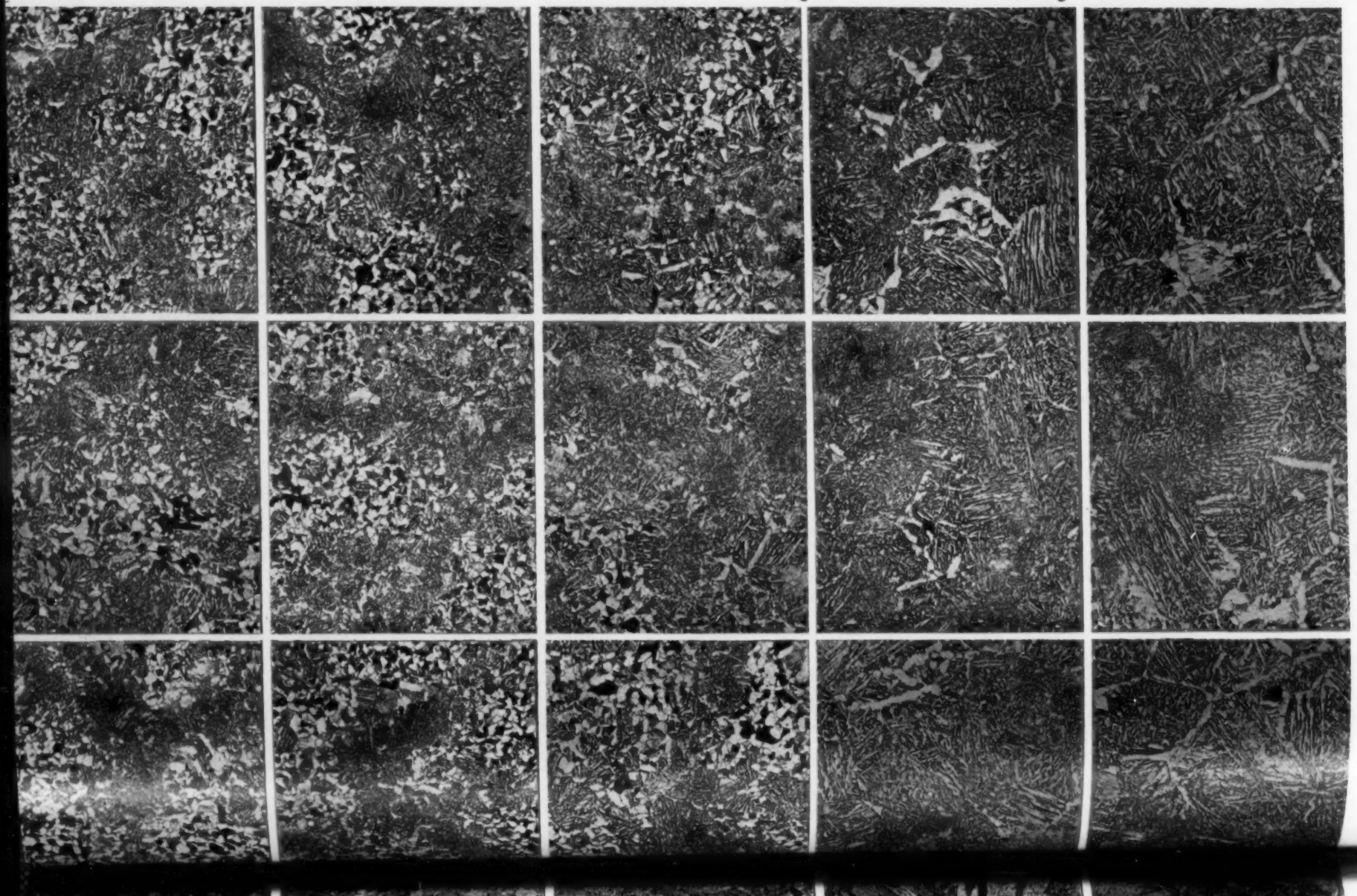




Fig. 27—Photomicrographs show the microstructure of Plate 4 after receiving homogenizing treatments indicated by the temperatures given below columns and times at right. All specimens nital etched. X100.

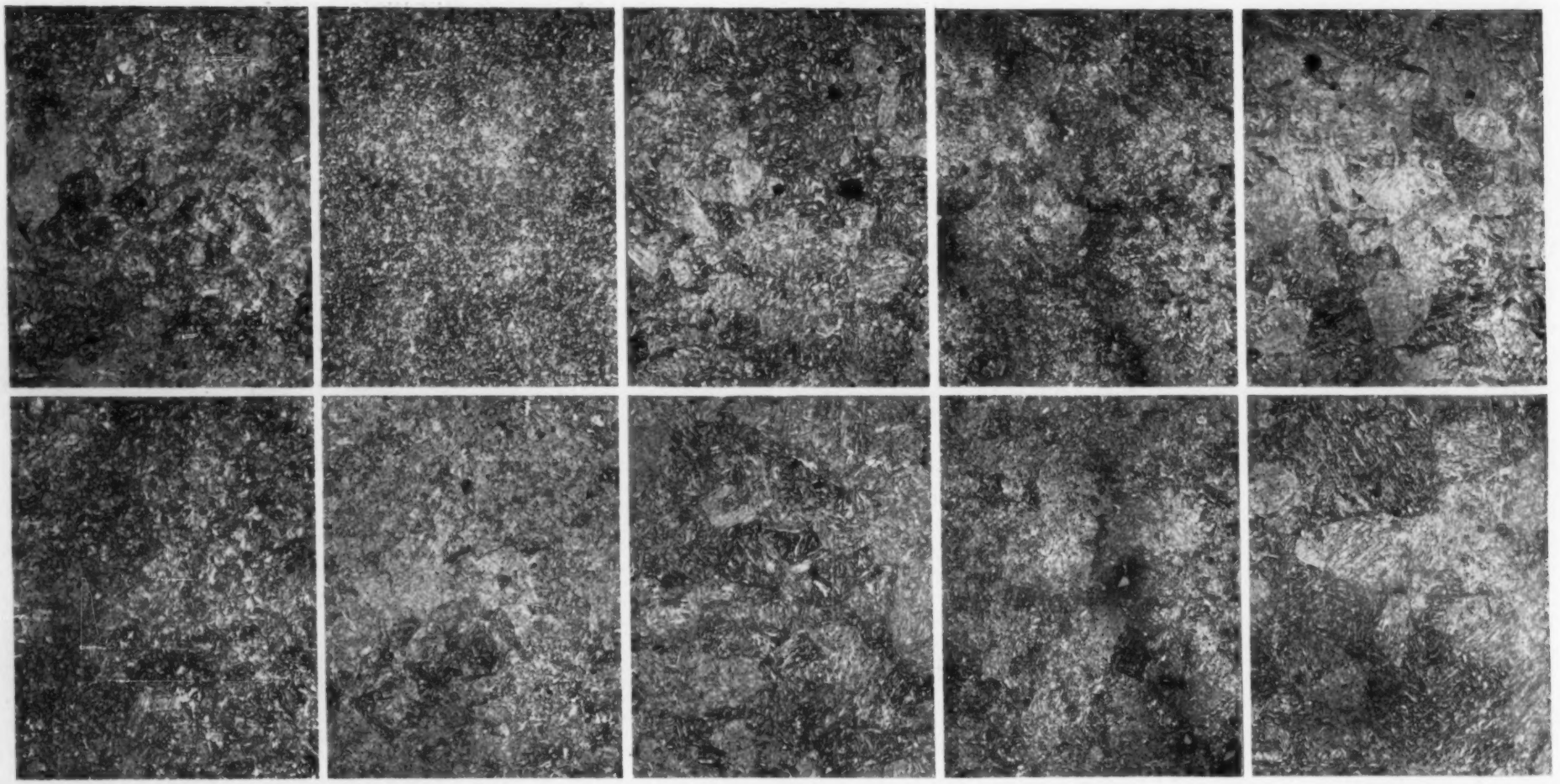


Fig. 30—Top row—Microstructure of the quenched and drawn cast plate specimens which had received no previous homogenizing treatment. Bottom row—Microstructures of the same steels after homogenizing for 12 hr. at 2050° F. followed by quenching and drawing treatments. All specimens nital etched. X100.



Fig. 28 (left)—Area near center of Plate 4 after quenching and tempering, showing alloy segregation. Note massive carbides and quenching cracks. No homogenization heat treatment. Nital etch. X100.

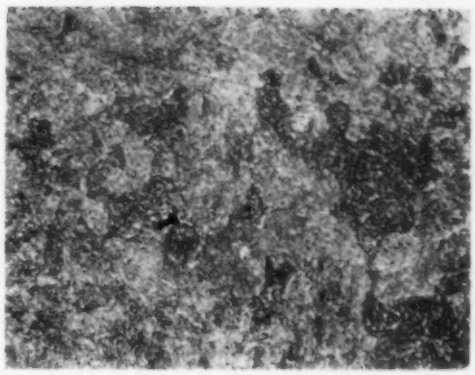


Fig. 29 (right)—Quenched and tempered structure of Plate 4 after a homogenization of 12 hr. at 2250° F. Segregation is no longer visible. Nital etch. X100.

Acknowledgment

This investigation constituted one phase of a research program sponsored by the National Defense Research Committee of the Office of Scientific Research and Development, under the classification—N.D.R.C. Research Project NRC-14. Grateful acknowledgement is made to the Office of Scientific Research and Development for supporting this program and for their permission to publish this information, and to the War Metallurgy Division of N.D.R.C. and the War Metallurgy Committee for supervising and directing the research. Thanks are also due to many members of the Battelle staff for their help, especially to Dr. C. H. Lorig for his guidance and direction, and to A. R. Elsea for his metallographic studies of the steels.

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DISCUSSION

Chairman: H. H. BLOSJO, Minneapolis Electric Steel Castings Co., Minneapolis.

Co-Chairman: CHARLES LOCKE, West Michigan Steel Foundry Co., Muskegon, Mich.

R. E. KERR[1]: During the war, we were producing breech rings in an Ohio steel foundry. One type of breech ring used on bombers called for an extremely thin section with high tensile properties. We used manganese-molybdenum steel and tried homogenizing some of them. We put them in bell-type annealers with large steel rolls at 2000° to 2300° F. for long periods of time, some of them for 20 hr. or more. We cut these rings, machined tensile bars from standard locations, and found little difference in yield strength or tensile strength, but noted a definite improvement in elongation and reduction of area.

C. E. SIMS[2]: We did considerable amount of work on homogenization treatments of steel in an entirely different field of endeavor, but the results coincide very well with those reported. When talking about homogenization, we should define what we mean. One can homogenize the carbon very readily, but when it is segregated it is a symptom of alloy segregation.

The best way to determine whether a steel is homogenized or not is to heat it above the critical temperature and cool it slowly, that is, give it a full anneal and see if the carbon distributes evenly or whether it is segregated. If the alloy content of the steel is segregated, the carbon will segregate also, but if the alloy content is evenly distributed, there will be uniform distribution of the carbon. Therefore, when talking about homogenization we are really talking about distributing the alloy which has been segregated because of dendritic segregation during freezing.

It appears that to get true homogenization in practice one must apparently go to an impractical temperature of at least 2200° F. If one uses what might be called a practical temperature below 2000° F., homogenizing requires an impractically long time. It would require probably a week in some cases. But we were able to get complete homogenization with temperatures of 2250° and 2350° F.

The results on Charpy values and tensile tests were similar to those reported. We sometimes lost a little yield strength and in some steels we gained yield strength, but not to a marked degree. We usually gained something in ductility. There was a small shift in the transition temperature of Charpy bars broken at low temperatures. It appears that any increases obtained in the ordinary mechanical properties of the steel are hardly worth the treatment.

C. R. JELM[3]: Were these tests made with normalized steels?

MR. SIMS: The tests were made with normalized steels. The benefits are of the same order and magnitude.

J. A. DUMA[4]: Have any tests been made on double homogenizing heat treatments at lower temperature to obtain information on the degrees of effectiveness of homogenization obtained? The authors state that a temperature of 2250° F. is required for optimum homogenization, but on the basis of the microstructures that we have examined, it appears that two low temperature homogenizing heat treatments produce almost the same results as a single high temperature homogenization treatment.

It is a well known fact in spheroidizing heat treatment practice that the action of repeated passage of the steel up and down through the critical temperature points shortens significantly spheroidization of the iron carbide in pearlitic carbon and alloy steels. It is probable that this action also manifests itself in double homogenizing heat treatments.

MR. KURA: I want to thank Messrs. Kerr, Sims and Duma for their comments. We referred to Mr. Sims' work several times in the paper.

Reference was made to tensile properties of breech rings. You will recall that we discounted our tensile data mainly because of porosity. A plate is one of the most difficult shapes to cast and make sound. Unfortunately, we cannot make any comparisons of tensile properties as a result of homogenization.

I know of no references to work done on the effect of double homogenizing. I see no difference between homogenizing two or three times at a given temperature, regardless of what that temperature may be, as compared to homogenizing for the accumulated time at the same temperature. I perceive it as somewhat akin to tempering. So far as the end results are concerned, there is no difference between double and single tempering of steels of low and moderate hardenability if the total time at temperature is the same in each case.

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[2]Battelle Memorial Institute, Columbus, Ohio.
[3]Thompson Products Co., Cleveland.
[4]Norfolk Navy Yard, Portsmouth, Va.

ACCELERATED TRAINING

American industry can not afford to lose the training gains which were made during the war period. In postwar operations, speedy and adequate training of workers will be an important factor in balancing the plant budget.

W. J. Hebard
Personnel Director
Continental Foundry & Machine Co.
East Chicago, Ind.

MANY CHANGES have been brought about in the foundry industry because of the termination of war production. In many respects plants are coping with different problems—new headaches to replace the headaches of wartime—but with respect to the subject of this paper, training, the problems are fundamentally the same. Only the externals are changed. Today and every day the plants are confronted with training problems.

It is true that apprenticeship in the foundry suffered a real setback during the war and is now receiving the greater emphasis that it deserves. It is true, likewise, that job-training has gone in the reverse direction.

Reduction of Training

During the war it was necessary to develop rather elaborate programs, with training staffs planning for large groups of trainees and special job instructors in every department. The end of the war, bringing a decreased turnover and, in many plants, decreased production, has brought about a reduction in the size of the training job—but the essence of the job is still there and it is important.

It is at this stage of the game that industry should be most concerned. During the war, costs were of secondary importance. Production in a hurry was paramount. Today, industry is extremely cost conscious—and it needs to be. Terminations, cancellations, price ceilings and wage increases have brought it up short against *cost*. Elimination of training programs can not be allowed merely because their cost can be measured and the cost of lack of training programs can not be measured. It can not be thought for a minute that going back to familiar prewar products means no further need for training.

No doubt, many will wonder why it is thought necessary to mention such a thing because the value of adequate training has been learned and is not likely to be forgotten in a hurry. Recalling a bit of history at this point may serve to explain the concern felt.

During World War I, American industry faced the same problems as it did in World War II. The "Preparedness" drive of 1916 paralleled the "Defense" drive of 1940. The "War Production" of 1917-1918—relatively at least—matched the "War Production Drive" of 1942-1945; and encountered the same problems of training quickly the thousands of green workers brought into the expanded plants.

Much was learned about the value of good training. An organization sprang up to afford an exchange of ideas among industrial training people. To sum up the knowledge of training acquired during the war, Charles R. Allen wrote his great book, *The Instructor, the Man, and the Job,* as a sort of Bible for industrial training.

When 1940 came a few corporations, mostly large ones, had fairly adequate training staffs, but in the main, American industries, and particularly the foundry industry, were not even training conscious. Fortunately, someone remembered Allen's book, got out his four-step method of job training, dressed it up in modern clothing, and brought it out as the now famous "J.I.T."—Job Instruction Training.

Government agencies, trade associations, and individual top managements, exerted strenuous efforts, first, to make industries aware of the need for adequate training, and second, to help them learn how the job should be done. The fact that the industries of America won out in the war of production was as much a tribute to the training efforts in industry as it was to the development of new techniques and new products, but it is submitted that the race might have been shortened, the cost might have been lessened, if industry had, through the years since 1919, profited by its experiences at that time.

"J.I.T." Valuable

It is to be hoped that training tools like "J.I.T." will not be permitted to gather dust for the next 25 years, and that industry not allow itself to slip back into the slow, inefficient, and costly method of hiring people and "letting them learn" their jobs. The fact that the actual cost of improper training, or no training at all, is not set apart on the books, but is hidden in the overall figure of tonnage per man hours, should not be overlooked.

For the most part the actual training job must be done by the line

Presented at an Apprentice Training Session of the Fiftieth Annual Meeting, American Foundrymen's Association at Cleveland, May 9, 1946.

supervisor. It is not often that the conditions of a particular training project in normal peacetime activity will justify the use of special instructors in the average foundry. However, there must be someone, somewhere in the organization, whose responsibility it is to anticipate the training needs, plan the training, and assist the line supervisor in doing an adequate job of training.

Production Planning

It is unfair to lay the whole job in the lap of the foreman. How elaborate, or how simple the set-up necessary for a particular organization, will depend upon the size and the type of the organization.

The responsibility for the planning of training springs from the planning of production. To illustrate the development of a specific training program which is designed to meet the needs of production planning, an example may be made of a large organization where the procedure is somewhat as follows:

When the manufacturing department plans to establish a new production line, the training department is called in, given all the detailed information as to the specific jobs which will be required, the number of people on each job, and the starting schedule.

From this information the training department makes an analysis of the training needs, taking into account for each operation how many people are needed, how many are already available, trained or partly trained in the organization, and how many must be fully trained. Each job is then studied to determine how much training is necessary and how it is to be given.

While the size of this organization justifies an elaborate training division, the same can not be said of the average foundry. However, the same fundamental problem is present, just as it is in the training of the casual replacement in an established production line.

The same function may be performed as a part-time duty of one man; it may be the full-time duty of one man, assisted by a committee of operating supervisors, or it may require a regular staff, depending upon the size of the organization, the rate of turnover, and the frequency of change in product.

It would be a waste of time to go into detail in recounting the value and importance of properly planned training for speeding up the process of bringing employees on a new job to full production. From the author's plant might be cited the case of training crane operators in 5 days, as compared with the "let them learn" process, which previously took from 30 to 60 days.

Experience in training women as coremakers and mold finishers in a two-week program, with what can be considered remarkably fine results, may be mentioned. Most foundries no doubt could point to similar experiences. However, the significant thing for consideration should be the method used to arrive at the objective.

For example: To train the crane operators an experienced craneman, who seemed to possess special aptitude as an instructor, was selected. He was given help to prepare the description and explanation which could be given as related instruction in a classroom; then arrangements were made for the use of a particular crane and he was helped to organize the steps by which the operation of the crane was to be taught.

In the case of the women who became coremakers and mold finishers, a special program of related instruction and actual experience was lined up with the aid of the local vocational school, and the trainees were given their two weeks of training in a classroom in the high school.

In short, in each case an attempt was made to arrive first at the answers to the question "What is to be Taught," "Where can it be Taught," "How can it be Taught," and "Who Should do the Teaching"? It is certain that these questions must be answered in every case and that there must be someone whose responsibility it is to work out the answers.

Interest in Training

Today, in thinking of training crane operators, perhaps only one or two being necessary in a month, the "Where" and "How" questions would find an entirely different answer. First of all, everyone concerned with training, from topmost management to the newest assistant foreman, must be convinced that planned training is an integral part of the line organization, in war or peace.

Someone must have the responsibility for directing and organizing specific training programs to meet specific needs. That someone must be informed fully of production plans and schedules which involve new employees for old jobs, or old employees for new jobs. Most important of all, he and the entire organization must be made to understand that he has the full and active support of top management. After that, it is up to him.

It is his job to work out reasonable and practicable plans for assisting the shop supervisors in their training problems. He has to get the answers to the "What and Where and How and When and Who." Perhaps the answer is right in the department, on the job with the foreman doing the work. It may be in a classroom or in a corner of the shop with an expert called in. Whatever the answers, the plan should be thoroughly made, properly executed, and adequately followed up. The results will show in lower costs, lessened turnover, quicker deliveries and better labor relations.

DISCUSSION

Chairman: F. C. CECH, Cleveland Trade School, Cleveland.

Co-Chairman: W. J. HEBARD, Continental Foundry & Machine Co., East Chicago, Ind.

C. J. FREUND[1]: *(Summary).* The statement which made the deepest impression on me was the need for the manufacturers to outline in advance the aims and purposes of their training program to insure success of the program.

I believe the foundry industry made a serious mistake in letting too much important foundry information accumulate in the minds of operating foundrymen and not making sufficient effort to get it in print for the instruction of the next generation of foundrymen.

The emphasis Mr. Hebard made on C. R. Allen's book shows how valuable a record, be it a book, file, letter, drawing, chart or graph is in training or in any other kind of work.

No amount of staff work, supervision, special training technology, or enthusiasm on the part of the training organization will accomplish much unless management is interested in the training program.

There has not been the interest on the part of management in training that there has been in other divisions of their business, with many exceptions. Unless those exceptions can be made to grow and accelerate the training movement will not have the successful career to which it is entitled and which is needed by the industry.

[1]University of Detroit, Detroit.

NONDESTRUCTIVE INSPECTION OF CASTINGS

Intelligent application of nondestructive test methods, old and new, by trained inspection personnel working with sound acceptance standards enables the foundry to determine exact locations of defective areas, change foundry techniques to eliminate them, and produce quality castings at lower cost.

Clyde L. Frear
and
Robert E. Lyons
Bureau of Ships, USN.
Washington, D.C.

IN GENERAL, nondestructive testing includes any method of inspection, practicable or impracticable, efficient or inefficient, which will disclose the presence of defects in a part without destroying the part itself. When applied to castings, the definition may be extended to cover those methods which, while they may remove part of the casting, leave the final part retaining its specific dimensions and still usable.

Examples of such a method are drilling out bosses and removal of extraneous metal such as padding and risers which were added to provide feeding, thus exposing areas which upon visual examination may give some idea of the soundness of the underlying metal.

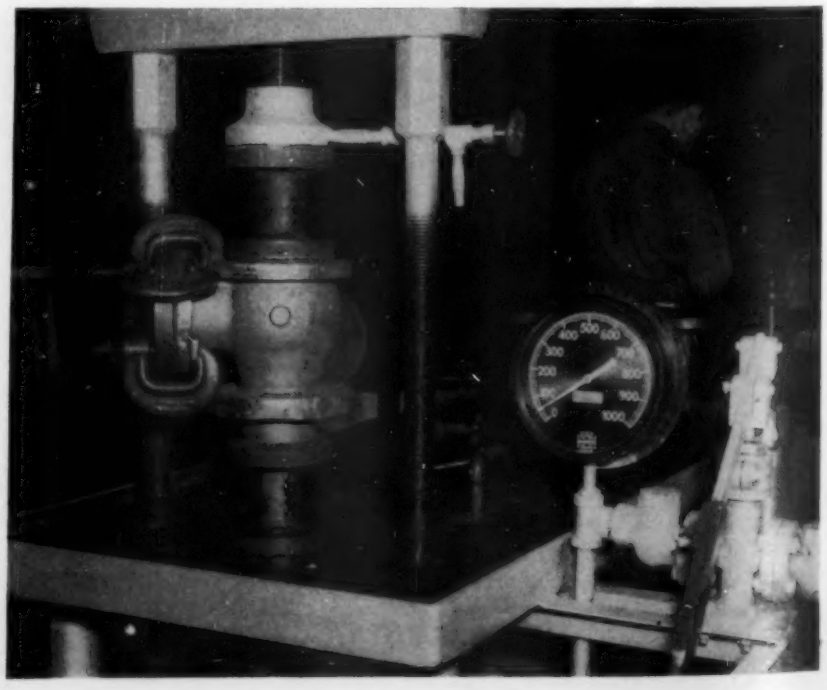
A thorough inspection of castings should be capable of determining the presence of internal as well as external defects, and provide a fairly complete knowledge of the extent of such defects. In the past, surface quality has often been an important factor in determining the acceptability of castings, and much development work has been carried out to improve the surface appearance. To the dismay of a number of foundrymen who have prided themselves upon the excellent finish of their castings, defects have been disclosed upon subsequent machining or pressure testing.

A fine finish is important in many items, among which are statuary, name plates, medals, hardware, etc., but is certainly of secondary importance when the casting is to resist pressure or heavy loading stresses. In such cases the major effort should be exerted in obtaining internal integrity which, in many cases, can be obtained only at the expense of surface appearance.

It becomes necessary then to provide some method of inspection which will show just what is inside the casting and permit of an accurate interpretation of the amount and extent of any discontinuities and their probable effect upon the final serviceability of the part. For years foundrymen have been sectioning castings to determine their internal soundness.

This is far from being a nondestructive test and, although it shows defects present in the area sectioned, it does not indicate the quality in any other area. The writers know of one case where a casting apparently was sound, as shown by sectioning, yet a large shrinkage cavity extended to within

Fig. 1—Typical arrangement of apparatus for hydrostatic testing of valves and fittings (3½-in. bronze angle valve under test).



This paper was presented and discussed at an Inspection of Castings Session of the Fiftieth Annual Meeting, American Foundrymen's Association, at Cleveland, May 6-10, 1946. The opinions expressed in the paper are those of the authors and not necessarily those of the Navy Department.

less than ⅛-in. from the cut and was missed entirely by the sectioning. Certainly as a method of correcting foundry technique the sectioning described was useless.

Although sectioning, or fracturing, is a valuable method of inspection in many cases for the correction of foundry technique, it is absolutely useless in checking those accidental defects which often occur despite the care used in the foundry and the standardization of technique.

Nondestructive Inspection Methods

Nondestructive inspection methods may be considered to include the following:

1. Visual inspection.
2. Sound or percussion tests.
3. Impact tests.
4. Pressure tests.
5. Radiographic examination.
6. Magnetic inspection.
7. Electrical conductivity.
8. Penetrants.
9. Supersonics.

A brief description will be given for each method and an attempt made to outline the advantages and disadvantages of each. No attempt will be made to list all methods which might be made to serve as nondestructive tests, nor to list many of the newer tests which are in a developmental stage. All methods listed are in regular use in one or more foundries.

Visual Examination. Visual inspection constitutes the oldest and most widely used inspection method in existence. In fact, it is so extensively used that many forget to list it as an inspection method. Its importance will become even more apparent when it is considered that the success of most of the nondestructive methods described in the following paragraphs depend upon visual inspection for interpretation.

As the name implies, this method consists of examining the surface of the casting with the unaided eye, the use of a hand lens or, in certain cases, low-power microscopes. The number of surface defects which may be found by a trained inspector but which might not be located by the ordinary observer is sometimes surprising.

Among the types of defects which are found by visual inspection are "burned-in" sand, blows, rat tails, cold shuts, swells, surface porosity and surface cracks. The hot tear is another important defect which is commonly considered as internal. Many of these, perhaps all of them, extend to the surface and may be picked up by visual inspection, unless the tear or crack is too narrow to be observed.

One of the most important uses of visual inspection is to determine whether a casting is within the specified dimensions. No details of this use need be given as the subject has been covered by papers previously presented to the Association.

Numerous defects which ordinarily are internal become exposed during machining. It is rather late to discover the presence of defects at this point, especially if the discovery is made after the finish machining operation has been completed. A number of metals, of which steel is an outstanding example, when found to contain defects after machining, can be repaired by welding or other suitable means, especially if a machining cut is still to be made.

Welding operations usually cause a small amount of warpage and, if the part is to be finished to close dimensional tolerances, every effort should be expended to determine the presence of any harmful defects and make necessary weld repairs before the finish machining operation.

This is true even when the repair is made in an unfinished area of a casting which has already been machined to dimension, as the heat from the welding operation may cause sufficient warpage to make the part useless because of misalignment of the finished areas. Added to this, stress-relief heat treatment, often necessary after weld repairs have been made, may injure finished surfaces sufficiently to make the use of the part impracticable.

Internal Flaws

Valuable and necessary as visual examination is, it will not take the place of inspection methods which will show the presence (or positive absence) of internal defects. However, it is necessary that the inspector, to be capable of correct interpretation of nondestructive test results, must have thorough training in locating defects and recognizing them as such.

Many inspectors evidently have lost sight of the fact that it is just as much their duty to accept satisfactory material as it is to reject material which is not suitable for the purpose intended. It is always easier, when in doubt, to reject a part, and if such doubt does exist perhaps rejection is the only answer, at least until such time as this doubt can be removed.

With training and experience the inspector will become more certain of himself and will be able to accept a greater percentage of parts without endangering their usefulness in service. It should be the chief purpose of the various methods to remove, insofar as possible, the doubt which may exist in the inspector's mind as to whether a part may be accepted, repaired, or rejected without qualification.

Sound or Percussion Tests. Next to visual testing, the sound or percussion method probably is the oldest method of inspecting for discontinuities without destroying the part. A number of variations and refinements of the method have been made, but basically it consists of suitably supporting the part or suspending by chains or other equipment, permitting the part to swing free of the floor and other obstructions, then tapping with a hammer.

Weight of the hammer blows must be selected and adjusted to set up vibrations in the part, thus producing a certain characteristic tone which may or may not be changed in pitch or quality by the presence of discontinuities which tend to change the wave length of the sound produced by the blows.

Limitations

This test is analogous to the testing of china and porcelain ware for cracks by tapping with a pencil or the fingernail. Such a method may serve to detect relatively large discontinuities in otherwise homogenous metal, and probably would be successful to a certain extent with forgings having simple shapes and fairly uniform cross sections.

On the other hand, it is not difficult to imagine relatively large internal defects which extend in certain directions but which would not alter the length or quality of the sound vibrations set up in the piece, in which case no indications of the defect would be noted.

Most castings are complex in shape, which in itself may hinder the production of a pure tone upon being struck with a hammer. The

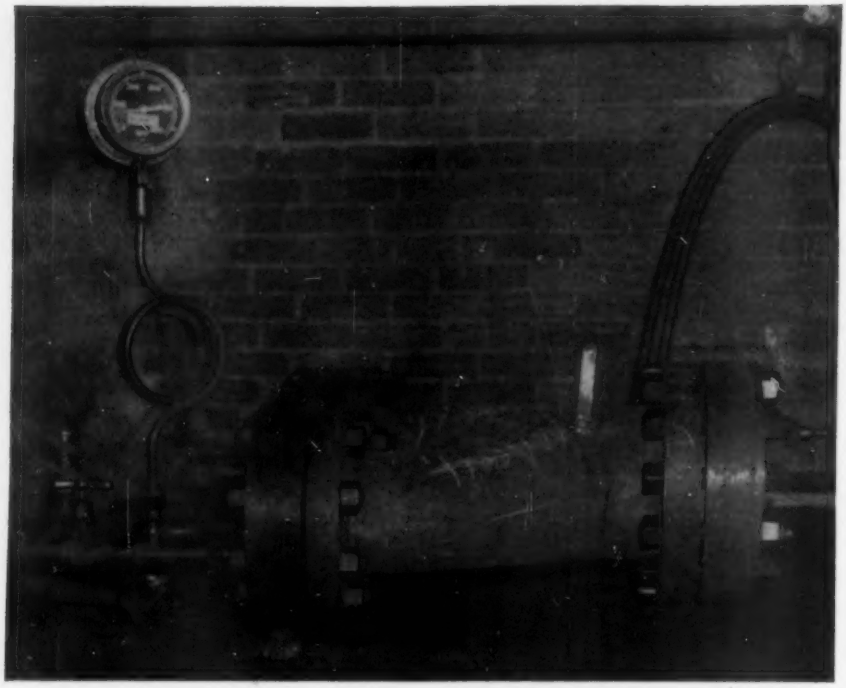


Fig. 2—High pressure steel fitting under steam test.

orientation of the crystals will also tend to dampen the tone.

Considering the harmless discontinuities which may exist in a large percentage of castings, all of which may have an effect upon the character of the vibrations, it can be expected that a relatively large internal defect might be required to change the tone sufficiently to be detectable by the ear with sufficient selectivity to distinguish between a harmful defect and one which would not be harmful, and to determine whether such a defect is located in a highly-stressed area.

The use of a stethoscope for locating abrupt changes in the tonal quality has been recommended, but even by this means it is doubtful if a small but harmful defect could be detected with certainty. In general, such a method would not be suitable with the metals having high damping characteristics.

Impact. As a method of nondestructive inspection, testing by impact must not be confused with the commonly known impact tests such as the Charpy and Izod tests, which require a specimen to be broken during the test. One form of nondestructive impact test would be an extension of the sound test described in the previous paragraph, the only difference being that larger hammers are used and heavier blows are struck.

Whether or not it may be considered as a nondestructive test often depends upon the amount of energy expended in the blows. Another version of the impact test, the drop test, consists of dropping the casting from a certain specified height, usually onto a steel or other solid plate of certain specified thickness and weight.

The expected value of such a test is the hope that all castings which contain harmful defects will break and be rejected automatically, and that those containing no defects will withstand the impact of the drop and be passed as acceptable.

Observations

However, instances have occurred in the experience of the writers where castings have been accidentally dropped from heights of less than that specified and onto wood blocks or concrete floors and suffered breakage in areas having no defects. In other instances, castings have successfully passed the specified drop test only to be put into service and break under loads considerably lower than those exerted in the test.

In this latter case, small defects already present in the casting probably withstood the test, but increased sufficiently under the influence of the impact stresses to lower the strength of the part to a point considerably below that required to withstand the stresses applied in regular service.

Another form of the test consists of dropping a specified weight from a predetermined height upon selected portions of the casting. This form of test is an improvement over the regular drop test in that the application of the load is under better control.

However, this does not eliminate the chance of breakage or propagation of defects. It has taken only a few instances of breakage similar to that described in the foregoing to make the writers believe that drop tests and other forms of impact tests are not to be trusted as acceptance tests for castings, and in many cases may do more harm than good.

In those cases where such a test is considered necessary, the tests should be followed by others capable of determining the presence of discontinuities which may have resulted from the impact stresses.

Pressure Testing. Pressure tests are used to locate leaks and to test the over-all strength of parts such as valves, fittings, boilers, etc., which are to carry fluids in service, and may be divided into hydrostatic, steam, and air tests.

Hydrostatic testing consists of sealing all openings in the casting except a small opening for the introduction of a water pipe and another for a vent, the latter opening being as nearly as possible at the highest point in the casting.

A vent is provided in order that all air may be removed from the part before the application of the pressure. The part then is completely filled with water and the pressure gradually increased to a specified point, which usually is one and one-half or two times the pressure for which the casting is designed.

For obvious reasons where leaks are suspected, the pressure should be applied slowly. Fig. 1 shows a typical set-up for hydrostatic testing of valve bodies and fittings. The illustration shows a 3½-in. bronze angle valve under test.

It will be noted that the inlet and bonnet openings of the valve body are closed by rubber gaskets, the necessary pressure to prevent leakage around the gaskets being applied by a large screw passing through the yoke shown at the top of the photograph. This screw is operated by

means of a hand wheel located above the yoke. The outlet is closed by a gasket and steel plate held in place by C-clamps.

The small valve shown at the top center is opened and water is pumped into the casting under test by means of the hand pump shown at the lower right. When water flows from the valve at the top, showing that all air has been forced out, the valve is closed and pumping continued until leaks are found in the valve body, or until the specified test pressure is indicated on the gauge.

With the test pressure being applied, the casting is carefully examined for small leaks and their location. If the operator has been successful in obtaining an absolutely tight system, with no leakage around the gaskets, through the pump or in the piping, then even a very slight leak or seepage will be indicated by a dropping off of the pressure shown on the gauge.

General Considerations

As fluid pressure is exerted in all directions, the hydrostatic test gives a measure of the over-all strength of the casting. Little or no chance of danger to personnel exists should the casting fail under the test pressure.

Water being almost incompressible, any enlargement of the part undergoing test will cause an instantaneous reduction or relief of the pressure, thus preventing shattering of the part. For this reason, application of pressure by a high-speed pump should be avoided. A small hand pump is preferred for this purpose.

In addition to testing for over-all strength, the hydrostatic test will locate individual leaks through the wall. The action of these individual leaks has given rise to the expressive but not highly accurate terminology of weepers, seepers, leakers and squirters.

The hydrostatic test shows over-all strength and the presence of leaks only *at the time the test is made,* and gives no real assurance that the part will continue to be free from defects in service. Cases have been known where a casting was shown to be sound by a hydrostatic test and, when the pressure was removed, the casting leaked under a reapplication of the testing pressure.

It is well known that most foundries desire, insofar as possible, that no machining be performed on the pressure parts of castings, and almost insist upon one unmachined surface in order that an unmachined skin will be left to resist the pressure, the interiors of most castings being less dense than the outer layers.

Another consideration is the fact that most castings are considerably overdesigned, not only to provide a generous factor of safety but to obtain walls which are sufficiently heavy to be run successfully. With this last in mind, it is easy to understand that a shrinkage defect which may occupy as much as 75 per cent of the wall thickness might not manifest itself as a leak under the specified pressure test.

Many shrinkage cavities appear to be smooth-walled when the casting is sectioned, and have been considered harmless provided that sufficient metal has been left in the wall to give the necessary strength.

However, close examination under the microscope has shown that many of these apparently smooth-walled cavities are filamentary in character, and sometimes these filaments have the characteristics of fine hot tears or cracks which extend almost through the casting wall.

Stress Variations

When such a casting is subjected to high-pressure steam or to alternate high and low pressures, the effect of the expansion due to heat or the straining of the walls under pressure may sooner or later cause the filamentary shrinkage cracks to extend, thus causing the defect to reach entirely through the wall. This phenomenon probably accounts for many of the leaks in pressure lines which have developed in service in parts which withstood pressure tests.

This fact should not be considered as condemning pressure tests, but is given as one explanation of something which has occurred many times. The writers are in favor of pressure tests on pressure castings, in conjunction with other nondestructive tests, as these tests most nearly approach actual service conditions. They wish only to point out

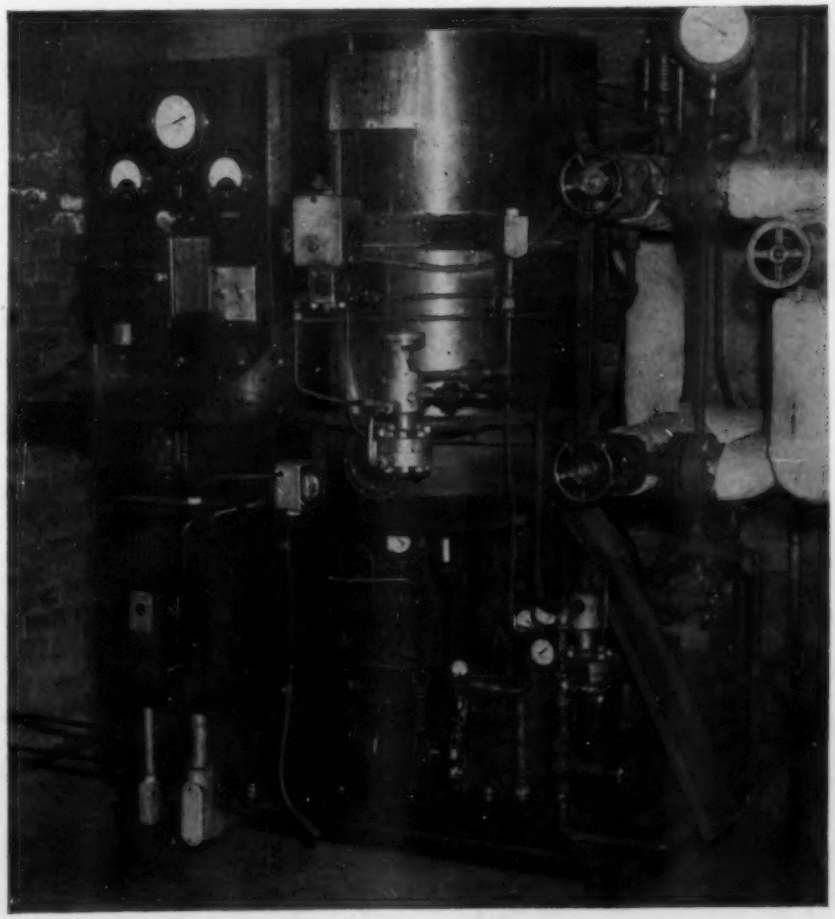


Fig. 3—Automatic boiler used in steam tests

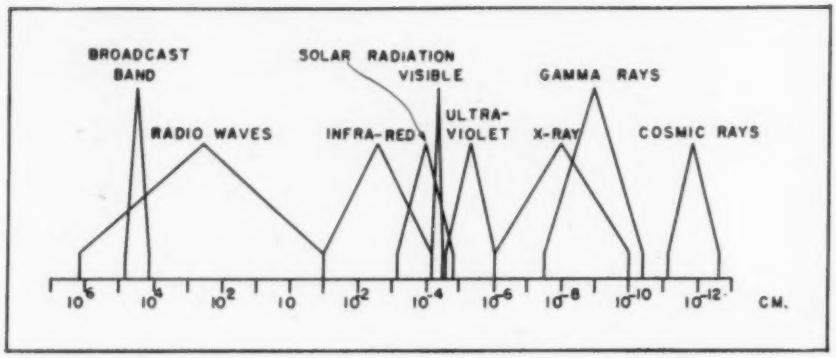


Fig. 4—Electromagnetic radiation diagram.

that no single test will give complete assurance of absence of harmful defects.

Steam Tests. When fittings, valves, etc., are to withstand high pressure steam service, a steam test is often applied. In most ways such steam tests are no better than hydrostatic tests except that the steam, being at high temperature, has a tendency to expand the metal and open up small defects which might have been missed by the hydrostatic test.

In addition, steam and other gases will pass through very small openings more rapidly than will liquids. Steam tests should be used only as adjuncts to hydrostatic tests, and should never be applied until the part has been hydrostatically tested at considerably higher pressure than the steam pressure actually applied in the test.

Because of high compressibility, steam (and other gases) will retain most of its pressure after considerable enlargement of the part, and any slight reduction in pressure may be built up again in a short time. The high pressure acting on the weakened walls of the part under test may cause breakage with sufficient violence to injure personnel by flying pieces or scalding.

Figure 2 shows a steam test being made on a steel "Y" casting for high pressure steam service. Although any source of high-pressure steam which may be available is suitable for this purpose, the automatic boiler illustrated in Fig. 3 is convenient for producing a continuous supply of steam under any predetermined pressure for any desired time with but little attention from the operator.

Air Tests. Testing with air is an adaptation of the hydrostatic and steam test, and is conducted in the same manner except that air is used in lieu of water or steam. In general, it is simpler to apply and requires comparatively inexpensive equipment, and because of its convenience is used to a great extent in the production inspection of small valves and fittings, especially those which are to operate under comparatively low pressures.

However, it may be used with larger parts, especially in those cases where a few pieces are to be tested and no regular equipment for hydrostatic testing is available. Small parts are usually tested by immersion in water or other suitable liquid. The air pressure is then applied internally and the air seeps through any openings in the part and forms bubbles of air on the surface, thus permitting ready location of leaks.

Instead of immersing the part in a liquid, which would be inconvenient with large parts, soap solution may be spread over the surface with a brush. Leaks will then be found by observing bubbles and frothing of the solution where air passes through the wall.

Radiographic Examination. Radiography was the first development which provided a means of actually looking through castings or other opaque materials and determining the presence of internal discontinuities and, in certain cases, heterogeneity of the material. Although it is 50 years since Roentgen discovered the ability of certain rays to penetrate metal, it was not until the 1930's that sufficiently powerful equipment was developed to make the use of x-rays practicable for inspecting castings.

Concurrent with the development of industrial x-ray equipment, the Naval Research Laboratory developed the use of gamma-rays emitted during the decomposition of radium and radium salts as a means of inspection.

These rays, which will penetrate materials commonly considered as opaque, are not mysterious if it is realized that opacity to light is merely the property of absorbing light rays to the extent that the rays are not transmitted through the material.

Characteristics of Short Waves

Many materials which are commonly considered as opaque become transparent in thin sections, a typical example being gold. With this consideration in mind, opacity becomes a comparative term, and it is readily conceivable that it might be possible to generate light rays to which ordinarily opaque materials would become transparent.

X-rays and gamma-rays are electromagnetic rays having wave lengths of approximately 10^{-8} to 10^{-10} cm. Light rays visible to the eye have considerably longer waves, their wave lengths lying between approxi-

Table 1

RADIUM DECOMPOSITION PRODUCTS

Name	*Type of Radiation ( ) = Weak Radiation*	*Half Life Period*
Radium	Alpha (Beta and Gamma)	1,590 yr.
Radon (niton, or radium emanation)	Alpha	3.825 days
Radium A	Alpha	3.05 min.
Radium B	Beta (Gamma)	26.8 min.
Radium C	99.97 per cent Beta and Gamma	17.9 min.
Radium C′	Alpha	Cal. 10^{-6} sec.
Radium C″	Beta	1.32 min.
Radium D (radio lead)	(Beta and Gamma)	19 yr.
Radium E	Beta	4.9 days
Radium F (Polonium)	Alpha (Gamma)	140 days
Radium G	——	——

mately 10^{-5} and 10^{-4} cm. (Fig. 4). As the lengths of these electromagnetic waves become shorter their penetrating power increases until, finally, they are able to penetrate objects opaque to ordinary light waves.

For the sake of emphasis it should be repeated that opacity and transparency are only relative terms. Thus a thick piece of glass would tend to be more opaque and less transparent than a thinner piece of the same glass, which is the same as saying that the thick piece would absorb more light rays than would the thinner piece.

Extending this reasoning, a piece of smoked glass or a lens from a melter's goggles would absorb more light than would a piece of clear glass of the same thickness. X-rays and gamma-rays follow these same rules. With thicker metal, more rays are absorbed and fewer pass through; also, the more dense the metal as measured by its specific gravity, the less the penetration of the rays.

Consider a piece of metal plate which is apparently sound by visual inspection but is known to contain a localized shrinkage defect in the center. If this plate is exposed to x-rays or gamma-rays, some of the rays striking the plate will be absorbed by the metal and stopped in their path, the remainder passing through.

Radiographic Indications

Because of less metal where the shrinkage occurs, more rays will pass through the plate at this spot. Rays passing through cannot be seen as they are invisible to the eye. However, they will cause a fluorescent screen to glow, the intensity of the glowing being proportional to the intensity of the rays striking the screen.

Therefore, examination of this screen discloses the location of the defect by the presence of a brighter spot. The silver coating of a photographic film will absorb a certain number of the rays passing through it, the number absorbed being proportional to the number striking it.

These absorbed rays have an effect on the film similar to that of light. Therefore, if a film is placed against the metal plate on the side opposite to the source of the rays, the rays passing through the plate and then through the film will affect the film to a greater extent immediately under the defect through which the greater number of rays penetrated.

If this exposed film is developed, it will show darker in that area where the defect was located. It becomes evident, therefore, that internal defects in a casting can be located by the passage of x-rays or gamma-rays through the casting and recording the intensity of these rays upon a photographic film.

Special films with an emulsion coating on each side have been developed for radiography, thus permitting greater ray absorption than single-coated films and hence recording the presence of defects with greater assurance and shorter exposure time.

Radiographic examination includes the use of x-rays and gamma-rays; also, the use of a few other materials which emit rays capable of penetrating heavy materials. Whether x-rays or gamma-rays are used depends upon the availability, power of the equipment, type and size of the material being tested, portability of the equipment, and often upon the convenience.

To a certain extent the two sources are interchangeable, but each has certain advantages and disadvantages which should be thoroughly understood, especially when deciding upon the purchase of equipment.

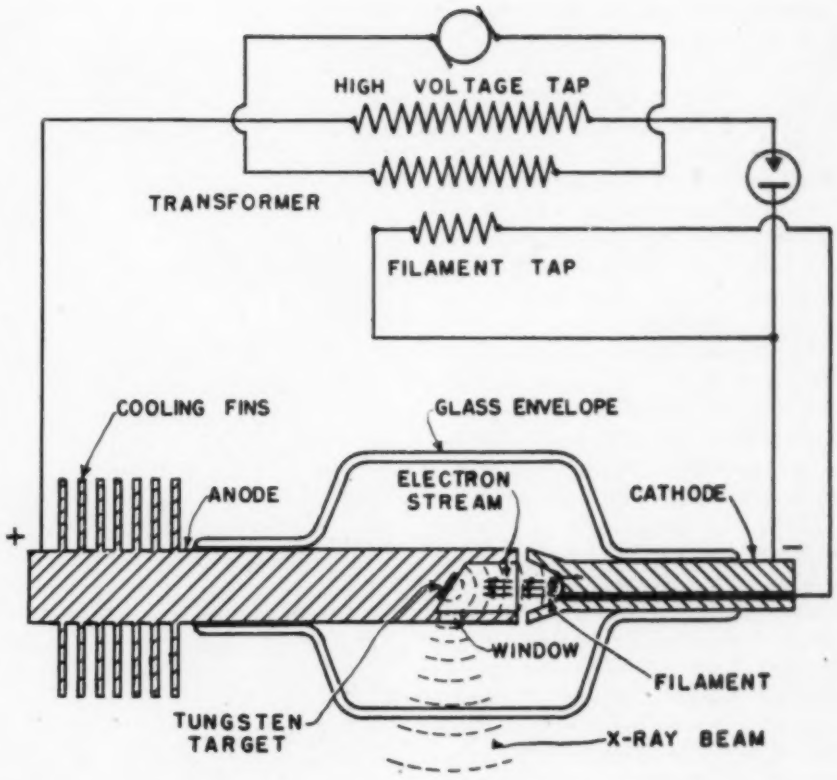


Fig. 5—Construction diagram of simple x-ray tube.

X-rays. These rays are produced by the impact of electrons upon a target, part of the mechanical energy resulting from the motion of the electrons being changed to energy in the form of electromagnetic waves.

Figure 5 shows a simple form of x-ray tube, which consists of a copper anode and cathode sealed into an evacuated, heat-resistant glass bulb. The inner end of the cathode is cup-shaped and is provided with a small, coiled tungsten filament, one lead of which is insulated from the cathode itself.

In operation this filament is heated by passing a current under an emf. of approximately 12 volts taken directly from a special step-down transformer or from a low-voltage tap from the main power transformer. The passage of the current and heating of the filament causes a build-up of electrons around the filament.

Electrons are negatively charged and have a certain small mass or weight. The outer end of the anode

is arranged for cooling, and the inner end adjacent to the filament is provided with a small tungsten insert or target. When a high-voltage direct current of the order of 50,000 to 250,000 volts is impressed on the tube as indicated in the diagram, the negative charge on the cathode rejects the electrons which surround the filament.

If the filament current continues to flow, electrons are continually formed and rejected by the high-voltage negative charge. The negatively charged electrons are held to a relatively straight and narrow path because of the opposition of the negatively charged cathode cup.

After traveling across the space separating the cathode and the anode, the electrons strike the tungsten target where their motion is suddenly stopped and their kinetic energy is converted, mostly into heat which is conducted through the cooling fins. Remaining energy is converted into electromagnetic waves which, theoretically, leave the target in all directions.

The X-Ray Beam

Most of the unwanted rays are absorbed by the metal of the anode and cathode, and often by lead which surrounds the tube. A "window" is provided to permit certain of the rays to leave the tube in the form of a cone or pyramid, depending upon the shape and location of this window, which is designed to give a beam of the desired dimensions. By manipulating the tube, the x-ray beam can be aimed in any desired direction.

The number of electrons emitted from the filament is proportional to the amount of current (usually measured in milli-amperes) impressed across the tube, and it follows that the intensity of the rays is also proportional to the current. The speed of the electrons is increased with increase of voltage (usually measured in kilovolts).

As the speed of the electrons is increased the wave lengths of the x-rays produced become shorter. In short, the wave length is inversely proportional to the voltage.

Sensitivity, or the ability to register small defects, is dependent upon the wave length of the x-rays, the greater the wave length, the greater the sensitivity. It would appear from

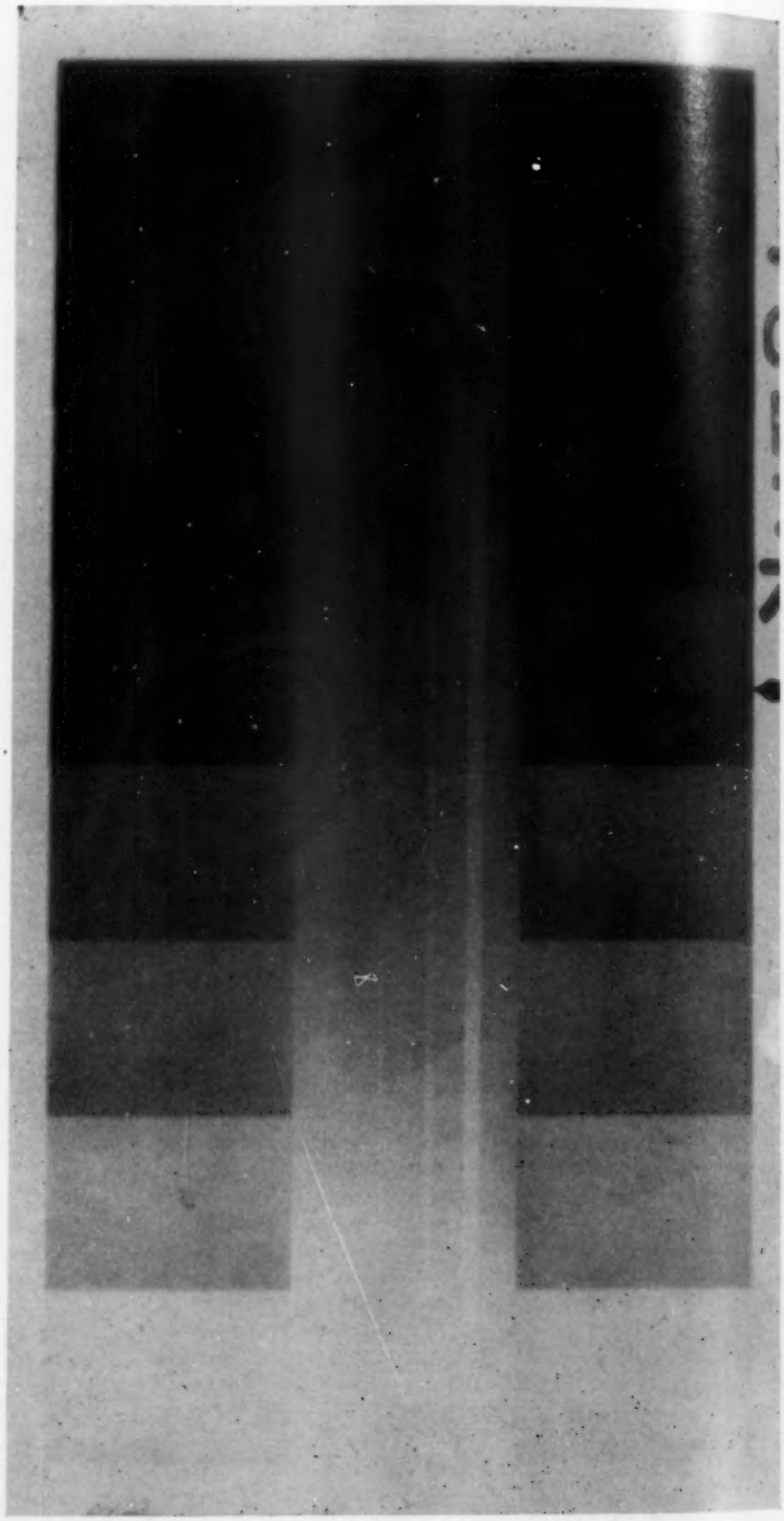


Fig. 6—X-ray radiograph of ⅛ to 1 in. stepped test block. Radiograph made using 185 kv. and timed for ⅝-in. section. Note: High contrast, and lack of detail at thin section (top) and heavy section (bottom); also, sensitivity indicated by clear rendition of milled slots at midlength.

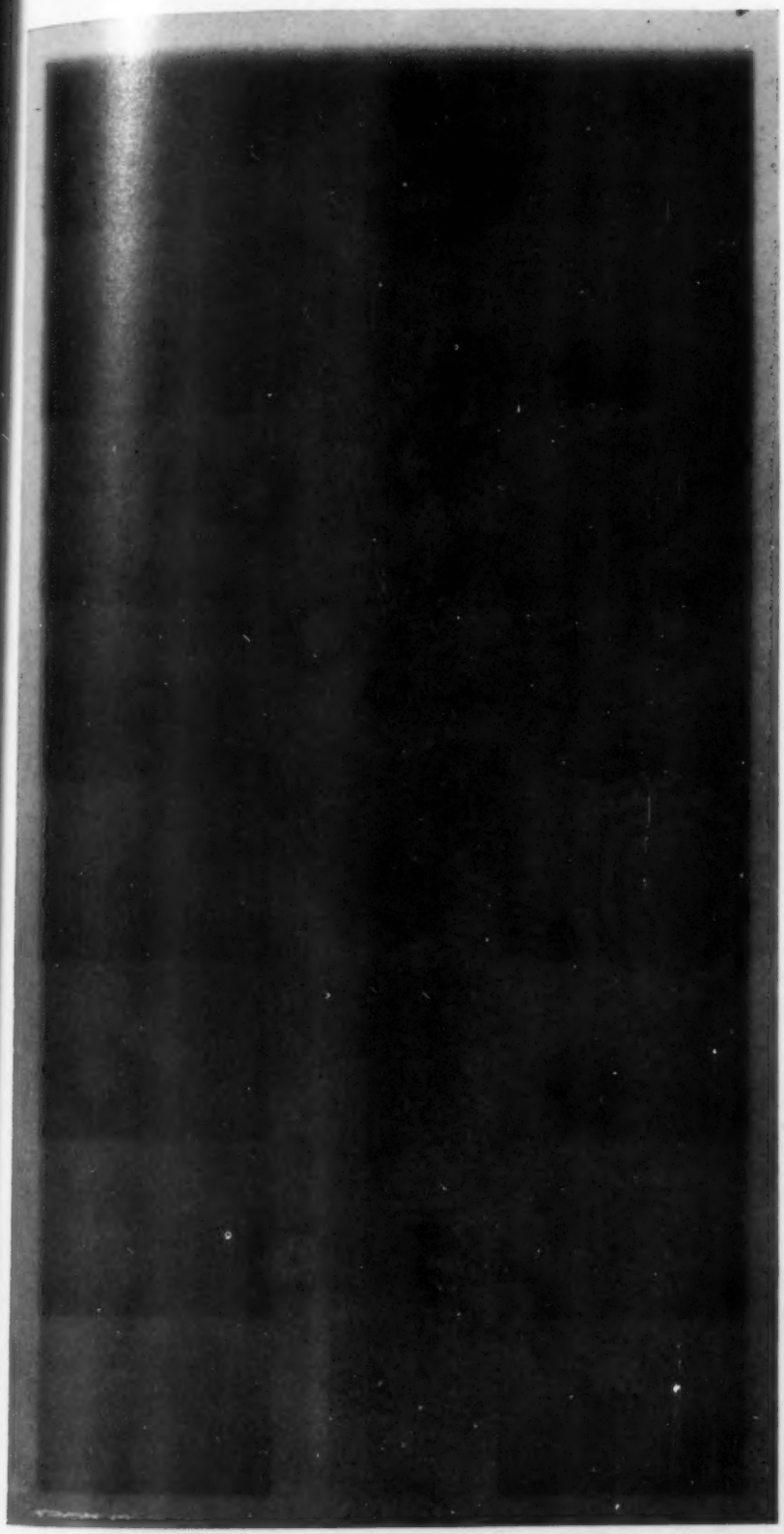


Fig. 7—Gamma-ray radiograph of same block shown in Fig. 6. Radiograph made using 500 milligrams of radium. Note: More uniform density and lower contrast by comparison with Fig. 6; also, decreased sensitivity indicated in less pronounced rendition of milled slots.

this that low-voltage x-rays should always be used for assurance of finding the smallest possible discontinuities, but other factors must be taken into account.

Probably the most important from a production standpoint is the fact that there are limits in thickness that rays from a particular voltage source will penetrate. Equipment of 150 kvp. (kilovolt-peak) is not of much use with steel thicknesses above one in. but is suitable for aluminum and magnesium of greater thickness.

Equipment rated at 220 kvp. is suitable up to 3 in. of steel, but above this the exposure time becomes excessive. Rays from 1000-kvp. equipment can be used practically up to 6 in. of steel, and often up to 8 in. or somewhat higher. Machines operating at 2000 kvp. have been introduced on the market, and can be used with considerably greater thicknesses.

Another difficulty is encountered in the use of low-voltage x-rays, especially in the inspection of castings. Few castings are uniform in cross section, and when radiographing nonuniform sections the problem of density and contrast enters.

Exposure Factors

Contrast can be defined simply as the difference in darkening of the film caused by the rays passing through a defect as distinguished from that caused by rays passing through the adjacent sound metal. Up to a certain point, contrast is synonymous with readability. Films from low-voltage x-rays should have more contrast than those from higher voltages.

However, if a wedge ⅛-in. thick at one end and 2 in. thick at the other is radiographed with a single exposure using 150 kvp. x-rays, the length of exposure (expressed in milliampere-seconds) being calculated for the average thickness, the image of the thin end of the wedge will be too dark to be readable with any illumination of the film, while that of the thick end will be so light that no defects will show.

Figure 6 illustrates this to a certain extent. In this case an x-ray radiograph, known as an x-ograph, has been made of a 4x8-in. steel plate which is one in. thick at one end and tapers by ⅛-in. steps to

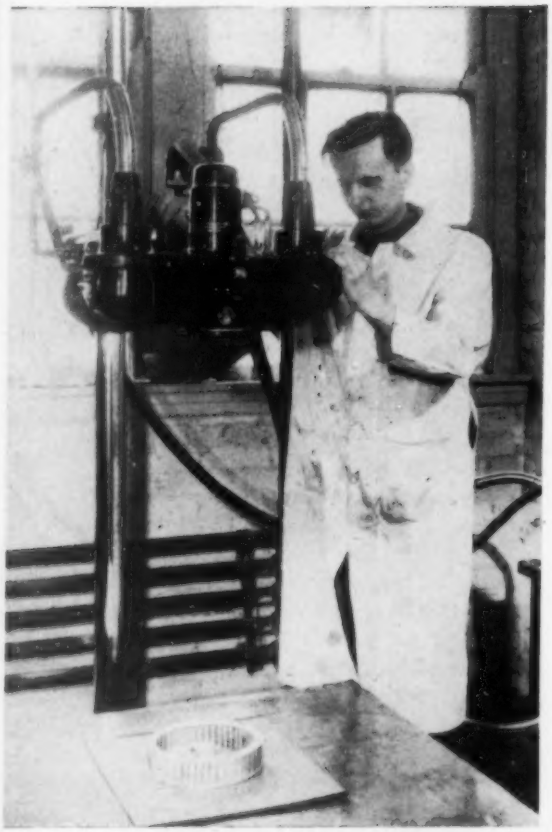


Fig. 8—X-ray head, 140 kvp. Transformer is separate from head necessitating high voltage cables.

⅛-in. thickness at the opposite end. A slot 1¼ in. wide has been milled across these steps to obtain a continuous taper from one-in. to zero thickness lengthwise of the plate.

In this tapered surface have been milled slots measuring 0.01x0.01 in., 0.02x0.02 in., 0.04x0.04 in., and 0.08 x0.08 in., which serve to show the sensitivity which is obtained at any thickness along the plate. This radiograph was made using 185 kv. and shows best rendition of the milled slots midlength of the block, the exposure having been adjusted for the ⅝-in. section.

Although the density obtained at the thin end of the block is not too great to be read by high intensity illumination, little or no detail can be noted by ordinary illumination. The same is true with the heavy end where the radiographic density is too low to register any but large defects. In the midportion of the block, the 0.01x0.01-in. slot is easily seen on the original radiograph, indicating better than 2 per cent sensitivity.

Fig. 9—X-ray head, 220 kvp. Transformer and tube are combined in head. Jib crane mounted.

Increased Penetration

By using rays with shorter wave lengths (those produced by higher voltages), this difference in densities between thick and thin sections becomes less. In other words, the "latitude" increases with decrease in the wave length. Latitude and contrast are inverse to each other.

If it is desired to gain latitude, a loss in contrast will result. In the radiographic examination of castings, the x-ray equipment selected must be a compromise between certain advantages and disadvantages and capable of operating under voltages which will penetrate the metal being tested in a reasonable length of time; also, which will give sufficient latitude in the range of thicknesses to be examined.

Gamma-Rays. Radiography using radium depends upon the emission of penetrating rays produced by the decomposition of radium or radium salts. Table 1 shows the various steps in the decomposition process.

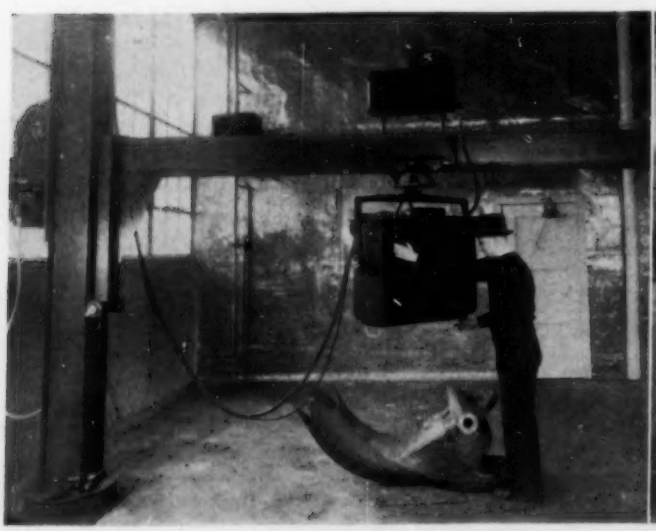
Alpha and beta rays produced by the decomposition have no direct use in radiography, as their path is so short that it is doubtful if they ever leave the container in which the radium is placed. Even if they did penetrate this container they probably would not reach the casting being examined, let alone penetrate it.

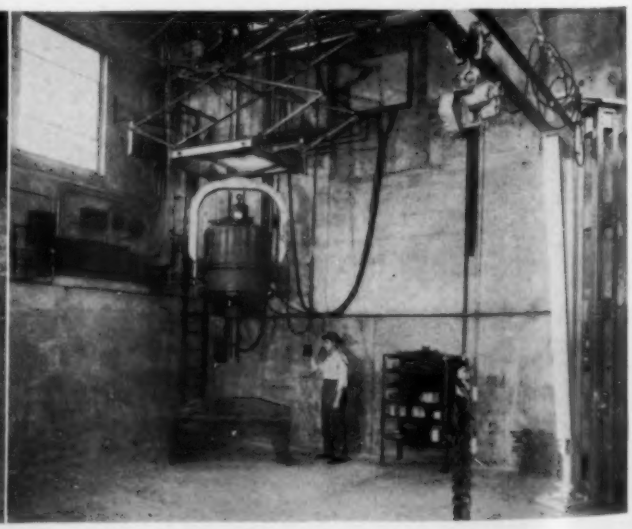
Gamma-rays, on the other hand, have very short wave lengths, and hence high penetrating power. As previously described under x-rays, because of the short wave lengths involved radiographs made with gamma-rays have high latitude, and little or no difficulty will be experienced in showing defects in heavy and light sections x-rayed on same film.

Using the same reasoning, the contrast will be lower than that obtained with x-rays having greater wave lengths, thus making it somewhat more difficulty to distinguish defects recorded on the film than would be the case with the same defect located by x-rays. In this connection, compare Figs. 6 and 7.

In the case of light metal castings and thin-walled castings having fairly uniform cross sections, the use of x-rays is preferred. Because of the bulkiness and weight of x-ray equipment, it usually is necessary to transport castings to a special area set aside for the purpose and where the

Fig. 10—X-ray head, 1000 kvp. Combined tube head and transformer. Pantograph mounted.





equipment has been permanently installed.

Figures 8, 9, and 10 show several x-ray installations, the control panels for each piece of equipment being placed in a separate room. For ease of transportation and setting up for radiographing, radium is much to be preferred, and in many places, such as those where no electrical power is available, it is the only equipment which can be used.

Even where a complete x-ray installation is available, radium is a necessity in certain locations in some castings, where openings are too small to insert a film or an x-ray tube. The radium itself (usually in the form of radium sulphate) is enclosed in a small, hermetically sealed silver capsule, which in turn is enclosed in an aluminum or steel "egg" from which the silver capsule is not removed during its use.

For safety to personnel during transportation from place to place, the radium container is placed in a carrying case provided with approximately 2 in. of protective lead on all sides.

One of the chief drawbacks to the use of radium is the long exposure time required as compared with x-rays. In many cases, convenience overbalances this deficiency. In other cases, several castings can be radiographed at the same time with the same unit of radium, which is not possible with most types of x-ray equipment.

Many of the larger castings cannot be x-rayed conveniently with anything less than 1000-kvp. equipment which, in addition to being expensive, is very bulky and requires a room with heavy walls to provide protection for operating personnel from the effects of the rays.

Where the production of a foundry is sufficient to cover the cost of high voltage x-ray equipment, its use is recommended because of its speed, especially in radiographing heavy sections, and because of its greater sensitivity. However, in most foundries the use of radium probably will continue.

Even the slow speed of gamma-radiography is not necessarily a handicap, as indicated in Fig. 11, which shows 22 valve bodies being radiographed at one time. With such a set-up it often is convenient to arrange a large number of castings in such a manner that the exposure will be continued overnight, when all personnel, including the radiographer, may be absent.



Fig. 11—Twenty-two valve bodies radiographed at one time. Circles painted on floor facilitate placing parts at correct distance from radium (center).



Fig. 12—Multiple set-up employing 1000 kup. x-ray equipment.

It should be mentioned that such multiple setups are not limited entirely to gamma-radiography, as indicated by Fig. 12, where several castings are arranged around the target of a 1000-kvp. x-ray. At present such x-ray set-ups are limited almost entirely to equipment capable of operating at 1000 kv. or higher.

After several years' experience in radiographing steel castings of various sizes and shapes, the writers can say that gamma-radiography has served the trade well in insuring satisfactory castings, and probably will continue in use in most foundries, but they recommend the use of x-rays where practicable and economical.

Before leaving the subject of radiography, it would be well to mention some of its limitations. Although it is at present the only nondestructive method which will show with certainty the presence of deep-seated defects and reveal the types of de-

fects present and their extent, it does fail to show many small defects with certainty, especially closed cracks, often the most harmful defects in a dynamically stressed member.

With low-voltage x-rays in some instances it is possible to locate defects the smallest dimensions of which are 0.5 per cent of the section thickness. In most production inspection, 1 per cent is as much as can be expected, and when using gamma-rays, 2 per cent is about the best that can be obtained.

In some cases it may not be possible to find defects which are smaller than 3 to 6 per cent of the section thickness. Fine surface cracks are difficult and often impossible to locate, particularly in heavy sections and other parts of castings of designs which do not lend themselves to radiographic examination.

Magnetic Inspection. As previously stated, radiographic inspection often is unable to reveal fine cracks and other harmful defects of dimensions smaller than a certain percentage of the section thickness.

In other cases, defects occur in large castings in locations which are of such shape and size that it is impossible to place the film or the source of radiographic energy to obtain suitable rendition of the defect on the film. Furthermore, harmful defects sometimes occur in the most highly stressed sections of the casting which, however, may be too heavy for radiographing. Many of these castings are of such design that to date it has not been found possible to prevent the occurrence of cracks with assurance, yet these cracks are at times so fine or tight that they are difficult or impossible to find by visual inspection, radiography, or other means of nondestructive testing so far described.

General Characteristics

It is desirable, therefore, to have a method of inspection which will locate the presence of such defects. Magnetic inspection serves this purpose provided the part being tested is capable of being magnetized.

In general, the magnetic test consists of magnetizing the area to be tested and applying magnetic particles in the form of a powder, by dusting or spraying with an oil suspension of the particles. If the piece being tested contains no discontinuities, the magnetic lines of force will tend to folow a continuous path through the material.

If a defect or other metallic discontinuity is present, these lines are forced to find a path around the obstruction. If the discontinuity is at or near the surface, some of the lines of force will pass through the air and cause the magnetic particles to pile up, thus making the presence of the defect readily apparent by visual inspection.

Although a strong permanent magnet or an electromagnet may be used to magnetize the part under test, magnetization usually is more conveniently accomplished by passing an electric current directly through the piece. The current in its passage through the metal produces circular magnetic lines of force in a plane at right angles to the direction of current flow.

Test Methods

The most popular procedures for applying the current to the castings are the prod and the all-over methods. A third method, consisting of wrapping a conductor such as ordinary welding cable around the part in the fashion of a solenoid, usually is more convenient in testing finish machined parts where even slight arcing at contacts might burn the finish.

The prod method consists essentially of pressing two conductors, spaced approximately 6 to 8 in. apart, against the surface of the casting at the area being inspected, thus placing that part of the casting in series with the electric circuit.

While the current is being applied, magnetic particles (iron filings cloaked in colored powder) are dusted over the casting surface between the prods and the excess powder gently removed by blowing obliquely over the surface by lung power, or preferably by a controlled air blast such as may be obtained by using a 5/16-in. I.D. air or oxygen hose with the air throttled to about 2 psi. by means of an acetylene or other low-pressure regulator.

Magnetizing Equipment. The chief requirement for magnetic powder inspection is equipment which will supply current of sufficient amperage at as low voltage as practicable.

Welding generators, when used with certain precautions, will serve to produce the required current for magnetic powder inspection when using the prod method. When amperage in excess of the capacity of a single generator is required, machines may be connected in parallel.

Arcing

When using welding machines as the source of current, a contactor switch should be provided to break the circuit before removing the prods since, because of the comparatively high voltage involved, an excessive amount of arcing will occur if the circuit is broken by removing the prods.

Arcing may cause surface cracking and, in the case of higher carbon and alloy steels, will produce hard spots which will interfere with machining. Arcing should be avoided especially on finished machined surfaces.

Storage batteries provide a readily portable source of current. Each storage cell will supply approximately 300 amperes and, by connecting the cells in parallel, multiples of 300 amperes may be made available. The low voltage (2 volts) eliminates the danger of harmful arcing, and therefore the need for contactor switches.

When using storage batteries care should be taken that individual contact periods be made as short as practicable, consistent with efficient examination, in order to prevent excessive discharge of the cells. Because of the rapid discharge an ammeter should always be provided in the testing circuit to insure the necessary current density at all times.

Specially designed magnetic equipment is available which will provide a convenient means of obtaining any desired amperage for the prod method, all-over method, or the wrap-around method. Although convenient contactor switches are supplied the equipment is designed to work at low voltages, thus minimizing the dangers resulting from arcing.

The all-over method is similar to the prod method except that, instead of spacing the electrical contactors a short distance apart, contact is made to each end of the casting, or other extremities chosen to control the flow of the current and the direction of the magnetic flux. The whole casting between the contacts becomes magnetized.

With this greater distance it is

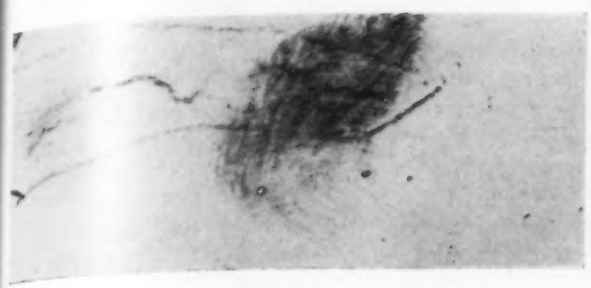


Fig. 13—Defect in bronze casting as seen by visual inspection.



Fig. 14—Same defect under black light after fluorescent oil treatment.

necessary to use higher currents, usually of several thousand amperes, depending upon the size and shape of the casting, to compensate for the greater spacing of the contacts. Magnetic particles are sprinkled over the whole casting surface and removed in the same manner as in the prod method.

In general, the all-over method is particularly applicable to large castings because of speed gained in its application, but is not as sensitive as the prod method by which better control of the magnetizing current may be obtained. The prod method usually is found more convenient and accurate for the inspection of smaller castings weighing under 500 lb.

Because of the possibility of using larger current concentrations, the prod method is more suitable than the all-over method for locating defects which lie slightly below the surface of the casting.

General Practice

Magnetic inspection is often limited to certain critical locations of castings. Care should be exercised in applying the magnetic particle test, as operators unfamiliar with the various types of defects and their location in the casting are likely to use the test indiscriminately, requiring the removal of metal under every indication, whether harmful or not.

It is not unusual to limit the use of the test to the detection of surface hot tears and cracks, other defects which are indicated by the test being dressed with a grinding wheel to determine the depth, often rechecking the ground area before chipping when this is found necessary.

A variation of the magnetic test consists of using magnetic particles covered with material which fluoresces under ultraviolet light. As the indications produced by the glowing particles are much more distinct than those resulting from ordinary particles, this method is especially useful in testing for defects in the bores of cylinders, inside of tubes and other locations where the lighting is such that the usual magnetic powder indications are difficult to distinguish.

Electrical Conductivity. This method is one of the simplest yet probably the least used of any nondestructive method known, and where applicable gives good results which are reproducible from casting to casting. It is practicable only where a large number of castings of the same design are made within close dimensional tolerances and rather close composition control.

Procedure

The method consists of passing an electric current through the piece, or selected areas on the piece, using constant voltage. The success of the test lies in the fact that, if all pieces are of the same size and contain no defects, the conductivity of each piece will permit the same amount of current to pass.

If a nonconducting defect exists in the casting, the actual conducting cross section is reduced and the resistance is increased proportionately. This increase in resistance permits a lesser amount of current to pass, which fact is determined by an ammeter in the testing circuit.

It is evident that this method is applicable only in cases where a large number of similar castings are to be inspected, and requires a number of parts which are known to be sound in order to standardize the equipment.

Penetrants. The writers have attempted to describe the fields of usefulness of radiographic and magnetic particle inspection, and to point out that the latter method is the only one on which reliance can be placed to show positively the presence of fine surface defects which cannot be located by visual inspection, radiography, or other known nondestructive tests.

Unfortunately, the magnetic particle method cannot be used with nonferrous or austenitic alloys, although some of the latter may be slightly magnetic because of a small amount of retained alpha iron, A test which will locate positively the presence of fine surface defects in nonmagnetic castings is desirable.

Oldest of such tests is the one commonly known as the "oil and whiting test," in which a thin oil is applied to the surface of the casting and allowed to remain in contact for a time sufficient for the oil to be absorbed into any openings by capillary attraction.

If all oil is then removed from the surface and the part allowed to stand for a time, a slight amount of the oil will seep from the openings and become visible on the surface. To expedite the process, the cleaned casting should be painted immediately with whitewash which, when allowed to dry thoroughly, not only tends to absorb the oil from the opening (thus hastening seepage), but also shows this seepage to advantage by the discoloration of the white surface.

Instead of using whitewash, powdered talc may be dusted over the surface. This is not as sensitive as a thin coating painted on, but the casting is easier to clean afterward.

A commercial process has been developed for the purpose, utilizing a highly penetrating oil in which is dissolved a substance highly fluorescent under ultraviolet light. After sufficient time has been allowed for complete penetration and the oil has been cleaned from the surface, a so-called dusting powder is dusted or poured over the surface.

This powder absorbs the oil from

the defect much as ink is absorbed by a blotter. The casting is examined under black light (an invisible form of ultraviolet light), and the oil absorbed on the surface by the powder fluoresces or glows, giving a strong indication of the underlying defect.

Figure 13 shows a tear in a bronze casting as seen by visual examination. Figure 14 shows the same defect treated with fluorescent penetrating oil and developing powder and viewed under black light.

Like magnetic inspection methods, this latter method using a fluorescent penetrating oil shows even the finest surface defect in great detail. However, unlike magnetic inspection methods it does not give indications of interior defects unless these actually reach the surface, in which case the indication is limited to the exposed part of the defect. For interior defects, even if close to the surface, it is necessary to use radiography or some other method of nondestructive testing.

Supersonics. The use of high frequency sound waves (pitch above the audible range) is at this writing the latest development in methods of nondestructive testing. If supersonic vibrations are set up at one end of an object, the wave travels to the opposite end of the piece and is reflected back to the point of origin, the time of travel being measured by an ocillograph.

Principle

If a discontinuity lies in the path of the wave, this path, and hence the time of traverse, is shortened, the wave being reflected from the defect. The wave automatically plotted on the ocillograph will indicate the location of the discontinuity and may indicate its approximate size.

Supersonics have been developed to the point where they have been used with considerable success in examining forgings, rolled plate and other simple shapes. Because of the usually intricate shapes of castings and the comparatively rough surfaces, and also because the most likely defects occur at locations of compound curvatures, it has not been found practicable to apply this method to casting inspection beyond a limited extent.

In the near future the process and equipment will doubtless be developed to the point where an additional and valuable nondestructive testing method will be available.

It has been the intention of the writers to cover briefly the different methods of nondestructive testing as applied to the inspection of castings, and to discuss the comparative advantages and disadvantages of each.

Test Value

The value of any test depends upon the ability of the inspector, and this probably applies especially to those who have the obligation to accept or reject castings because of the results of nondestructive tests. The writers have seen cases where serviceable castings have been scrapped, and also where castings have been accepted only to fail later in service.

An inspector who has not had an exhaustive training and been able to prove his ability to interpret the results of tests correctly should not be permitted to pass final judgment upon the acceptability of a casting. Cases may be mentioned where newly hired personnel without manufacturing experience has been placed in inspection departments and, after a few hours' instruction, permitted to reject parts with finality.

It is the writers' opinion that a person cannot be a successful inspector until he has had thorough training in manufacturing processes as well as in the technique of inspection methods and the interpretation of tests.

Regardless of the inspector's ability in interpretation, he cannot do a satisfactory job of inspection unless he has been given standards of acceptance to which he can work and which will aid his judgment in making decisions. It is not to be expected that the inspector will do his work with constant reference to the standards.

Acceptability

Unless the foundry technique is poor or entirely lacking, no question will arise concerning the acceptability of good castings, and the same applies to those which are rejectable without question. Only in borderline cases will it be necessary to refer to the standards.

As further interpretation experience is gained, the standards will become more and more like a fire insurance policy, the cost of which is paid willingly with the hope that it never has to be used.

It is believed that the more important nondestructive test methods which are currently being used and are applicable to the examination of castings have been discussed in this paper. Detailed descriptions have been omitted as separate articles in the literature describing details of technique and the necessary equipment are available.

It must be borne in mind that, although an excellent job of inspection can be accomplished by the use of these methods, no method yet developed is sufficient in itself to determine with finality the presence and extent of all types of defects.

Tests for Defects

Rather, each method has its own advantages for certain types and locations of defects, and each method usually serves best when used in conjunction with one or more of the other tests. Constant development is underway with most of the methods to make them more sensitive, enlarge their field and shorten the time necessary for inspection.

Foundry Use of Nondestructive Inspection. Visual inspection is probably used in all foundries. However it has been the experience of the writers that many castings with small surface blemishes, which would in no way affect the serviceability, have been rejected because visual inspection failed to give any indication as to whether the discontinuity on the surface extended into the body of the casting.

This has been the case especially with pressure castings and with structural castings which might be subjected to severe loading and reversed stresses in service where an internal defect would be liable to progression as a result of these stresses. In such cases it becomes evident that a test which would give proof of internal soundness would introduce considerable savings in the foundry.

In spite of this possibility, many foundries have shown hesitancy in accepting these new methods of inspection, probably fearing that any method which would indicate the presence of defects not found by previously known methods could not fail to increase the percentage of rejected castings.

It is true that this has happened in a number of cases when radiog-

raphy and magnetic inspection were first used, as these methods began to show up defects of which the particular foundries were unaware.

However, when these same foundries began using the results of this inspection to determine the exact location of defective areas, then changing the foundry technique to eliminate these defective areas, it was found that most of the castings could be made without harmful defects. Also, by knowing the exact extent of all internal defects it became a simple matter to suggest design changes which would reduce the liability to unsoundness.

Use of radiography, for instance, has almost eliminated the necessity for sectioning or breaking castings to determine soundness. One steel foundry reported that the cost of castings which would have been cut up for inspection, had radiography not been available, had paid the cost of all radiographic inspection in that foundry, including the cost of equipment, materials and personnel.

It is probable that other foundries also have found this to be true. It is the writers' opinion that with the adoption of practical nondestructive testing methods and with a sincere effort to improve the quality of the castings, foundrymen will find that sound castings can be produced economically and the use of nondestructive testing a distinct asset instead of the liability which many have considered it.

Acknowledgment

Grateful acknowledgment is made to the Boston Naval Shipyard; the New York Naval Shipyard; the Naval Research Laboratory; Bath Iron Works Corp., and General Electric X-ray Corp. for the photographs and radiographs used in this paper.

DISCUSSION

Chairman: H. W. WARNER, Allis-Chalmers Mfg. Co., Milwaukee.

Co-Chairman: E. G. LEVERENZ, American Steel Foundries, East Chicago. Ind.

MEMBER: Suppose you section a casting and find no defects. You go into production and later find that there is a casting leak. What type of nondestructive test should be used? The defect is there. Sectioning did not reveal it.

MR. FREAR: As has already been stated in the paper there is no single type of nondestructive test which will show all defects which may be present in a casting. If the defect is a very fine crack, radiography might not show it, especially if the plane of the crack lies at an angle with the direction of the rays. If the crack is at or near the surface, however, it will be found easily by magnetic inspection.

With the possible exception of supersonics, there is no nondestructive test developed to date which will determine with certainty the presence of fine internal defects. It is hoped that some such method will be developed. Without knowing the type of defect, it is difficult to answer the question, but probably a combination of radiography, magnetic inspection and perhaps hydrostatic testing would serve to locate the defect.

The hydrostatic test would be more effective if applied twice, that is, by loading, removing the load, and then reloading. If the casting withstands all these tests the chances are that no further trouble will be experienced.

J. NABREZNY[1]: The author spoke of sounding of castings as being an obsolete method of inspection. In the malleable foundry industry, in conjunction with visual inspection, it is one of the best methods of determining cracked castings before the castings are annealed. Also, it is one method of determining whether a casting is properly annealed. It must be done by properly trained personnel. In some industries sound inspection is of little value because of the material being inspected.

MR. FREAR: We did not mean to convey the opinion that the use of percussion testing was obsolete, even though we did not like it in the majority of cases. Of course, in those cases where a defect can be found by this method, it is a good one to use, but most castings are of such shape that the presence of harmful defects would not be determined.

MR. NABREZNY: With visual inspection you can examine a casting carefully and find no cracks, but if you sound that same casting you will find a dead sound and by more careful inspection you will locate the crack. We have found castings that we have sounded and they sounded dead and by breaking up the casting we have located the failure. It is not 100 per cent reliable and even radiographic inspection is not 100 per cent, but I think it is one of the best ways in the malleable industry for segregating cracks in the hard iron.

MR. FREAR: We must admit that in preparing this paper we omitted this very important test on hard malleable iron, even though one of us has used it on hundreds of such castings. It is an excellent method for this purpose, and although not the only one which would serve the purpose, it is certainly the most economical. With respect to its use on annealed malleable castings, I believe the same reservations in its use are applicable as with other types of castings.

MEMBER: I believe Mr. Frear mentioned that magnetic inspection would show up defects in a casting within 3/8 in. of the surface. What kind of defects will that show up and will it show them up if the surface itself is sound?

MR. FREAR: It will show up almost any kind of defect but the finer the defect is the surer you are of finding it when it is below the surface. Of course, as a defect gets larger, or as it gets deeper under the surface, the pattern on the surface gets broader and it becomes more difficult to interpret. However, when the defects spread out so that interpretation becomes difficult; then they become easier to detect by radiography. This shows, I believe, how these two tests work hand in hand better than any others, and although neither by itself will determine the presence of all defects, the use of the two methods together gives us the nearest known approach to the ideal.

CHAIRMAN WARNER: A gentleman asked me about nondestructive inspection of castings, particularly on starting a production run. Sometimes the customer wants from 1,000 to 5,000 castings immediately. It has become the practice under the circumstances to check them visually with the result that you find yourself in a predicament when the customer in machining them runs into segregations along the line causing a leak or defective casting. He may find 50 per cent of these castings defective. Through either improper gating arrangements, solidification of the iron or steel, or pouring temperatures, he may get 50 per cent scrap out of one of the pieces. In this case the foundryman is definitely at fault because visual inspection did not insure sound castings.

This gentleman wanted to know whether it would not be advisable to test a trial run by magnetic powder inspection. It is a good idea. You do not have to make a complete examination of the castings but you can take one or two out of every 50 or 100 pieces and inspect by either magnetic powder or x-ray. In the particular job under discussion there was a defect about 3/8 in. below the surface of the casting. It would be difficult to pick it up through magnetic powder test unless the defect showed through minor outside surfaces. Then it would show up. Otherwise, on that particular job it would be necessary to use x-ray examination.

[1]Michigan Malleable Iron, Detroit, Michigan.

REPORT OF A.F.A. COMMITTEE

ALUMINUM ALLOY DIE CASTING

Although the fundamental information on die casting herein presented will be of specific interest to the producer and the consumer of die castings, it should be of general interest to everybody in the foundry industry.

Committee on Die Castings
Aluminum and Magnesium
Division, A.F.A.

THE THREE MAIN requirements for the commercial production of die castings are as follows:

1. A properly designed metal die to produce the desired part economically;
2. A machine with a means of injecting metal into the die at the proper speed and pressure and which will hold the die components in alignment under operating conditions; and
3. An alloy which has the proper characteristics for casting and which will give the desired physical properties to the casting.

Die Casting Dies

The design and construction of the die casting die is extremely important to the success of the satisfactory commercial production of die castings. The dies for die casting aluminum are made from alloy steels nominally consisting of 5 per cent chromium, 1 per cent molybdenum, and 0.4 per cent carbon. This steel is heat treated after machining and will be satisfactory for making 50,000 to 500,000 castings or more, depending on the size, shape, etc., of the casting.

Heat Treatment of Dies. The dies are heat treated by placing the blocks comprising the die in a heat treating box with a few pieces of charcoal, or by imbedding the blocks in cast-iron chips. In either case, the box is covered, the joint sealed to exclude air, and the container and contents placed in a muffle furnace where they are heated at a rate of about 100° F. per hour until a temperature of 1200° F. is attained. They are held for two hours at 1200° F. to equalize the temperature of the box and contents before increasing the temperature at the same heating rate to 1600° F., at which temperature they are held for an additional two hours. The temperature is increased at the same heating rate to 1800-1850° F., at which temperature the dies are held for one hour per inch of thickness plus one hour.

The die blocks are removed as quickly as possible from the container and allowed to cool in still air to about 300-500° F., so as to minimize distortion. They are then tempered at 1050° F., for six hours per inch of thickness, removed from the furnace and cooled in air. Hardness should be between 416 and 460 Brinell (444 desired) with a 3000-kg. load. Although little scaling takes place, that which occurs can be removed by descaling process.

It is the practice of some die casting plants to produce a limited number of sample castings from a die before heat treatment, but most plants discourage this practice because, in the running of sample castings, the slides and cores are apt to gall, and the alloy cast may "solder" (alloy) to the soft die walls. Heat checking of the die also is accelerated by this practice.

The die steel can be nitrided after heat treatment to produce a harder surface by treating at 950-1050° F. in a closed container with dried ammonia gas.

The die itself consists of two parts, namely, a cover die and an ejector die, which define the surfaces of the desired part. The addition of an ejector plate and ejector pins behind the ejector die provides a means for removing the casting from the die. Such a die is shown in Figs. 1 to 4.

Interior openings are formed in die castings, as in sand castings, by the use of cores, and reduction of machining cost thus is obtainable as might be expected. Since any but parallel cores would interfere with the removal of the casting from the die, cores must be provided with means for removal from the casting before they interfere with either the opening of the machine or the ejection of the casting.

Casting Ejection Mechanism

A die casting is designed to stick to the ejector die and is removed with the aid of pins (called ejector pins) which are held by the ejector plate. The plate can be moved by a rack and pinion, which can be operated manually, hydraulically, or by means of a linkage to a part of the casting machine. Direct connection to hydraulic cylinders also is used, as are various cam arrangements.

Ejection can also be accomplished by means of a bar on the casting machine, which stops the ejector plate as the die opens and so receives its name "bump ejection." A more recent method involves the use of tapered blocks in which die cavities are cut. When these tapered blocks are moved for-

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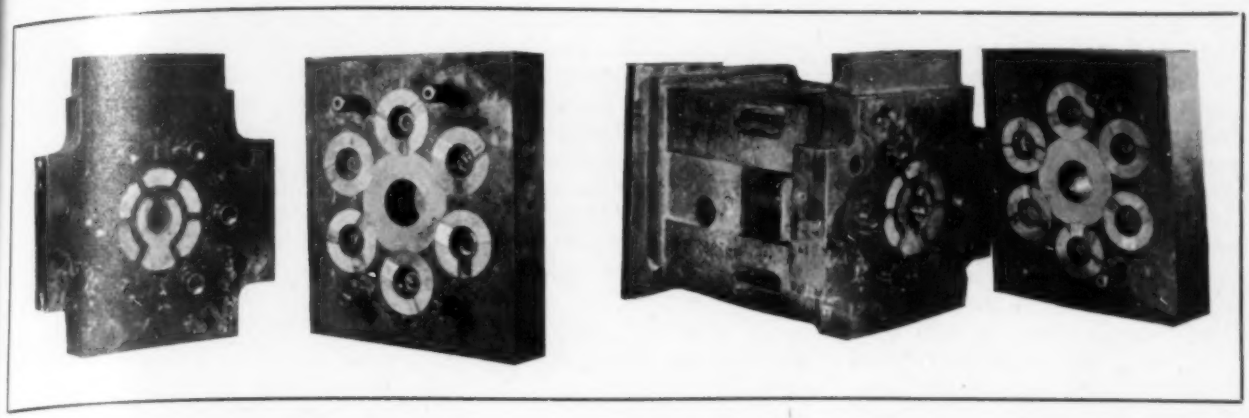


Fig. 1 (left)—Cover and ejector die blocks. Fig. 2 (right)—Ejector die block mounted on die base with ejector pin plate assembled and cover die block.

ward by ejectors, they carry the casting with them and open up in proportion to the taper. By this method some external undercuts can be cast direct, and the undercut will clear the die as it opens so that the casting can be removed.

The ejector plate usually contains several large pins, ground so as to be flush with the face of the die when the ejector pins are in the casting position. When the die is closed, these so-called "surface pins" enable the ejector pins to retract without being bent or broken by contact with the cover die, and remain always in position to eject.

Special properties which are not characteristic of the cast alloy frequently are required at certain points in a casting. Under such requirements the die and the part to be die cast are designed so that previously fabricated parts (termed inserts) having the desired properties are anchored in the casting by the cast metal (Fig. 5). Inserts provide greater wear resistance, strength, hardness, conductivity, bearing, magnetic and spring properties, and simplify design by joining several castings into a unit when coring and undercuts make such devices necessary.

These inserts are usually keyed, knurled, or undercut in such a way that they will not turn or pull out of the casting, but can be designed so that some relative motion between the insert and the casting will take place. They may be metallic or non-metallic, iron, steel, brass, bronze, graphited bronze (oilless bearing), phenolic and other type resins, and even wooden inserts have been used. Inserts may be produced as screw machine products, stampings, forgings, etc.

Dies designed for one impression only are called single dies; if for more than one impression of the same part, they are called multiple dies, which may yield, in some cases, as many as forty impressions. Dies can give impressions of different parts, so that all the castings required for one assembly may be made in one shot. Such dies are called combination dies, which can be made so that some of the impression blocks can be replaced by others. The latter are referred to as interchangeable dies.

Holding dies, made by machining cavities in the large die blocks into which standard small dies for various parts can be inserted, are a variation of the interchangeable die and are known as unit dies. Holding dies will hold six unit dies at one time, and each of which, depending on the size of the casting to be made, may yield impressions for several parts.

The unit die method has made it possible to produce small parts economically by dividing among six dies the labor required to make each shot. Obviously, the same alloy should be specified for the parts to be run simultaneously. Where dimensions must be held to the lowest possible tolerances, single dies must be used.

Die Casting Machines

The die casting machine can be considered as consisting of two major parts, namely, one to hold the die, and the other to hold and/or inject metal into the die cavity. Basically, the die supporting part of the machine is the same, regardless of the method used for metal injection. It consists of a rigid frame and a means for closing the die, locking it under pressure, and later opening it to allow the casting to be removed. The schematic drawing of a die casting machine shown in Fig. 6 is self-explanatory.

Machine Frame. The frame of the machine must be rigid enough to support without deflection the weight of dies which may be placed on it. It must be adequately massive so that the pressure exerted on the injected metal does not stretch the structural bars and allow the die to open during the "shot." This condition would not only make it difficult to maintain the dimensions of the casting across the die opening, also, because of the loss of pressure on the metal, would lead to decreased density of the casting. The liquid metal which would be sprayed from the die in such a case would be dangerous not only to the operator of the machine but to others in the vicinity.

Although machines may be operated both vertically and horizontally, most modern machines are built for operation in a horizontal position. This feature makes it easier to change dies.

Die Actuating Mechanism. The devices for opening and closing the die are usually hydraulic cylinders directly acting on the movable plate, or in combination with a mechanical linkage or toggle. The pressures required to hold the die together are 600 tons in many cases, and probably sometimes more.

There are two methods of injecting metal in general use. They are the "gooseneck" method and the "cold chamber" method.

"Gooseneck" Method. The older

or "gooseneck" method involves metal injection by the application of air directly on the surface of the liquid metal in a container called a "gooseneck." The "gooseneck" is supported in the molten bath and is refilled after each shot by dipping the nozzle under the surface of the bath.

The steps in the casting cycle using the gooseneck machines are as follows:

1. Close the die, then raise the gooseneck from the "at rest" position in the bath to the "casting" position, with the gooseneck nozzle held firmly in the bushing at the pouring gate.
2. Make shot by applying air pressure on the top of the metal in the gooseneck, forcing liquid metal through the gate into the cavity.
3. Maintain this pressure until the casting has solidified.
4. Lower the gooseneck to starting position.
5. Open the die and eject the casting by moving the ejector pins forward.

The process is repeated after retracting the ejector pins. A simple sketch of the "gooseneck" is shown in Figs. 7 and 8.

Objections to Gooseneck Machines. This type of die casting machine has several objections which have resulted in constantly diminishing use. Since the gooseneck and metal pot are made from cast iron and the gooseneck moves up and down in the bath at each shot, the molten aluminum alloy dissolves iron from these parts and it becomes almost impossible to control the iron content in the casting. The iron content by specification is usually allowed to rise to 2 per cent in castings made by this process, which does not give the best properties to the metal.

It has not proved practical to use air pressures over 500 to 700 psi. in this process, and the castings are not as dense as those made by the newer "cold chamber" process with its higher pressures.

The gooseneck machine is also more dangerous to operate than the cold chamber machine, because the metal may squirt out between the nozzle and the die. The metal, as it enters the pouring gate, is sprayed into the cavity and these oxide-covered sprayed particles trap air from the gooseneck, nozzle, gate and the die cavity in the casting.

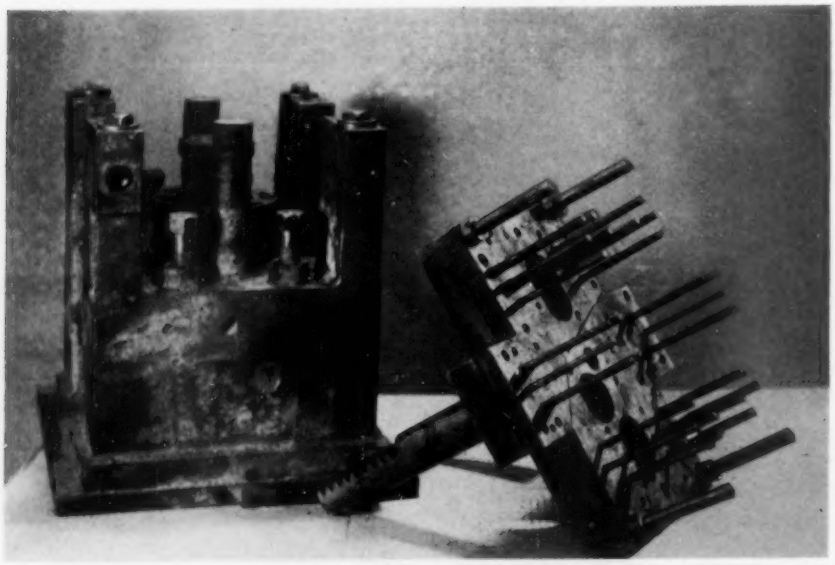


Fig. 3—Die base (left) and ejector pin plate assembly.

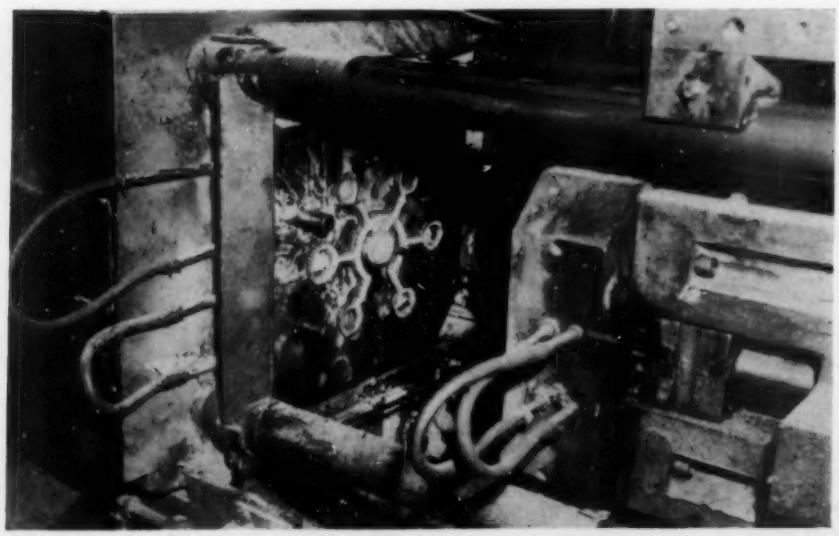


Fig. 4—Die mounted on machine.

"Cold Chamber" Method. The "cold chamber" method is now the process in most general use. In this method, a piston operated by a hydraulic cylinder applies pressure on the metal in a chamber which is charged by hand ladling. This mechanism is used both vertically and horizontally, although most machines use the horizontal type.

The steps in the cold chamber casting cycle are as follows:

1. Close the die.
2. A ladle of the proper size, which will furnish just enough metal to fill the cavity, gates and overflows, and leave a plug in the chamber about ¾ in. to 1 in. thick is used to ladle metal from the holding pot to the chamber.
3. The piston is moved forward by the hydraulic cylinder, with the speed controlled by a valve set to fit the conditions, and forces the metal in the chamber through the gate into the die.
4. When the metal has solidified, the die is opened, and since the pressure is still being exerted by the hydraulic cylinder, the plunger pushes out the slug remaining in the chamber and is then withdrawn to its starting point. Cores are withdrawn and the casting ejected.

Objections to Cold Chamber

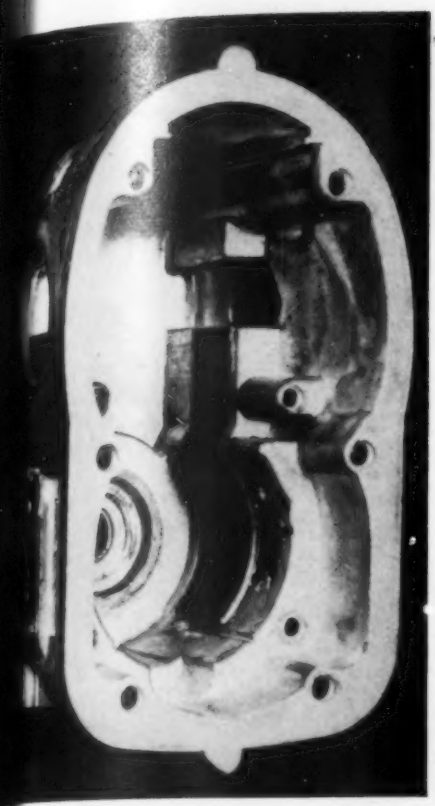


Fig. 5—Electrical motor housing showing inserts.

Machines. The objection to the horizontal cold chamber is that the shot sleeve is only partially filled with metal, which starts to solidify along the walls of the sleeve as soon as it is poured. The advancing plunger gathers up liquid and solidified metal, and the air above it, and forces them all into the die. Practical considerations, such as speed of operation, have caused the horizontal chamber to be favored over the vertical. The vertical sleeve does not have the objections just cited, but the plunger tends to stick and the process is slower.

Advantages of Cold Chamber Method. Since the hot metal is in contact with the walls of this cylinder for only a short time and is ladled out of a pot which can be coated or enameled to prevent iron pick up, the iron content of the casting alloy can be controlled to a much lower figure than in the gooseneck machine. Iron contamination can, if necessary, be kept at 0.8 per cent or below. Castings made from alloy of low iron content solder to the die in many cases, and for this reason most specifications allow 1.3 per cent Fe, which causes only a slight reduction in physical properties of the casting.

Injection plunger speeds may vary from 250 in. per sec. to 5 in. per sec. At the higher figure, the metal is sprayed into the die, while at lower speeds the metal feed changes to a flowing stream. Larger pouring gates are required for slow plunger speeds. Slower speeds are used only for heavy-walled castings.

In other words, the volume of metal entering the die per unit of time, determined by the gate cross-section and the pressure on the metal, is the deciding factor. However, it frequently is impossible in thin-walled castings to use a gate sufficiently large to fill the cavity properly at low speeds if the gate must feed into too thin a section.

The pressures involved in this method vary from 2500 psi. on the metal to as high as 100,000 psi., although most work is done somewhat below 20,000. Denser and stronger castings can be produced by the cold chamber method than can be produced on the gooseneck machine.

A die casting alloy should have a short freezing range to minimize cracking of the casting in the die during solidification. It should have a composition low in contaminating residual impurities, and of course should be gas free and dross free. In order to achieve these purposes, it is necessary to start with metal of the desired purity, and then control the process so that the metal in the finished product is unchanged

Most die casting concerns now purchase their metal already alloyed. However, all incoming metal should be carefully analyzed to make certain that it meets the specifications under which it was purchased.

Whenever the hourly requirements for molten alloy are high—for example, 1000 lb. per hr., melting can best be accomplished in refractory-lined reverberatory type furnaces. Although they usually are fired by gas or oil, in special cases coal has been used as a fuel. No iron is picked up by the molten metal. The large surface of this type furnace aids in securing contact between the molten metal and the flux, and makes drossing more complete. The movement of the metal through the furnace is slow, compounds present will be deposited as a sludge in the bottom of the furnace and later drawn off.

The metal drawn out of the reverberatory usually is delivered in pre-heated bull ladles directly to the pots at the casting machines. If molten metal is delivered in this fashion, the temperature control of the metal in the pots can be maintained within closer limits than if ingots or solid scrap are added directly to the pots.

If the demands of the casting room are too low to justify the use of a reverberatory furnace, other methods of melting can be substituted. These range from horizontal cylindrical furnaces holding perhaps 1500 lb., through tilting furnaces holding 500 lb., to melting furnaces placed at each machine, where the metal is melted and ladled into the holding pot at the machine as required.

Metal Temperature Control

Both the melting furnace and the holding furnace should be temperature controlled to avoid overheating of the metal and to maintain the temperature found by experiment to produce the best castings. Automatic control has been found superior to manual control of temperatures.

The holding furnaces usually are gas-fired, designed to allow holding 250 to 300 lb. of molten aluminum at the machine. Either cast iron pots (which can be coated with enamel or other materials to reduce the iron pick-up) or refractory pots are employed, the latter being more brittle than the cast iron but not soluble in the aluminum alloy.

In the die casting industry, gates and sprues represent a considerable percentage of the weight of an untrimmed casting. Such scrap, plus reject castings, represents 50 to 75 per cent of the metal melted.

Reverberatories are operated by charging as much metal into the melting chamber as will sink into the bath and by repeating this process as soon as a charge has melted, until the furnace is full. Usually a ratio of two or three parts of gates and other scrap is melted with one part of ingots in each charge.

A sample for analysis is taken and is analyzed by spectrograph, a complete analysis of a heat being made in 20 to 30 minutes. During this period, if much dross is on the surface, it is fluxed with mixtures of

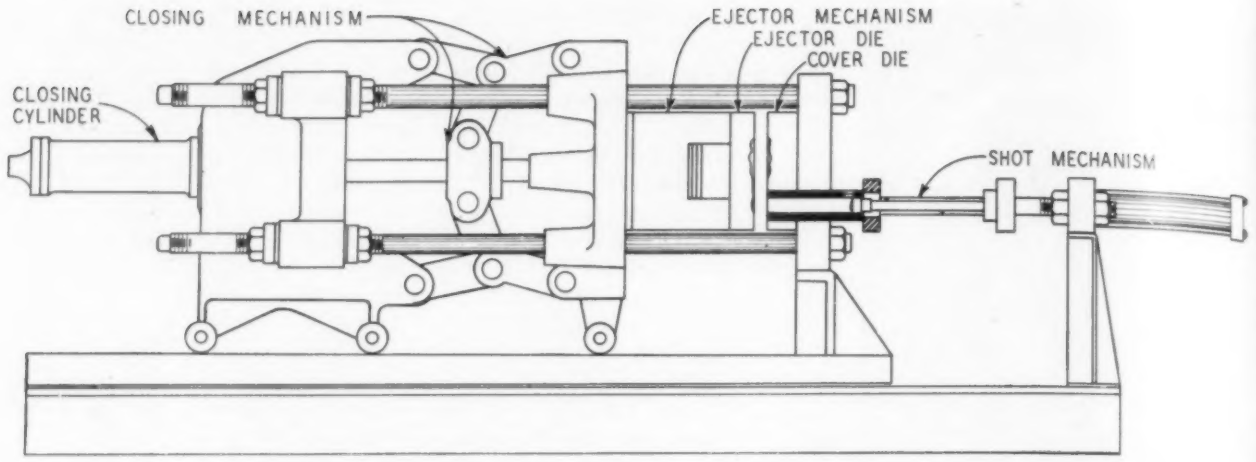


Fig. 6—Schematic drawing of die casting machine.

salts such as sodium fluoride, potassium fluoride, sodium chloride, etc., which can be mixed in the plant or purchased as a proprietary flux. The dross should be cooled as quickly as possible to prevent oxidization of the contained aluminum, which usually is sold to a secondary aluminum dealer for recovery.

Metal Analysis Control

When the analysis has been received, metal composition is corrected and the metal taken by bull ladles to fill holding pots as needed. As much as 4000 lb. can be taken from a reverberatory holding 7000 to 8000 lb. before recharging. In the meantime, if metal demand is great, other reverberatories are run in overlapping cycles.

A reverberatory can be continuously charged at one end and discharged at the other, but control of the metal analysis is less exact with the batch method of melting.

Inasmuch as most foundries use more than one alloy, it is desirable to analyze the holding pots regularly to avoid mixing of alloys. The high magnesium type alloys usually are melted in small melting furnaces, as are alloys of high purity where the quantities desired are low. It is also desirable to flux or skim the holding pots several times a day to remove the dross, which may cause machining difficulties in castings.

Aluminum Base Alloy Properties. The outstanding characteristics of aluminum die castings are:

1. Low specific gravity.
2. Corrosion resistance.
3. Retention of luster for longer periods of time after polishing.
4. Freedom from dimensional changes.
5. Retention of properties at subnormal temperatures.
6. High thermal and electrical conductivity.

This combination of properties has resulted in wide use of aluminum die castings. The average specific gravity of aluminum die castings is 2.80, as compared with 1.80 for magnesium base alloys. For many applications, especially small parts, this weight differential is negligible.

The ability of a part to take and hold a polished luster is most important, and advantage is taken of this property in the finishing of many die castings.

Aluminum die castings have excellent corrosion resistance in the as-cast state. When it is desired to increase resistance to attack by corrosive media, the anodizing treatment may be resorted to, since it greatly enhances corrosive resistance.

Aluminum is one material capable of withstanding extremely low temperatures without affecting original ductility. For this reason, aluminum die castings have found extensive use in aircraft refrigeration and similar applications where temperatures of —60° F. or lower are encountered.

Many aluminum alloys have been used for die castings, but most die casters now use three main types, as shown in Table 1.

Silicon-Aluminum Type Alloys. Alcoa alloy 360 is a high strength alloy which probably will replace Alcoa 13 (ASTM S9). The former does not show the segregations of primary silicon in the center of heavy sections when cast on certain type casting machines. These segregations, known as eutectic chill flakes, are frequently found in castings of the 12 per cent silicon alloy. Either of these alloys is used when a fluid alloy is required to fill complicated and thin-walled shapes, where good corrosion resistance is required.

Difficult to Cast

A special grade of the 12 per cent silicon alloy (Alcoa 13x), containing a maximum of 0.8 per cent iron, is somewhat more difficult to cast than the standard Alcoa 13 (ASTM S9), but produces castings with higher ductility, shock resistance and corrosion resistance.

Copper-Silicon Type Alloy. This alloy is an outgrowth of World War II. It is based on the use of aircraft scrap, most of which has a 4 per cent copper content. By eliminating most of the magnesium in the scrap and adding up to 9.5 per cent silicon, an aluminum alloy is obtained which has been found quite satisfactory as a die casting alloy. It possesses good casting characteristics and has excellent physical properties.

This alloy was used almost exclusively for aluminum ammunition component castings and other uses, to such an extent that it can be considered a general purpose aluminum die-casting alloy. In the author's opinion, this alloy may become the general die casting alloy for post-war use, because of low cost.

Magnesium-Aluminum Type Alloy. The ASTM G2 alloy is the best

alloy where high impact resistance, high ductility and optimum corrosion resistance are desired. It takes and retains a beautiful luster when polished, and has optimum machinability properties. However, it is more difficult to cast than the silicon type.

Special properties undoubtedly can be obtained by casting other alloys than those listed in Table 1. However, in an effort to prevent contamination, every foundry endeavors to standardize on those alloys which, in the main, will be satisfactory for most of the castings required. Any castable alloy can be die cast by setting up separate melting pots, etc., but this should be avoided whenever possible.

Some special die castings have been produced which can be both solution treated and precipitation hardened, although die castings as a class cannot be given a solution heat treatment because they will blister. However, a low-temperature aging will result in enhanced properties if the alloy is normally improved by complete heat treatment.

Since die castings are made in steel dies, it is obvious that the shape of the casting must be such that it can be withdrawn readily from the die. This is accomplished by constructing the die without undercuts (except as noted below), and with proper tapers on all surfaces.

Tapers. All external surfaces should be designed with a continuous taper, maximum at the parting line of the die. It should be as large as possible, with a minimum allowable taper of 0.010 in. per in., except for small depths where, if required, straight sides can be cast. The minimum taper that should be designed for cores is 0.015 in. per in., and the largest dimension must be at the base of the core. The reason for the larger draft allowance on internal surfaces is readily understood when the large thermal contraction of aluminum and its alloys is considered.

Undercuts. Internal undercuts can be cast only by the use of collapsible cores. Such undercuts slow the casting cycle, cannot be controlled as accurately as can sections of solid steel, and add to die cost.

External undercuts are produced by loose pieces in the die, which are ejected with the casting and must be removed and replaced in the die for another shot by the use of slides, or by "angular ejection." The use of these methods may lead to excessive "flash" on the casting which must be removed, and may result in reduction of dimension accuracy, higher die cost, and slower speed of production.

Redesigning

Therefore, before planning to cast undercuts, it is always wise to consider redesigning the part so that it will accomplish its purpose as a simpler casting. However, in many cases, undercuts should be used and will show savings.

Cores. Cores should be designed to provide the highest possible ratio of diameter or thickness to length. Long, thin cores should be avoided, as they have a tendency to burn out and bend or break, thus increasing die maintenance. Through cores, with both ends supported, should be used for small holes wherever possible. Interlocking cores may increase die maintenance costs through breakage.

Parts should not be designed with openings completely through the casting, unless required. The flash that will appear around the opening at the far end will require an extra cleaning operation. Where possible, cores should be used to maintain uniform section thickness and for reduction of weight. They should also be used wherever openings are to be drilled or tapped, since the density of metal is greater in heavy sections around cores than in uncored heavy bosses, where shrinkage voids tend to appear.

Cores can be used to produce openings of shapes that would be difficult if not impossible to machine. Very small openings are not cored, since the cores would require frequent replacement, and such openings usually can be drilled more economically. No opening less than 3/32 in. diameter should be cast, and the length of the core should not exceed four times the diameter. Cores of ½ in-1 in. diameter may have an unsupported length of six to eight times the diameter of the opening.

Bosses. Bosses are frequently provided on die castings and should be cored to provide metal of sufficient density for tapping. Outside threads should not be used on bosses, but where outside threads are required,

Table 1

Composition and Physical Properties of Aluminum Die Casting Alloys

	Alloy Type			
	Silicon-Aluminum		*Copper-Silicon-Aluminum*	*Magnesium-Aluminum*
Designation				
ASTM	S9	—	SC6	G2
Alcoa	13	360	A380	218
Composition (per cent)[1]				
Copper	0.6	0.6	3.0–4.0	0.2
Silicon	11.0–13.0	9.0–10.0	7.5–9.5	0.3
Iron	1.3	1.3	1.3	1.8
Magnesium	0.1	0.4–0.6	0.1	7.5–8.5
Manganese	0.3	0.3	0.5	0.3
Zinc	0.5	0.5	0.6	0.1
Nickel	0.5	0.5	0.5	0.1
Tin	0.1	0.1	0.3	0.1
Total other impurities	0.2	0.2	0.5	0.2
Properties[2]				
Tensile strength, psi	40,000	42,000	41,000	42,000
Yield strength, psi. (0.2 per cent set)	18,000	23,000	23,000	23,000
Elongation, per cent in 2-in.	2.0	2.0–4.0	3.0	5.0
Charpy impact, ft. lb.[3]	2.0	2.0–5.0	3.5	10.0
Specific gravity	2.66	2.68	2.76	2.53
Weight lb./cu. in.	0.096	0.097	0.099	0.091
Melting point, °F	1065	1110	1090	1160

[1] All percentages are maximum unless a range is shown.
[2] Typical physical properties on die cast aluminum test bars. These properties do not exactly represent the properties of castings but are used as a guide for engineers in design.
[3] Impact determined on ¼-in. x ¼-in. bars.

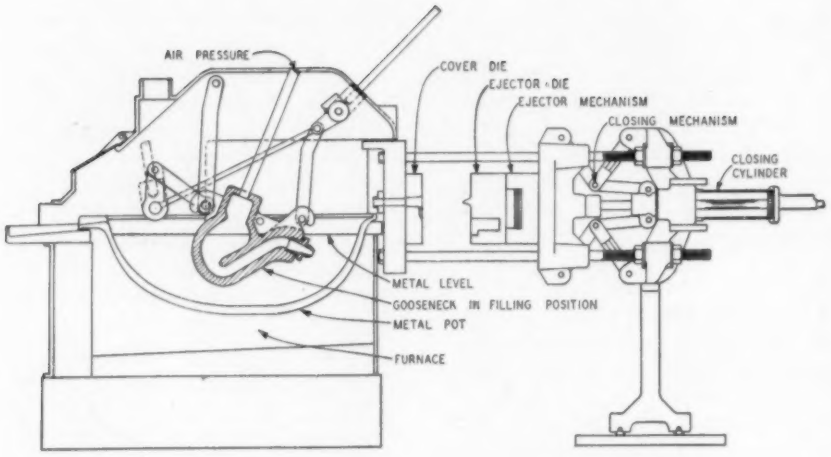


Fig. 7—Aluminum gooseneck machine in open position.

an insert of some other material should be used. Speed nuts should be employed if possible.

Section Thickness. A casting must have a minimum thickness of 0.060 in. to cast well in small areas, and as much as 0.090 in. or more on long, flat sections. The usual requirement is that the thickness be one-thousandth of the area cast. Section thickness should be as uniform as possible, and where changes in section occur they should be made as gradual as possible by means of fillets. Fillets should be used at all changes of section and should be as large as practicable.

Too great a section thickness should be avoided because, the thicker the section, the greater the tendency toward porosity. Contrariwise, the thinner the section, within the limits of casting practice, the less the porosity.

In order to avoid heavy sections, ribs are used to provide greater strength and stiffness than would be afforded by a flat, thin section. Shadow marks on the flat side over a rib may be avoided if the thickness of the section is great enough, perhaps 0.100 in. or if the rib width is kept below 70-80 per cent of the section thickness.

Threads. Threads are not ordinarily cast on aluminum die castings. However, external threads have been cast by the angular ejection method on millions of shell parts during the war—with a coarse pitch of about 12 threads per inch. Internal threads usually are made by drilling and tapping a cored hole. Cut external threads are not used in aluminum since they are stress raisers, hence not desirable.

Tolerances. The basic advantage of die castings over other aluminum castings is that a die casting is cast to the finished size in most cases, with little machining required in other cases. The surface is smooth, and openings of any shape can be cast to precise dimensions and to exact spacing.

The dimensional variation between castings from the same die is low. For example, tolerances on those parts of a casting made in the solid die (exclusive of dimensions across the parting line, or across lines where slides join the die) are only 0.0015 in. per in., with a minimum tolerance of 0.003 in., although the required tolerance across the parting line is greater.

Where machining to close tolerances is necessary, a stabilizing treatment may be given the castings. This consists in heating the castings to 450° F., and holding at this temperature for two to five hours. The casting and solidification strains are relieved, and practically no further change in dimensions will be experienced.

Since a die casting is made from rapidly chilled metal, the grain size is small and physical properties of test bars as die cast are higher than test bars from the same alloy as-cast in sand or gravity permanent molds.

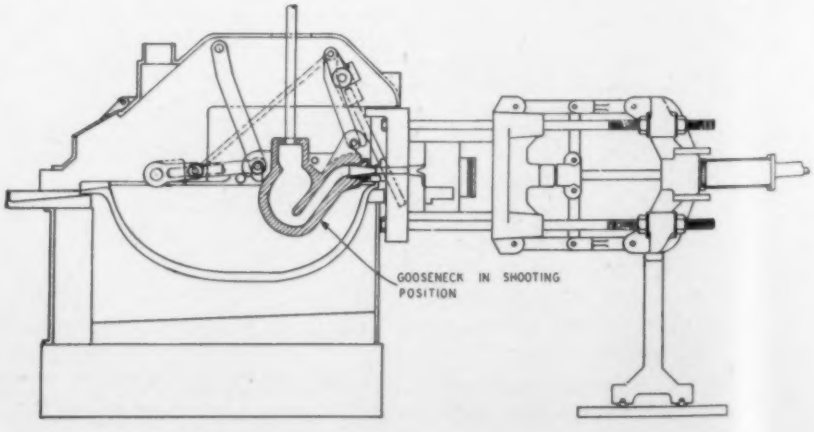
Size of Die Castings. At present, aluminum die castings do not usually exceed 15 lb. in weight, although there is no practical reason why castings considerably larger could not be made if required. However, the dies would be much more expensive, and larger machines might be required than those now in general use.

Trimming Die Castings

When a die casting comes from the machine, it consists of the casting or group of castings connected by gates to the sprue and surrounded by overflows, flash and ejector pin pads. These must be removed before the casting is used. Unless the gate is very heavy, or the casting quite fragile, the gate is broken off before removing other excess metal. When the gate cannot be broken off, it is either sawed or cut off in a chopper.

Whenever possible, the balance of the group is placed in a trimming press, operated either by foot or mechanically, and all excess metal around the parting line is removed simultaneously. Sometimes it is necessary to separate the castings made at one time and to trim them

Fig. 8—Aluminum gooseneck machine in closing position ready to make shot.



individually. Hand filing is avoided, except in cases where trimming either cannot be done or can be done only incompletely. Small lots are usually hand filed. Some surfaces require grinding on disc grinders or sanding belts.

Inspection

Before using aluminum die castings, and in the process of casting, various means of inspection are employed, depending on the end use of the part. The only way of non-destructively checking the internal condition of the casting is by means of x-ray, either on film or by fluoroscopic screen.

Process Inspection. Die casting producers utilize their radiographic equipment during the initial stages of production in order to correct gating, venting, overflows, etc., until the die is producing acceptable quality castings. By means of regular checking, quality is maintained. The castings may also be subjected to percentage checks up to 100 per cent, if the application requires it.

Fluoroscopic inspection is less expensive than film radiography, but is less sensitive, in the author's opinion, since it is only capable of detecting defects three to five times as large as can be detected by film radiography. In all cases, the end use of the casting determines the extent of inspection required.

End Use Tests. Castings are designed for use where definite mechanical properties are required. Tests subjecting castings to loads equal to, or greater than, service requirements are applied, insuring satisfactory performance.

Castings frequently are required for pressure applications. In such cases, they are pressure tested by applying gas or hydraulic pressure on specific sections.

Aluminum Die Casting Applications

The applications of aluminum die castings extend to practically all industries. A few examples of their use by industry follow:

Household Utilities. Vacuum sweepers, washing machines, ironers, refrigerators, food mixing devices, cooking utensils, waffle molds.

Electrical Industry. Motor housings, motor end shields, generator housings, instrument cases and frames, telephone, radio, radar and television parts, phonograph parts and record changers.

Optical Industry. Cameras, projectors, binoculars, microscopes.

Aircraft. Various starter motor parts, conduit boxes, airscoop, carburetors, hardware.

Business Machines. Typewriters, adding and tabulating machines, stenotype cases.

Automotive Industry. Trim, carburetors, generator parts, brake shoes and hydraulic brake parts.

Marine Industry; Portable Tools.

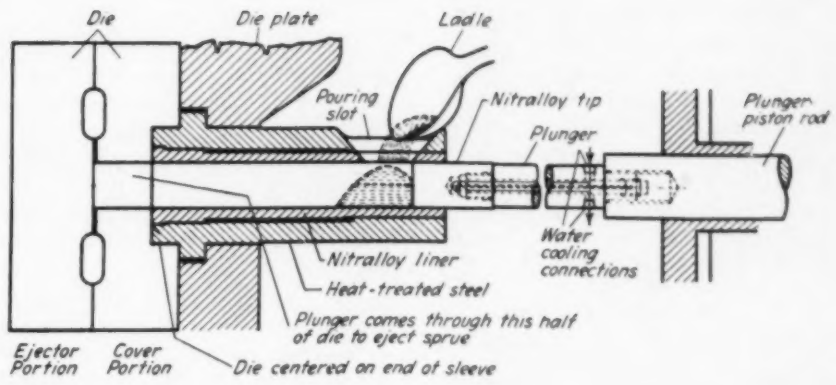


Fig. 9—Sectional diagram showing the die, cold chamber, and horizontal ram or plunger (in charging position) of a modern cold chamber type of die casting machine.

Die Cast Surfaces

Surface as Cast. A die casting has a very smooth surface as cast, but a large flat area is difficult to polish to a brilliant luster, hence the area usually is designed with a raised surface. A depressed design on the surface results in high die costs, since it requires cutting away die surface to leave raised design.

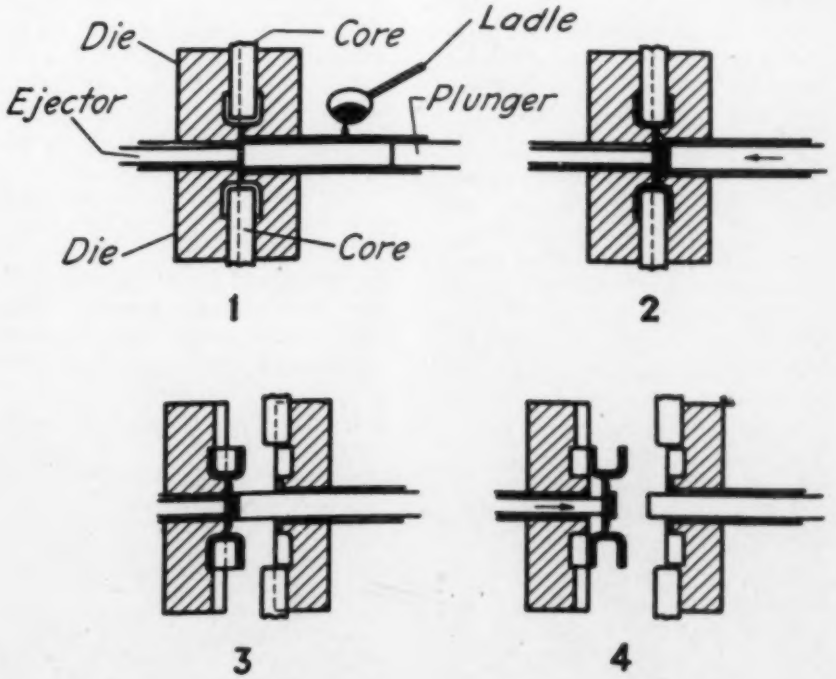


Fig. 10—Diagram illustrating four stages in the production of cup shaped die castings in a cold chamber machine having a horizontal plunger: 1—Ladling metal into cold chamber. 2—Molten metal after it is forced into die. 3—Cores withdrawn, die opened and plunger advanced to eject casting from fixed half of die. 4—Plunger withdrawn and ejector advanced to force gate of casting from ejector half of die.

Where a number or design is required below the surface, it usually is obtained by placing the design on a standing pad, thus forming raised letters on the casting below or flush with the surrounding surface.

Ejector pin marks are undesirable therefore are frequently placed on pads which, attached to the casting, can be broken off in trimming.

As-Cast Surface Standards

No standards are available for surface finish in the die casting industry. However, the notes below may prove helpful:

Decorative Finish. This finish represents the ultimate in casting surface finish and can be attained with a 220-grit grease wheel at the parting line only, followed by buffing of the entire part. Parting line must be evenly matched and solid; shrinkage areas at parting line are *not* acceptable. Solder and waves are *not* acceptable.

Service Decorative Finish. This finish is required on parts from which service as well as ornamentation are expected. In addition to the 220-grit grease wheel at the parting line, spot polishing with the same wheel is anticipated. Slight solder, waves and small surface run marks are allowed. Parting line must be evenly matched, and shrinkage spots at parting line held to a minimum. Slight porosity at gate is acceptable.

Commercial Die Cast Finish. Polishing of parting lines and spot polishing to start with a 180-grit grease wheel. Light solder, waves, run marks and shallow surface porosity are acceptable. Slight porosity at gate and parting line, and slight shrinks at parting line, are acceptable for this grade.

Mechanical Finish. A finish specified where solidity and strength are the prime requisites. Porosity *cannot* be accepted, but slight shrink spots and waves may be acceptable. Parting lines should clean up in die trimming or should be filed.

Paint Finish. Finish specified where a ground coat is applied before the finish coat. Surface imperfections such as run marks, cold shuts, slight surface porosity and small shrink spots along the parting line are acceptable. The parting line should clean up when ground with No. 150 emery.

Finishes for Die Castings

Aluminum die castings are covered with an oxide film which constitutes an important advantage from a service standpoint. For numerous applications, the ability of aluminum alloys to form and maintain this natural film is sufficient protection, and further finishing is not necessary.

For many other applications, however, where a more decorative appearance or better resistance to service conditions may be desirable, a wide assortment of durable and attractive finishes is available. These finishes are classified as follows:

Mechanical Finishes developed by abrading or working the die-cast surface by mechanical means to obtain a polished or textured effect.

Chemical Finishes, produced by treating the castings with chemical solutions to give etched surfaces or oxide coatings having various properties for specific application.

Anodic Finishes, formed by anodically oxidizing the castings in a suitable electrolyte to form a substantial oxide coating, which may then be given supplementary treatments to seal or color it.

Electroplated Finishes, applied by the electro-deposition of metals, such as nickel or chromium, on the aluminum surface.

Paint and Organic Finishes, which are clear and pigmented coatings applied by spraying, brushing or dipping.

The methods used or combined, in a commercial finishing procedure for any die casting, will depend upon sales and service requirements, cost considerations, design of casting (size and shape) and alloy type. Where no background of experience exists, it is advisable for the user of aluminum die castings to consult with the die caster or an experienced jobbing finisher.

DISCUSSION

Chairman: R. E. WARD, Eclipse Pioneer Div., Bendix Aviation Corp., Teterboro, N. J.

Co-Chairman: A. T. RUPPE, Bendix Products Corp., South Bend, Ind.

R. C. CORNELL[1]: I would like to ask Mr. Fox, if he were to buy a die casting machine with his own funds and put it in his own plant, what aluminum-base alloy would he use?

MR. FOX: It would depend on my economic status. I would probably use the cheapest I could get hold of, in which case it would be the 4 per cent copper, 9 per cent silicon alloy.

MR. CORNELL: Suppose someone else was buying your alloy? Which would you use? Suppose you were not concerned with what it cost, but you wanted the alloy which would be the easiest to cast from the mechanical standpoint.

MR. FOX: Generally speaking, the 12 per cent silicon alloy is a general all around alloy. We are now investigating a modification of this type alloy containing 10 per cent silicon and 0.5 per cent magnesium for having superior properties than the straight 12 per cent silicon alloy.

A. SUGAR[2]: The 12 per cent silicon alloy would be the easiest from a casting standpoint, but it would not be the best as far as machinability is concerned. Is that right?

MR. FOX: Yes, but it depends on your machining set-up.

MEMBER: Mr. Fox mentioned that the vertical type injection system on a cold chamber die casting machine was not good because it was too slow. What causes it to be too slow?

MR. FOX: It requires two pistons and you have your attendant difficulties with pistons. It is a two-man operated machine, known as the Polak type machine. We like it because it is capable of holding the molten alloy in a definite place and you can fill up all cylinder areas. You have no air space, whereas if you pour into a horizontal cylinder you have air space to contend with. Until it can be made economical and there is less maintenance on it than on the horizontal machine, the horizontal machine is generally more adaptable.

J. J. MCCLAIN[3]: I would like to ask Mr. Fox if he has been able to make successfully and consistently any one type of die casting which can be heat treated.

MR. FOX: We have made and Mr. Cornell of Dow Chemical has made a casting which was radiographically sound. We never tried to heat treat it. That, of course, is the way to determine maximum soundness. By a prescribed control of the speed of shot, the pressure, size of the gate, thickness of the gate, and all other variables Mr. Cornell produced a sound die casting. It is a tail wheel casting for an airplane.

MR. CORNELL: That is right. I might offer a little comment on that problem of blistering of die castings. It is not necessarily a function of the radiographic soundness of the casting that you end up with. You can get a casting that is absolutely sound but it may be made in such a manner as to cause it to blister on being heated above a certain temperature, whereas another casting which is not sound, in fact, is quite porous, will not blister. It is more a function, as we have found it, of the way you make the casting rather than the quality that you end up with. In other words, it is not the porosity in the casting, generally speaking, which makes it blister.

MR. MCCLAIN: We of General Motors have analyzed die castings from various sources. It is necessary to eliminate as much of the entrapped air and gases as possible in order to produce a radiographically sound casting. It seems to be our trouble in finding castings that are free from entrapped gas and air. In heat

treatment this aluminum die casting blisters. I just wondered whether you had had any experience in eliminating that trouble.

Mr. Fox: A casting, unless it is definitely specified for a certain use requiring maximum soundness, is an ordinary casting. For instance, a gas meter housing should be sound at about 3 lb. gas pressure. To obtain the required soundness in the casting you must gate, vent, use the right die and casting metal temperature and the right pressure. If it is not called for, it is not usually done because of the added expense.

W. E. Martin[4]: At Sperry Gyroscope Company, Inc., we had the case of some little aluminum die casting levers which required increased strength. Some porosity could be found by radiography. Several thousands were solution heat treated at 960° F., quenched, and hardened at 300° F. None of them blistered. I cited this to somebody, and to prove it, got together a group of castings found to be sound by x-ray examination and they blistered on heat treatment.

Mr. Cornell: I think the crux of the matter is whether you want to have a casting that you can heat treat or whether you want a casting that is of sound structure. If you want a casting which will be resistant to leakage, then you want a sound casting. If you want one that you can heat treat, that is another story. There are two different problems there and an effort should be made to keep them separate.

Member: I would like to ask Mr. Cornell, if the blistering is not a function of the soundness of castings, what did you find was?

Mr. Cornell: We classify the factors in several general ways. One was that the section of the casting was one indication of whether or not it would blister. Generally speaking, a thin section would blister in preference to a heavy section casting, and the speed of injection or the gate velocity, whichever you prefer to term it, is the governing factor. Castings can be made in thick sections which are both sound and which do not blister, but it is a matter of how you make the casting and not so much what you end up with. In other words, fast filling generally tends to make castings which will blister, whereas solid front filling makes castings which will not blister.

D. Basch[5]: We have obtained large die castings, in sections varying from about 2 in. down to ⅛ in. with thin fins between them. We worked with Mr. Fox and his crew and the die castings not only, under a very sensitive x-ray, showed dense structure like wrought, but could be heat treated, solution heat treated, without the slightest blistering and with full realization of all the required properties.

What you say about the x-ray picture not being a criterion as to whether the casting can be heat treated or not is absolutely true. It depends on the gases which are held in equilibrium at room temperature and which are released at the higher temperature; it is also a matter of how you make the castings, which machines you use, what control you use.

I think when we take the general routine of die castings used for the over-the-counter articles, we cannot guarantee that the castings can be heat treated. As a matter of fact, in the majority of castings, they cannot be heat treated. But if you set out to make a casting which must be susceptible to heat treatment, which must pass the heat treatment without blistering or without any of the other bad effects which militate against anodization or any other treatment, you can do it. That is one of the MUSTS on the agenda of the die casting industry—to develop that. We cannot get such castings today as a regular routine matter, but there is nothing impossible about it. Do you not agree?

Mr. Cornell: You are entirely right, Mr. Basch.

R. F. Hauser[6]: We have gone into that problem extensively at Bendix and what Mr. Cornell and Mr. Basch have stated is also our conclusion. The x-ray does not indicate the amount of blistering that is likely to occur upon heat treatment of the castings. We have a magnesium die-casting alloy which we are normally heat treating at about 500° F., and although we have produced many castings I have yet to see one of them blister.

We have found that the amount of porosity that shows in the x-ray will certainly vary with the alloy composition. We find a large amount of that porosity is shrinkage porosity, and if it is eliminated you will have a sound casting, but if it is not eliminated, it still will not have anything to do with the casting blistering during heat treating.

Member: I would like to ask Mr. Fox whether or not he has data on physical properties of successfully heat treated aluminum die castings.

Mr. Fox: No, I do not think there are data available on that. In many cases, the as-cast condition of the die casting is superior to any heat treated sand casting except that of Alloy 220.

Mr. Basch: We ran many tests on a large number of heat treated castings and the physical properties of those heat treated castings (made of No. 356 alloy with 7 per cent silicon) came entirely up to our specifications, which required, in the heat treated condition, a minimum of 5 per cent elongation and about 24,000 psi. yield strength. In comparing the same castings with the forged part made out of alloy No. 25-S, the over-all end test indicated that the die castings made out of alloy No. 356, heat treated, came within 5 per cent of the forgings in tensile strength.

The way we tested them was not by tensile strength but in a breakdown test under high centrifugal force at a minimum speed of 50,000 rpm. and about a 6½-in. dia. With very few exceptions they withstood a long time cast at 5,000 rpm. without failure. There were no cracks between the thin fins and the large section or anywhere. That indicates what you can get with die castings when you go about it right.

Mr. Martin: These Recommended Practices are being drawn up for the most part to help new consumers or producers of die castings. In writing them, we should try to anticipate the possible ways people new to the industry are going to get into trouble. Most new producers are likely to have a bad case of hard spots. Further discussion on how to prevent hard spots should be in order and it is my opinion that the best way to minimize trouble with hard spots is to keep the metal temperature above 1200° F.

Mr. Fox: The proper use of the proper type of flux and degassing should be employed. We have found that the magnesium in the magnesium-aluminum type alloy having a greater specific gravity than aluminum or magnesium oxide tends to settle and be held in suspension.

Hard spots are associated with oxide inclusions, particularly in that type of alloy. In careless handling of the casting machine the operator can take up or ladle in some dross from the top, and those oxide spots are those so-called hard spots.

The prevention of hard spots is relatively easy. If you use a cover flux on the molten metal after first properly cleaning the melt you will then prevent it from oxidizing. We have found that magnesium oxide in that type of alloy has been detrimental, especially if you keep the melt uncovered for long lengths of time.

Mr. Martin: I had in mind trouble from hard spots in aluminum-base alloys containing silicon and copper or silicon with fairly high iron content.

Mr. Fox: It is a matter of not entraining that dross that accumulates on the top of the bath, keeping the melt clean, especially where you ladle out of that bath.

Mr. Martin: Do not hard spots of iron silicide start settling out when the metal temperature gets down near the melting point?

Mr. Fox: Yes, you will get a sludge, but that is not generally the cause of hard spots. The hard spots are associated more with oxides. This sludge that settles out is of a high iron-manganese-silicon composition. It may run as high as 8 or 10 per cent iron and similar amounts of manganese and silicon, but this seldom is the cause of hard spots. If you keep the temperature above 1200° F. very little sludge is apt to form.

W. J. Klayer[7]: Mr. Fox in his recommended practice talked about 500 psi. maximum for the air type die casting machines and 30,000 lb. for the other type. They are not synonymous. I think in one you have psi. and in the other you have lb. on the ram.

Mr. Fox: What I was talking about was the difference between a gooseneck type casting machine and the cold chamber type machine. We meant 500 psi. air pressure on the gooseneck machine as against 5,000 psi. up to any desired pressure on the cold chamber type machine.

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[1]Litemetal Dicast, Inc., Jackson, Mich.
[2]American Metal Co., Ltd., New York.
[3]Allison-Bedford Foundry, Bedford, Ind.
[4]National Smelting Co., Cleveland.
[5]Almin Co., Schenectady, N. Y.
[6]Eclipse Pioneer Div., Bendix Aviation Corp., Teterboro, N. J.
[7]Aluminum Industries, Inc., Cincinnati.

CORES FOR AUTOMOTIVE MALLEABLE CASTINGS

Sound castings through good core practice... a knowledge of the components of core ingredients, proper proportions, mixing and baking are important operational factors.

W. G. Ferrell
Superintendent of Foundries
Auto Specialties Mfg. Co.
St. Joseph, Mich.

AN EXACT KNOWLEDGE of the core sand components is important if good cores are to be obtained. A check should be made for clay content, and then for materials that will burn at contact with the molten metal. A determination of the size and shape of the sand grains will give an indication of the amount of oil to use, and also how the core will break down during solidification of the metal. Cores will burn out more freely when made with sand having a rounded grain.

Core Oils. The use of a good core oil is important, for it is generally known that a multitude of materials can be added to the oil, which have no basic value in the making of cores. Therefore, unless the purchaser has the necessary equipment for checking core oil, he must know the content of the oil. Chemical analysis of core oils will not indicate the quality of the cores or castings. The checking of core oil is a study in itself, and is most important to all foundries.

Core Oil Composition

However, to the average foundryman, the pertinent facts are: Is the oil right? Will it make good cores? Will it make good castings today, tomorrow and next week at the lowest possible cost? Although core oil is a widely used material, the actual makeup of its component parts is not generally known to the user. The author knows of no checking method that can be employed which will reveal to the user the exact amounts of linseed oil and other materials that are used in the manufacture of core oil.

It is possible to check the flash and fire points, specific gravity, saponification value and acid number, but the purchaser cannot go back to the core oil manufacturer and claim a deficiency in linseed oil or rosin because the check made will not prove it. The best practice is to secure a reliable manufacturer who can be depended upon for the production of consistently uniform oil.

Mixing Core Sand

An error generally made in mixing core sand is in the use of too much oil. It is much better to use just enough oil to allow handling of the cores. A rammed core, or one that is properly blown, needs but little oil. Another common error in foundries is the addition of more oil to prevent breakage by the careless molder or core handler. It is the author's contention that training employees in the proper handling of cores instead of using excessive amounts of oil will prevent much scrap caused by strains and tears, and also save expensive oil. The use of a minimum amount of core oil will aid in eliminating objectionable gas and odors sometimes present in foundries.

In this foundry it has been found that with the use of a high grade oil and a ratio of 1 part oil to 200 parts sand, and as high a ratio as 1 to 400, good cores for automotive malleable will result. By using such small amounts of oil the amount of water can be increased at mixing time without encountering trouble due to the standing of cores in the green state. This will aid in producing a good surface finish when using corn cereal binders, due to the migration of dextrine to the core surface at baking time.

In preparing a new core, basic information as to size of core and shape and size of the cavity to be cast is available, but the proper amount of core oil is not known. Therefore, to be on the safe side, it is common practice to use more oil than necessary. If a good casting is produced on this first try, a check is seldom made as to maintaining, or perhaps improving, castings quality with the use of a lesser amount of oil. A casting produced with less oil will be less liable to strains or the possibility of tears, in addition to savings resulting from the minimum use of expensive oil.

Cereal Binders

Corn Flour Binders. Use of corn flour is a general practice in making cores for malleable iron castings because of the high amount of green bond obtained without ill effects during casting. A number of corn flour binders are made by the wet-and-dry process. The author feels that the wet-process material has an advantage in that it will tolerate a greater amount of water at the time of mixing. This will aid the foundry having no drying equipment for raw core sand.

It is sometimes assumed that the checking of cereal binders is not important, as there is no reason for the binder to contain any material other than corn; yet samples checked by

Presented at a Malleable Foundry Practice Session of the Fiftieth Annual Meeting, American Foundrymen's Association, at Cleveland, May 7, 1946.

the burning method have been found to contain as much as 3 per cent clay or bentonite mixed with cereal binders.

Core Sand Mixing. In manufacturing automotive castings it is well to use as little bonding material as possible. It is the author's opinion that the exclusion of clay in core sand mixtures will produce better castings and require less core oil, with a resulting decrease in scrap. Clay additions force an increase in the amount of oil, producing a core which is hard all the way through, causing tears, cracks and strains in the castings. Where the need is found for higher hot strength to prevent gate burn-in, a far better practice is to paint the cores at gating points with a silica wash.

Batch Components

Core sand mixing is an important operation. Components of each batch of sand should be ascertained and recorded, as this is the only means by which results can be checked.

A number of different types of core sand mixers are available, some capable of thoroughly mixing a batch of core sand in 3 min., while others take as long as 10 min. The following method is suggested for checking the proper mixing time cycle:

Weigh or measure the sand and flour—mix together for 1 min. Then add water and oil, running total mixture an additional ½ min. Start taking test samples of sand every ½ min. until 20 samples have been taken, being careful to mark the time on each. After the 20 samples have been taken, three tensile strength test cores are made from each. Bake and cool carefully, check each for core hardness, and then run tensile strength tests. The first test samples will have a wide range of strength, but as testing proceeds the range narrows until the values are closely grouped. Proper mixing time is that of the test sample developing the highest strength, and mixing beyond this time is of no value. A few of the various core sand mixtures in the foundry with which the author is associated are shown in Tables 1 and 2.

In conclusion, it is recommended that foundries making automotive malleable iron castings observe the following points: Know the sand and mixing cycles; be certain of the oil and oil manufacturer; use as little core oil as possible; bake cores until all oil is thoroughly oxidized, leaving the core an even brown in color, and the castings will be sound as far as the cores are concerned.

DISCUSSION

Chairman: A. M. FULTON, Northern Malleable Iron Co., St. Paul, Minn.

Co-Chairman: C. F. JOSEPH, Saginaw Malleable Iron Div., General Motors Corp., Saginaw, Mich.

MR. JOSEPH: Did you run tests on various core oils to determine the best oil for use or have you just standardized on one oil and continued using it over a period of years? What is your practice in selecting your core oil?

MR. FERRELL: Over the period of a great number of years the picking of core oil was rather a haphazard procedure. We then adopted the chemical method of checking core oils, and that was not satisfactory. We then employed a research chemist to determine why we could not select a core oil.

There are three or four things which may be checked in a core oil. However, these do not always indicate exactly what is wrong with an oil. We have found it most satisfactory to purchase oil from a reliable manufacturer and take his word for it. We buy the best core oil there is and use as little as possible, depending on the reliability of the manufacturer.

DAVID TAMOR[1]: Can you tell us something of the hardness that you obtain in your production cores?

MR. FERRELL: The scratch hardness runs anywhere from 65 to 85 on regular production cores. We use a very soft core. It may be surprising to some men why we use such a soft core. We malleablize our iron of 1.40 to 1.60 per cent silicon in 15 hr. in electric roller-hearth continuous ovens.

MR. TAMOR: Is the 65 core hardness for extremely thin work, where you want collapsibility?

MR. FERRELL: That is right. The range is fairly high. Where we run low collapsibility we encounter very thin castings where we are in danger of getting cracks or tears.

E. G. STORIE[2]: What type of mixing equipment is most satisfactory?

MR. FERRELL: We find that a fast-action mixer is better for core sand.

P. R. SCHILLING[3]: Is there any reason, if you use such a diluted mixture of 1:200, that you would not find it profitable to stick to straight linseed oil?

MR. FERRELL: We have not because that necessitates a drier. Linseed oil is not rapidly oxidized. It requires an accelerator. The core should not be in the oven very long when it contains so little oil. If the core is in the oven a long time, the oil burns up leaving cores with no bond.

VICTOR PASCHKIS[4]: Has any attempt been made to bake the core by dielectric heat? You mentioned that the time of keeping the core in the oven has some bearing, and of course the time of oxidation has. In dielectric heating, you heat the core all the way through, and technically it should be superior, although it may not be economically possible to do so.

[1]American Chain & Cable Co., York, Pa.
[2]Fittings Limited, Oshawa, Ont., Canada.
[3]Superior Steel & Malleable Castings Co., Benton Harbor, Mich.
[4]Columbia University, New York.

Table 1

CORE SAND MIXTURES

Components, Properties	*Rear Axle Carrier, Main Body Core*	*Slab Type Gate Core*	*Front Wheel Hub Core*	*Truck Differential Case, Body Core*
Lake Sand, lb.	400	400	400	400
Cereal Binder, lb.	4	4	4	4
Core Oil, No. 45, pt.	1½	1½	—	—
No. 110, pt.	—	—	1½	1
Moisture, per cent	2.5	2.5	—	—
Green Permeability	205	219	—	—
Dry Permeability	253	243	—	—
Tensile Strength, psi.	105-110	175-180	55-60	55-60
Surface Hardness	60	70	—	—
Sintering Point, °F.	2417	2467	—	—
Baking Time, hr.	—	—	1	1
Temperature, °F.	—	—	450	450

Table 2

LAKE SAND SCREEN TEST

Screen Mesh No.	*Retained on Screen, %*
6	0.0
12	0.0
20	0.0
30	0.0
40	1.0
50	21.0
70	62.5
100	14.4
140	1.1
200	0.0
270	0.0
Pan	0.0

Grain Fineness No. 51.13

MR. FERRELL: We are now working on that, and have purchased a dielectric heating machine of 12 kw. size for that purpose. It looks as though only 3 or 4 min. would be required to bake a core which formerly required 1½ hr. of baking time.

CHAIRMAN FULTON: What type of core and what section of core is made with a 1:400 mixture?

MR. FERRELL: That mixture is used in a very thin slab core for Ford castings. We have one particular Ford casting that has a very thin band around it. We find that with 1.40 and 1.60 per cent silicon iron, we are forced to use practically no oil or the casting will crack. That is where we use the 1:400 mixture.

EARL M. STRICK[5]: In mixing the core sand, is there any difference in adding the oil or the water first?

MR. FERRELL: Our experience has been that adding the oil first does not produce results equivalent to adding the oil after the water. Our experience and tests prove that the water must be added before the oil to get efficient bonding action. Our total mixing time is 2½ min.

CHAIRMAN FULTON: Do you start your mixture from a dry sand?

MR. FERRELL: Yes.

CHAIRMAN FULTON: Then you know definitely the amount of moisture?

MR. FERRELL: We do. We dry all core sand coming into the plant.

F. L. HARRIS[6]: Have you any hot strength tests to correspond to the hardnesses 65 to 120?

MR. FERRELL: I think there are some in the paper. Are you referring to sintering tests?

MR. HARRIS: I was thinking more of the hot strengths at 2000° F., in other words, collapsibility that you get with those hardnesses.

MR. FERRELL: I do not have them. All I can give you are the sintering points. We find sintering of the 1:400 mixture occurs at 2467° F.

MR. HARRIS: Do you have tensile strengths that correspond with those hardnesses?

MR. FERRELL: We do. The potential strength on these cores will run 110, 110, 105, 110. Surface hardness was 6.

MILTON TILLEY[7]: Would you describe your core-baking process? Do you put all sizes of cores through the same oven? What are the baking time and temperature? Do you over-bake your small cores and under-bake your large cores? Do you do all your baking in the oven, or does some take place outside the oven?

MR. FERRELL: We try to bake the cores completely in the oven in order to get rid of the smoke. We do segregate the large cores. It is impossible to bake all the cores with that small amount of oil in the same oven. So our baking temperature is 450° F. and our time is 1 hr. and 10 min. in the oven.

O. J. MYERS[8]: I was interested to learn of your test for mixing the core sand. Is the green bond also influenced by that mix, and do you get maximum green bond when you get your maximum tensile strength?

MR. FERRELL: No. We are interested only in getting the maximum oil out of it, the strength in it. That is why we say that that method of testing is efficient as far as the oil is concerned. We do say that when you reach the end point, when the tensile strength of the core is grouped closely together, there is no further need of going into it. We find a wide range. Of course, different mixers will give you longer times.

MR. MYERS: How much water do you use?

MR. FERRELL: We use 1½ to 2 per cent moisture.

MR. MYERS: Is it put in only for green bond strength?

MR. FERRELL: Yes.

MR. STRICK: Have you tried anything to substitute for corn flour? If so, what affect did it have on the collapsibility of the cores?

MR. FERRELL: The ones we have tried have affected it very badly. We have used wheat flour and other forms of flour. We have used other chemical compounds. We have not found one that produces the desired results. As a matter of fact, we are going back to regular malleable iron because the 1.40 to 1.60 silicon iron requires very collapsible cores.

G. L. GALMISH[9]: What happens if you run 1.40 to 1.60 silicon?

MR. FERRELL: The iron is very prone to hot tear, and there are certain shapes of castings that cannot be manufactured with it. It makes a good malleable iron after it is annealed.

B. L. HAYS[10]: Are most of your cores made on blowers?

MR. FERRELL: About 95 per cent of our cores are made with blowers.

CHAIRMAN FULTON: Do you find that a blown core requires more or less oil than a rammed core?

MR. FERRELL: We do not need as much oil in the blown core. We find the blown core gives us better results for it has greater permeability than the rammed core.

MR. STRICK: Do you have iron penetration caused by some of your cores? If so, what do you do to eliminate it?

MR. FERRELL: We have very little iron penetration. If we have iron penetration in the gates, we put a little core wash there, rather than increase the oil for in our condition we dare not increase the oil. That would increase the hardness, and we must avoid that. We use a little surface core wash when we get iron penetration.

CHAIRMAN FULTON: We assume then that the higher silicon content iron you run, the softer the core you use.

MR. FERRELL: That is correct. It is not possible to make the grade of high silicon iron using hard cores. Our cores are so soft we have to make special containers to handle them. They are set in at the core room and carried to the molder in those containers. The cores have to be handled very carefully.

CHAIRMAN FULTON: Do you get many cracked castings due to cores?

MR. FERRELL: I do not think a soft core will cause a crack. We know that lack of collapsibility is the cause of tears. Strain cracks may be caused by something different, possibly gating.

MR. TAMOR: Would you comment on the use of wood flour to produce a collapsible core with thin walls?

MR. FERRELL: We do not use wood flour. We have tried it and found no advantages.

MR. TAMOR: I do not think there is much question but that wood flour in cores gives good collapsibility. It will impart it to molding sand. The reason we are using corn flour is to get green strength and collapsibility. Wood flour will give collapsibility but will not give green strength in the cores, and you have to use something else with the wood flour which probably gives you a harder core and less collapsibility in the finished product.

[5]Erie Malleable Iron Co., Erie, Pa.
[6]Belle City Malleable Iron Co., Racine, Wis.
[7]National Malleable & Steel Castings Co., Cleveland.
[8]Werner G. Smith & Co., Minneapolis.
[9]Michigan Malleable Iron Co., Detroit.
[10]Union Malleable Iron Works, E. Moline, Ill.

COPPER AND THE STEEL CASTING

An Alloy for POSTWAR CONSIDERATION

H. F. Taylor
H. F. Bishop
and
R. C. Wayne

COPPER HAS BEEN EMPLOYED as an alloying element in many of the so-called high yield strength, low-alloy structural steels for a number of years. It has not been used widely in wrought steels for machine parts and there are no S.A.E. or N.E. steels which contain copper other than in residual quantities.

Although the limited supply of copper in time of war alone could account for the omission of this element in the N.E. steels, two other reasons exist for not using it. When more than 0.4 per cent copper is present, steels become subject to edge-cracking in the rolling operation unless the temperature is carefully controlled or unless sufficient nickel is present to counteract this difficulty.

Mechanical Properties Affected

The second reason is that the beneficial effects on mechanical properties conferred by copper are more pronounced after normalizing and tempering than after quenching and tempering, in which latter condition most of the S.A.E. and N.E. steels are used; moreover, the benefits are most evident when the carbon content is 0.20 per cent or lower, and most of the S.A.E. or N.E. steels are used at higher carbon contents unless they are case-hardened.

Difficulties outlined in the foregoing naturally restrict the use of copper to structural steels, certain special purpose steels, steels for heavy forgings (which usually contain appreciable quantities of nickel) and steel castings. Although copper has not been used widely in the steel castings industry except by the Pacific Car and Foundry Co. and by the Ford Motor Co., the excellent casting characteristics and high physical properties, which can be developed in both light and heavy sections, should make the copper alloy steels of interest for postwar castings.

Although much work has been done on the heat treatment and physical properties of copper steels by other investigators, it was considered advisable to check certain of these results and also to study weldability, mass effect, and high and low temperature properties of these steels to determine if it was desirable to widen their sphere of application. The results which are reported herein were obtained before copper became a critical metal during the war.

Brief Review of Existing Knowledge. Several outstanding advantages of copper-silicon cast steels are worth considering in choosing a high strength, low alloy steel of weldable grade for castings.

From the standpoint of casting long, thin members, copper steels are unique. When sufficient copper is added in combination with higher than normal silicon, steels become very fluid in the molten condition. Finlayson[10] has reported that boom castings with overall lengths of approximately 11 ft. and maximum metal thickness of ½ in. can be molded in green sand and poured through a single gate at one end.

Both Kinnear[4] and Nehl[2] made early investigations upon copper cast steels containing normal amounts of manganese and silicon. Kinnear was the first to predict the practical advantages of the precipitation hardening (aging or age hardening) of copper steels. This heat treatment gives maximum strength with little or no change in ductility over the annealed steels.

Iron-Copper Diagram

According to the iron-copper equilibrium diagram of Norton[6], iron can dissolve about 1.5 per cent of copper at 1531° F. (833° C.) but retains only about 0.35 per cent in solution below 1200° F. (650° C.). However, the copper dissolved in excess of 0.35 per cent is precipitated only upon very slow cooling (some of the dissolved copper being retained in solid solution at the center of heavy sections even after air-cooling) and it is difficult to produce the desired sizes of precipitated particles in the cooling cycle.

Therefore, to obtain optimum particle size consistently and throughout varying cross-sections, it is good practice to retain, at room temperature, all of the copper which was in solution at the heating temperature by using the proper cooling rate and then precipitating the copper by reheating. The control of time and temperature is not too critical, as will be shown by data which are included in this paper.

Copper steels differ from most

The opinions expressed are those of the authors, H. F. Taylor (now Research Associate at M.I.T., Cambridge, Mass.), H. F. Bishop, and R. C. Wayne, Steel Castings Section, Division of Physical Metallurgy, Naval Research Laboratory of the Office of Research and Inventions, Washington, D. C., and do not necessarily reflect the views of the Navy Dept.

other precipitation hardenable alloys in their capacity to increase in strength and hardness without a serious loss in ductility. This may be due to the fact that the submicroscopic, keying particles of copper precipitated upon artificial aging are relatively soft, and yield plastically. under deformation.

A high yield to tensile strength ratio is typical of precipitation hardened alloys; in "as normalized" copper steels, the marked increase in this ratio is undoubtedly due to some precipitation of properly-sized copper particles upon air cooling.

Copper Does Not Oxidize

Another advantage of using copper as an alloying element in steel for castings is the fact that copper is not lost by oxidation or volatilization in the melting process. Heads, gates, and other return scrap yield all of their copper upon remelting.

A mass of data on copper steels has been compiled by Gregg and Daniloff[7]. Lorig and his co-workers[8, 9] have contributed much to the knowledge of copper steels, and other investigators have been active in this field. A chronological selected bibliography on copper steels appears at the end of this paper.

Specimen Preparation. The steels used in this work were melted in a 300-lb. acid induction furnace. Sheared S.A.E. 1015 steel and ingot iron were used as melting stock. Ferro-alloys were used for alloy additions except nickel and copper, which were added as commercially pure metals. The steels were sand-cast in 60-lb. coupons of the type shown in Fig. 1. The projecting fingers were sawed from the main bodies of the coupons before they were heat treated.

Shoulder-grip, 0.505-in. diameter tensile specimens were machined from these blanks and tested according to conventional methods unless otherwise noted. Blanks for impact specimens were cut from the bottom portions of the coupons and were heat treated before they were machined to standard Charpy key-hole notch specimens unless otherwise indicated. Each physical property test was made in duplicate.

Effect of Copper on Tensile Properties. Figure 2 shows the tensile properties of a steel containing 0.15 per cent carbon, 0.40 per cent silicon and 0.60 per cent manganese at eight copper contents from 0.09 per cent to 2.10 per cent in two conditions of heat treatment; normalized, and normalized and aged.

Tensile Strength Affected

Effect of aging upon tensile strength and yield strength begins to appear when about 0.50 per cent copper is present, and the strengthening effect increases with increasing copper content until about 1.5 per cent is present; beyond that amount, little additional strengthening occurs. By adding 1.5 per cent copper to the base composition and aging after normalizing, the yield strength is doubled, increasing from 38,000 to 80,000 psi., while the tensile strength is increased from 60,000 to 100,000 psi.

Reduction of area and elongation values are erratic after aging and are lower than might be expected. However, this particular base composition is not well suited to the development of the best properties in precipitation hardened copper steels and serves only as an illustration of the limits within which copper is effective as a precipitation hardening agent.

Effect of Manganese on Tensile Properties. Figures 3 and 4 illustrate the influence of manganese on the mechanical properties of low carbon copper steels. Manganese slightly improves the strength of copper steels which are annealed, and normalized and tempered but, as the manganese content increases, the steel becomes less responsive to the aging heat treatment.

Effect of Silicon on Tensile Properties. It has been stated that silicon and copper are effective in increasing the fluidity of molten steel. A series of copper steels with increasing silicon contents was cast into test coupons to determine whether or not the increased silicon would have any detrimental effect upon the physical properties. The results obtained from tests on these steels after four different heat treatments are shown in Figs. 5 and 6.

When both the manganese and silicon are about 1.00 per cent, a

Copper increases the yield and tensile strengths and yield ratios of steels to which it is added. This increase in yield strength to tensile strength ratio makes low alloy copper-bearing cast steels particularly useful for castings where a saving in weight is desirable. These improvements are obtained without materially affecting ductility. With manganese and silicon present in amounts higher than normal, an excellent combination of properties is obtained.

With copper contents above about 0.6 per cent these steels can be precipitation hardened (aged) without liquid quenching to further increase the yield and tensile strengths. In the annealed condition a copper steel containing 0.15 per cent carbon with manganese and silicon each at approximately 1.00 per cent will have a yield strength of 60,000 psi., and by aging this steel after normalizing a yield strength of 90,000 psi. is possible. Copper steels in the aged condition have low impact resistance as measured by Charpy and Izod tests, but have good elongation and reduction of area. Because a supersaturated solution of copper in ferrite is obtained by even very slow cooling, sections as large as 8 in. can be hardened throughout by the precipitation strengthening draw after normalizing.

Copper cast steels are best below 0.20 per cent carbon, and if kept within this limit have good weldability. With approximately 0.10 per cent carbon present, these steels are readily weldable without preheating or postheating at yield strengths above 50,000 psi. The combination of copper and 1.00 per cent or more silicon produces excellent fluidity in these steels. They do not become brittle at subzero temperatures, and high temperature tensile tests indicate that copper steels containing about 0.50 per cent molybdenum show promise for service at moderately high temperatures. The resistance of copper-bearing steels to atmospheric corrosion is well known, and curves are presented showing copper steels to be more resistant to scaling during heat treating than are plain carbon or carbon-molybdenum steels.

very good combination of properties is obtained. Figure 7 shows some properties of high silicon-manganese steels with the copper varying from about 1.10 to 1.80 per cent.

It is a peculiarity of this composition that aging after normalizing increases the reduction of area value by about 5 per cent as well as the yield and tensile strength, while the elongation decreases slightly. This anomaly has been noticed many times and is caused by the fact that the ductility of the aged tensile bar manifests itself in a localized area with a very pronounced "necking."

Effect of Carbon. Since copper is soluble in ferrite, the precipitation hardening effect is most pronounced in low carbon steels. Copper steels should not contain over 0.25 per cent carbon and preferably the carbon content should be less than 0.20 per cent, not only because the best combination of physical properties is obtained within this range, but also because these steels are weldable, as is shown later in this paper.

Stress Relieved

In the stress relieved condition, obtained by annealing or normalizing and tempering, a copper steel containing 0.15 per cent carbon, as shown in Fig. 8, has a yield strength of 60,000 psi. with excellent ductility and impact resistance. When this same steel is normalized and aged, the yield strength is raised to almost 90,000 psi., as shown in Fig. 9, and good elongation and reduction of area are maintained. The impact resistance of copper steels always drops considerably on aging.

The curves of Fig. 9 show how the precipitation strengthening effect decreases with increasing carbon. With 0.10 per cent carbon the yield strength of a normalized steel is increased by 20,000 psi. due to the precipitation hardening draw, while a steel containing 0.25 per cent carbon shows an increase of only about 10,000 psi. Figure 10, taken from work by Williams and Lorig[9], shows the same behavior for rolled steels over a much greater carbon range.

Cumulative Effect of Copper, Manganese, and Silicon. It was found that in order to obtain the best combination of strength and ductility in copper steels either the manganese or silicon contents or both should be about 1.00 per cent. Low carbon copper steels containing 1.00 per cent manganese and 1.25 per cent silicon have been used in making commercial castings with very good results. This alloy presents no melting problems, and has excellent fluidity, and high strength and ductility.

Table 1 lists the properties of a plain carbon steel and also a copper steel of the same carbon content but with 1.25 per cent siiicon and 1.00 per cent manganese present. Table 2 lists the improvements in tensile and yield strengths which are caused by the copper and the excess silicon and manganese over that present in the base steel; the effect of each element is determined separately by increasing the amount of one element in the steel and holding the other two constant.

The sum of improvements due to each element on the yield and tensile strengths of the steel is 17,000 and 14,500 psi., respectively, while the improvement due to the combination of these elements, as shown in Table 1, is 26,000 psi. in yield strength and 18,000 psi. in tensile strength.

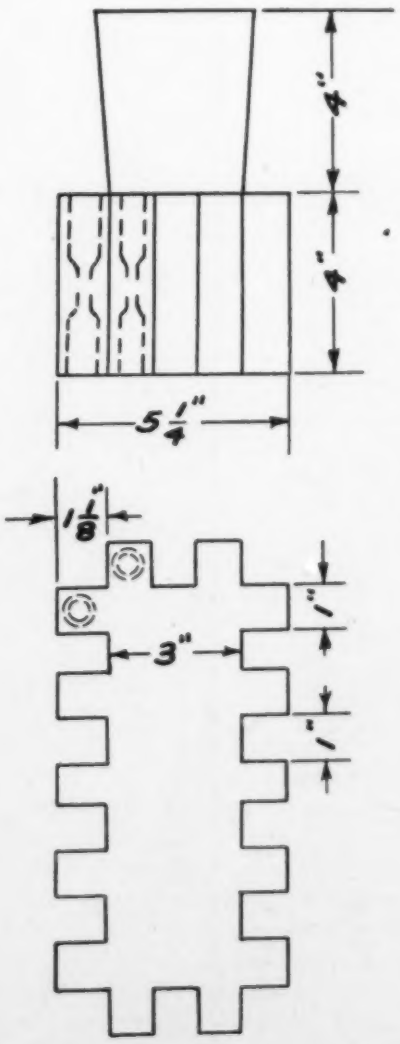


Fig. 1—Views of test coupon.

Table 1

Comparison of Plain Carbon and Copper Bearing Cast Steels

Normalized 1700° F. (930° C.)
Tempered 1200° F. (650° C.)

	Composition			
Steel	*C*	*Mn*	*Si*	*Cu*
A	0.20	1.0	1.25	1.75
B	0.20	0.7	0.35	0.00

	Physical Properties			
Steel	*Tensile Strength, psi.*	*Yield Strength, psi.*	*Elongation, per cent in 2 in.*	*Reduction of Area, per cent*
A	86,000	62,000	28.0	57.0
B	68,000	36,000	32.0	56.0
Increase	18,000	26,000	—4.0	1.0

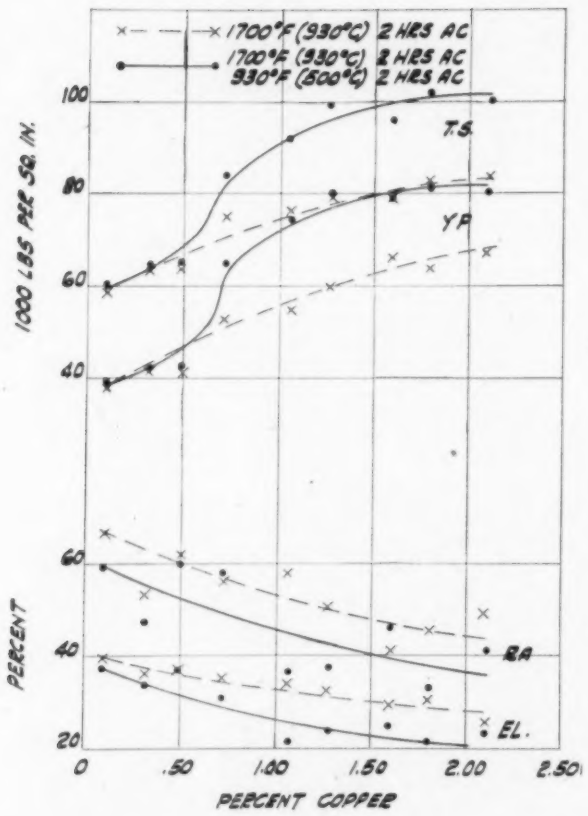
This increase in the yield strength is not accounted for when the effects of each element are added together, the cumulative effect being roughly twice the total effect of each element considered separately. It has recently been shown that the cumulative effect of alloy additions is obtained by multiplying factors for each element in the alloy rather than by adding these effects directly.

Copper Affects Yield Strength

Tables 1 and 2 show the marked influence of copper on yield strength. High yield strengths and yield ratios are characteristic of copper bearing steels. The data presented in these tables are for normalized and tempered steels. By aging these steels, further increases of between 20,000 and 30,000 psi. can be obtained.

Aging Time and Temperature. The tensile properties obtained by aging a copper steel at 870° F. (465° C.) and 930° F. (500° C.) for varying lengths of time are shown by the curves of Fig. 11. By aging at 930° F. (500° C.) maximum strength is reached in about 3 hr., and the strength decreases after about 6 hr. when the over-aging effect becomes pronounced.

When aging takes place at a lower temperature, 870° F. (465° C.), changes occur less rapidly, with maximum strength being reached after about 8 hr., and no marked over-aging is apparent even after 16 hr. For commercial practice, a good



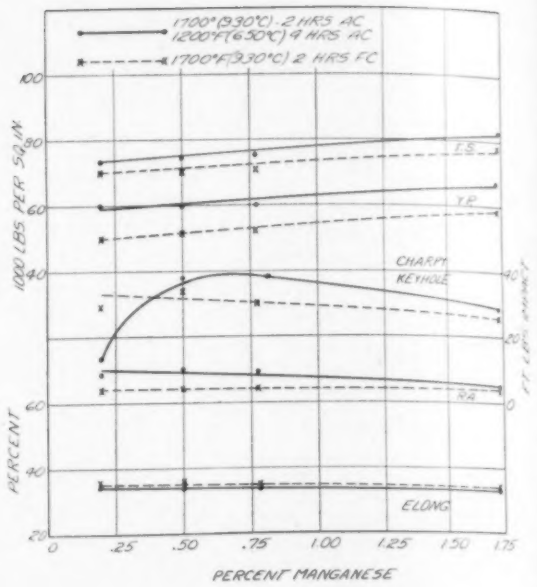
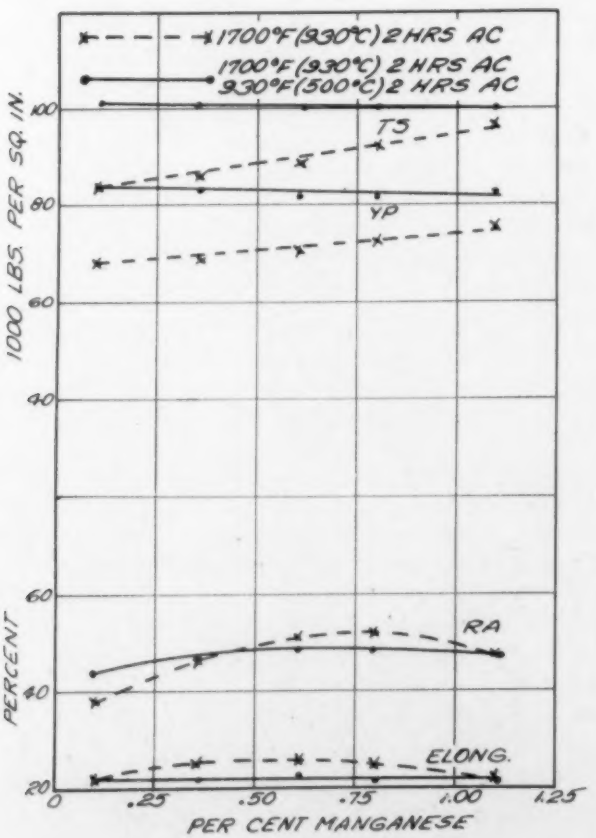


Fig. 2 (left)—Effect of copper on mechanical properties of copper steels. Composition, per cent —C, 0.15; Mn, 0.60; Si, 0.40; Cu as shown.

Fig 3 (above)—Effect of manganese on mechanical properties of copper steels. Composition, per cent—C, 0.08; Si, 1.00; Cu, 1.75; Mn as shown.



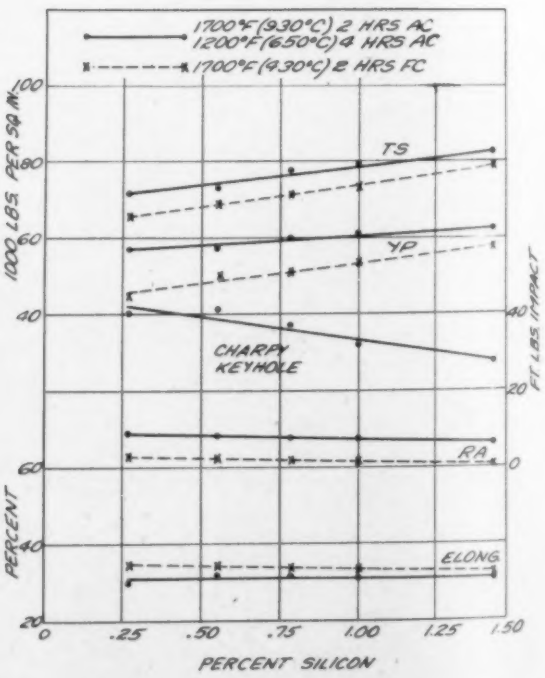


Fig. 4 (left)—Effect of manganese on alloy of composition, per cent—C, 0.12; Si, 1.07; Cu, 2.08.

Fig. 5 (above)—Effect of silicon on alloy of composition, per cent—C, 0.12; Mn, 0.83; Cu, 1.85.

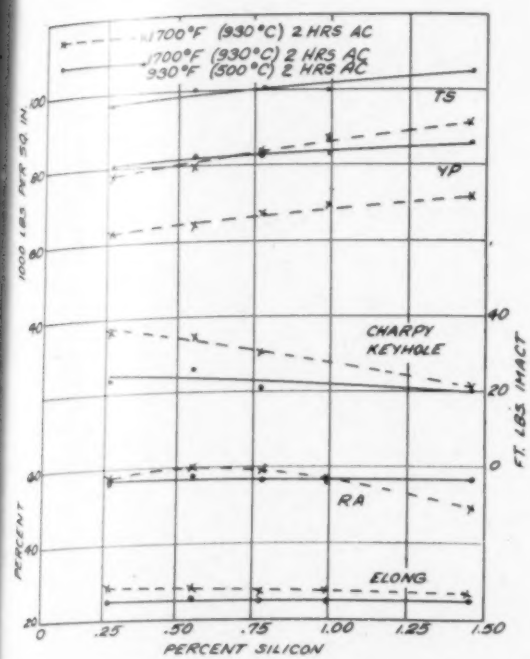


Fig. 6—Effect of silicon on mechanical properties of copper steels. Composition, per cent—C, 0.12; Mn, 0.83; Cu, 1.85; Si as shown.

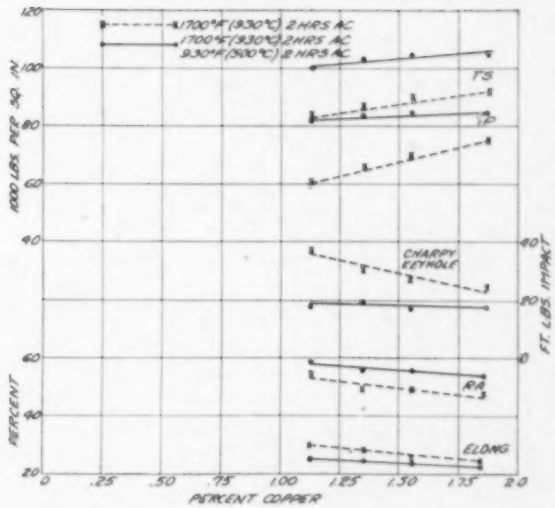


Fig. 7—Effect of copper on mechanical properties of copper steels. Composition, per cent—C, 0.16; Mn, 0.96; Si, 1.07; Cu as shown.

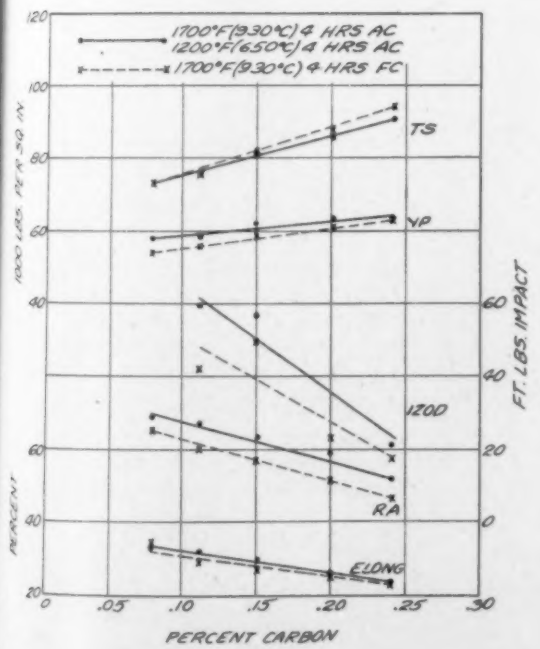


Fig. 8—Effect of carbon on mechanical properties of copper steels. Composition, per cent—Mn, 1.0; Si, 1.0; Cu, 1.75; C as shown.

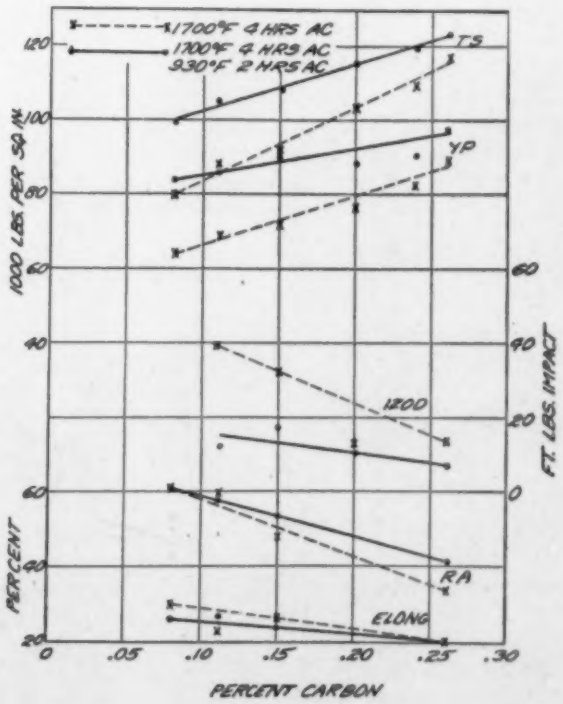


Fig. 9—Effect of carbon on mechanical properties of copper steels. Composition, per cent—Mn, 1.0; Si, 1.0; Cu, 1.75; C as shown.

Table 2

EFFECT OF COPPER, MANGANESE, AND SILICON ON COPPER STEELS

Normalized 1700° F. (930° C.) — Tempered 1200° F. (650° C.)

Composition, per cent				Properties			
C	Mn	Si	Cu	Tensile Strength, psi.	Yield Strength, psi.	Elongation, per cent in 2 in.	Reduction of Area, per cent
0.17	1.21	0.97	1.50	87,000	66,000	30	61
0.17	1.21	0.97	—	81,000	57,000	31	63
Increase caused by raising Cu from 0.0 to 1.5 per cent				6,000	9,000	—1	—2
0.08	1.00	1.00	1.75	78,000	63,000	33	67
0.08	0.70	1.00	1.75	77,000	62,000	33	68
Increase caused by raising Mn from 0.70 to 1.00 per cent				1,000	1,000	—	—1
0.12	0.83	1.25	1.85	82,000	62,000	31	67
0.12	0.83	0.35	1.85	72,000	57,500	31	69
Increase caused by raising Si from 0.35 to 1.25 per cent				10,000	4,500	—	—2
Total of Increases				17,000	14,500	—1	—5

Table 3

WELD PROPERTIES USING TEE BEND TEST

(Test pieces annealed after casting.)

Heat No.	Composition, per cent C	Mn	Si	Cu	Mo	Test Piece	Heat Treatment	Max. Load, lb.	Angle at Max. Load, degrees	Angle at Fracture, degrees
B	0.21	0.75	0.40	—	—	Cast Fillet	As Annealed	8,400	71	
						Weld Fillet	As Welded	8,100	62	
						Weld Fillet	1200° F. (650° C.) 1 hr. AC	8,800	71	
G	0.17	0.61	0.44	—	0.51	Cast Fillet	As Annealed	10,300	73	
						Weld Fillet	As Welded	7,850	56	
						Weld Fillet	1200° F. (650° C.) 1 hr. AC	8,100	75	
M	0.11	0.63	0.92	1.79	—	Cast Fillet	As Annealed	8,250	68	92
						Weld Fillet	As Welded	8,000	55	55
						Weld Fillet	1200° F. (650° C.) 4 hr. AC	9,350	65	over 120
D	0.16	0.78	0.45	2.02	—	Cast Fillet	As Annealed	9,300	72	
						Weld Fillet	As Welded	9,150	29	
						Weld Fillet	1200° F. (650° C.) 1 hr. AC	9,600	63	
H	0.16	0.90	1.11	1.67	—	Cast Fillet	As Annealed	10,500	66	
						Weld Fillet	As Welded	8,700	30	
						Weld Fillet	1200° F. (650° C.) 1 hr. AC	8,700	53	
L	0.15	1.08	0.98	1.77	—	Cast Fillet	As Annealed	9,750	60	66
						Weld Fillet	As Welded	8,150	29	29
						Weld Fillet	1200° F. (650° C.) 4 hr. AC	9,500	65	75
F	0.20	1.12	1.12	1.95	—	Cast Fillet	As Annealed	11,250	63	
						Weld Fillet	As Welded	8,950	20	
						Weld Fillet	1200° F. (650° C.) 1 hr. AC	10,000	45	
J	0.24	1.12	1.05	1.70	—	Cast Fillet	As Annealed	11,500	71	75
						Weld Fillet	As Welded	7,100	20	20
						Weld Fillet	1200° F. (650° C.) 4 hr. AC	8,500	36	36

aging treatment for copper steel castings is 900° F. (480° C.) for 5 hr., which is midway between the two extremes shown in the curves.

This aging time and temperature are not very critical, as the curves are fairly flat at maximum strength values, and varying the time at temperature an hour more or less, or the temperature by ± 25° F. (15° C.) will not appreciably affect the properties obtained. With castings of light section, the upper limit of aging can be used due to the fact that thermal equilibrium in the casting is reached quickly, while for heavy sections more uniform properties throughout the casting are obtained by aging at a lower temperature such as 870° F. (465° C.) for 8 hr.

Copper Steel in Heavy Sections. An advantage of the copper-iron precipitation hardening system is that drastic quenching is unnecessary prior to aging. Cast cylindrical sections up to 8 in. in diameter were found to respond to the aging treatment after air cooling. The strength and hardness at the centers of 2, 4, 6, and 8-in. sections are shown graphically in Fig. 12. As the section size increases, the improvement in strength on aging is reduced, because, since the cooling rate from the solution temperature is slower, less copper is retained in supersaturated solution.

Nevertheless, with these steels, the strength at the center of an 8-in. section can be increased by 10,000 psi. simply by aging after normalizing. The 2-in. cylinders increased in strength about 18,000 psi. after aging. Hardness surveys across each of the sections tested showed the hardness to be quite uniform although the hardness level became lower as the section size increased.

Cylinders Water Quenched

A series of these cylinders was also water quenched from 1700° F. (930° C.) and subjected to the aging treatment. Hardness surveys were then made across disks cut from these cylinders in both the "as quenched" and in the quenched and aged condition. The hardness across these sections after quenching was not as uniform as it was after normalizing, but decreased from the surface to the center.

In Fig. 13 the hardness at the center and surface of these cylinders is shown before and after aging. Whereas after air cooling from the solution temperature the aging effect diminished as the section size increased, when these sections are water quenched the aging effect becomes more pronounced as the section size increases to 8-in. diam.

In the small sections a certain amount of hardening is present upon quenching due to the carbon present and, when these steels are aged, the increase in strength due to aging is superimposed upon the softening due to tempering. The large 8-in. section, although water quenched, cools at a rate which is comparable

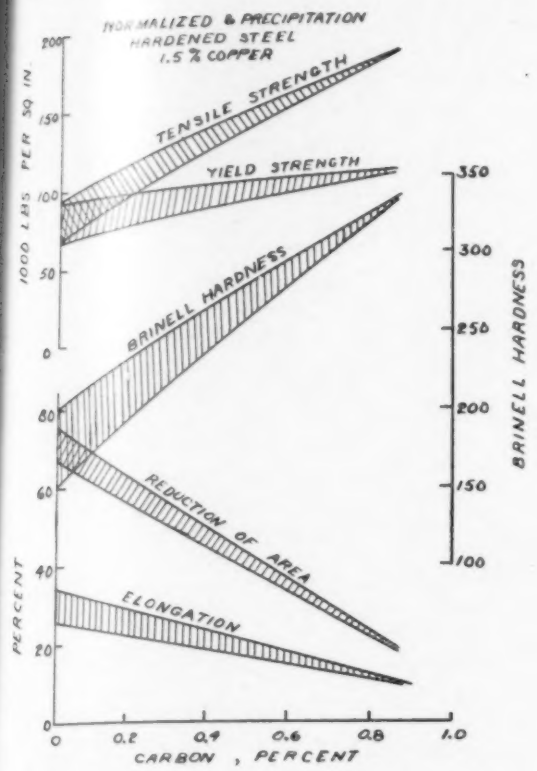


Fig. 10—Magnitude and direction of property changes brought about by precipitation hardening 1.5 per cent copper steels of varying carbon contents. (Williams and Lorig, "Metals and Alloys," March, 1936.)

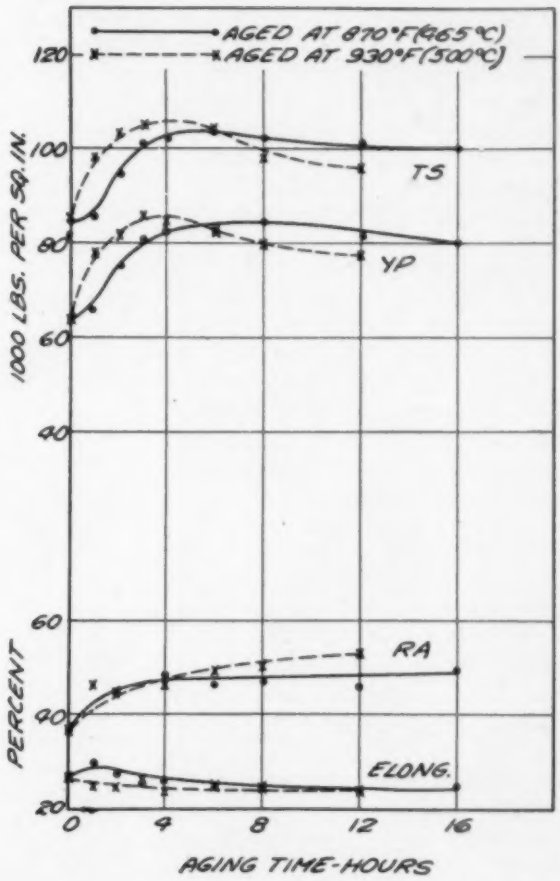


Fig. 11—Effect of aging time and temperature on steel of—C, 0.14; Mn, 0.56; Cu, 1.36; Si, 0.99.

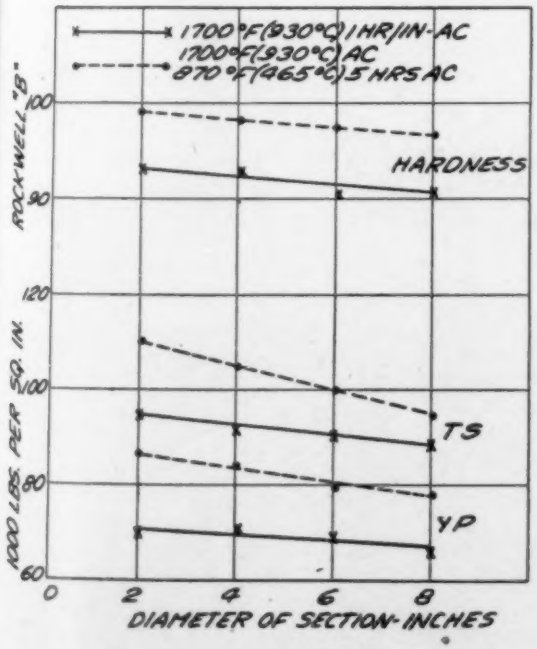


Fig. 12—Effect of aging on large sections after normalizing treatment. Composition, per cent—C, 0.16; Mn, 1.02; Si, 1.09; Cu, 1.49.

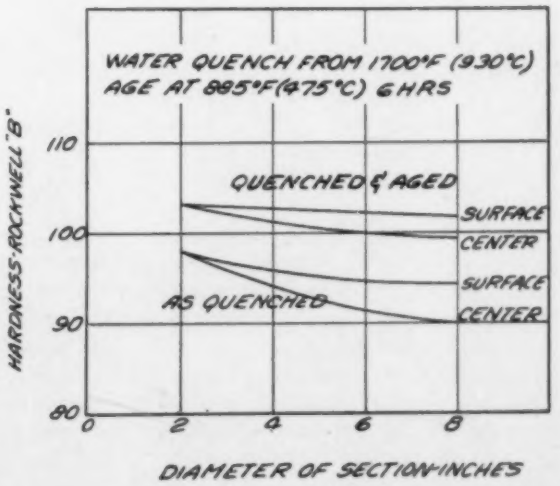


Fig. 13—Effect of aging on large sections after water quenching treatment. C, 0.14; Mn, 0.80; Si, 1.31; Cu, 1.48 per cent.

to the cooling rate of an air cooled one-in. section.

The hardness at the center of the 2-in. section increased by 5 points on the Rockwell B Scale, whereas hardness at the center of the 8-in. section increased 10 points. When strength in large sections is desired it can be secured economically by quenching a copper steel and aging, rather than by obtaining sufficient hardenabiilty by means of more expensive alloying elements.

Fluidity of Copper Steels. Figure 14 shows fluidity curves of a copper silicon steel and also of a plain carbon steel as measured in the spiral fluidity mold developed at the Naval Research Laboratory[18]. For any pouring temperature the copper silicon steel flows from 6 to 8 in. farther in the spiral fluidity mold than does the plain carbon steel.

Low Temperature Properties. It was thought that these steels would be desirable for use in ships' castings because their high yield strength would make it possible to design on a higher strength basis with a corresponding weight reduction of the castings. Thus it would be necessary that they exhibit good properties at low temperatures.

Impact Tests

Low temperature impact tests were made on a copper steel, and the results are compared with those obtained from a plain carbon and a 2.90 per cent nickel steel which is noted for its low notch-sensitivity at low temperatures. These steels were all normalized and tempered at 1200° F. (650° C.), and contained 0.15 per cent carbon.

The nickel steel, as shown in Fig. 15, is superior in impact resistance at low temperature, but the copper steel is distinctly better than the plain carbon steel. Tempering at 1200° F. (650° C.), which is well above the age-hardening temperature, places the copper steel in its softest condition. Aged copper steels have low impact values, and these steels in the aged condition would undoubtedly show poor impact properties at sub-zero temperatures.

High Temperature Properties. The curves of Fig. 16 show the results of short-time tensile tests made at temperatures as high as 1400° F. (760° C.) with 0.313 in. diameter specimens. All specimens were annealed before testing. A well-insulated furnace was used and temperatures were held within ± 5° F. (3° C.) by means of an accurate controlling pyrometer which was checked regularly with a portable potentiometer.

Temperatures were measured with chromel-alumel thermocouples spot-welded to the specimens. The tests were not started until each specimen had been held at temperature for ½ hr. and the rate of loading was low; at least 5 min. were taken to load each specimen.

Curves Plotted

Stress-strain curves were plotted from the loads indicated by the testing machine and extensions read with an optical extensometer accurate to 0.0002 in. Platinum reference lines were used on the specimen at a gauge length of 1¼ in. and sightings were taken through quartz-covered slits in the furnace.

It was thought that, since molybdenum has been used to improve the high temperature characteristics of carbon steel, it might act similarly in copper steels. This would be desirable because of the greater inherent strength of copper steels and because of their good behavior in the foundry.

Each point on the curves of Fig. 16 represents results from two tests which did not vary by more than 5 per cent. The copper steel shows consistently higher proportional limits and tensile strengths than does the carbon-molybdenum steel. The coper-molybdenum steel has the highest tensile strength of the three steels at all temperatures, but the proportional limit is initially lower than it is for the plain copper steel.

Indication of Tests

Although these tests serve only a superficial purpose, they do indicate that copper-molybdenum steels have promise for service at moderately high temperatures. Their high proportional limits and tensile strengths make them worthy of consideration for service at temperatures below 750° F. (400° C.).

Scaling Resistance of Copper Steels. The beneficial effects of small amounts of copper for increasing the resistance of steel to atmospheric corrosion are well known. The curves of Fig. 17 show that the presence of copper in steel also increases its resistance to oxidation in furnace atmospheres during heat treatment. These results were obtained on small machined specimens of identical shape; each specimen weighed slightly less than one lb. The specimens were weighed in previously ignited crucibles and placed in a furnace with normal atmosphere at 1600° F. (870° C.).

After an exposure of one hr. the pieces were cooled in a dessicator and weighed. Without removing the oxide scale, heating, cooling, and weighing were repeated several times to obtain the data shown in Fig. 17. The increases in weight of the specimens are due to the oxygen which combined with the iron to form the scale and may be taken as a measure of the relative amounts of scaling on carbon steel, carbon-molybdenum steel, and copper steel.

The copper steel showed a consistently greater resistance to oxidation than did the carbon and carbon-molybdenum steels, which oxidized at about the same rate. An interesting difference in the steels was in the type of scale formed.

Scale Formed

That scale which formed on the carbon and carbon-molybdenum steels blistered and flaked off leaving a fresh surface for further oxidation, while on the copper steel the oxide layer was more dense and tightly adherent. A copper-rich film is left beneath the scale which is extremely resistant to oxidation, and this probably is the greatest factor in preventing scaling of copper steels.

In larger scale tests on sandblasted castings which weighed approximately 5 lb. each, the plain carbon steel gained 1.25 per cent in weight at 1700° F. (930° C.) for 18 hr., while the copper steel gained only 0.96 per cent by weight.

Weldability of Copper Steels. The weldability of copper steels was studied by means of the tee-bend test. In this test two castings were made, as shown in Fig. 18, for each composition tested. One casting was made with square corners at the junction forming the tee, and the other was made with filleted corners. The plates were annealed after casting and a weld bead was laid along each internal corner of the unfilleted plate so that the two plates had the same shape, one with cast fillets and the other with weld fillets.

The plates were then sectioned to

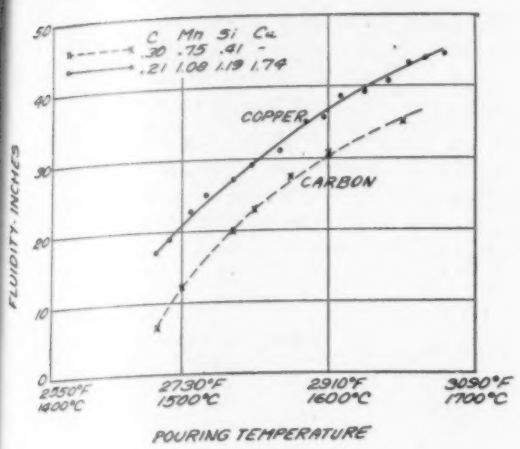


Fig. 14—Fluidity of a plain carbon and a copper steel as measured by fluidity spiral.

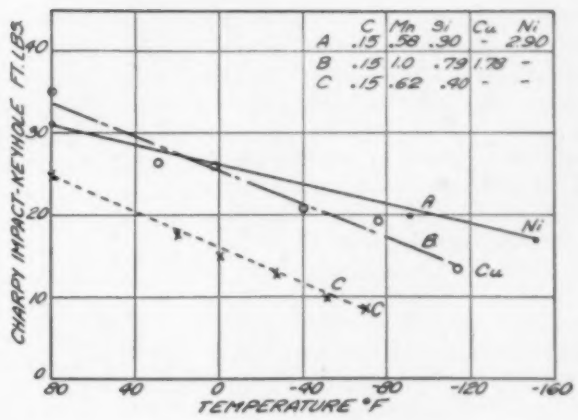


Fig. 15—Low temperature impact properties of plain carbon, nickel and copper steels. Heat treatment—double normalize and temper 1200° F.

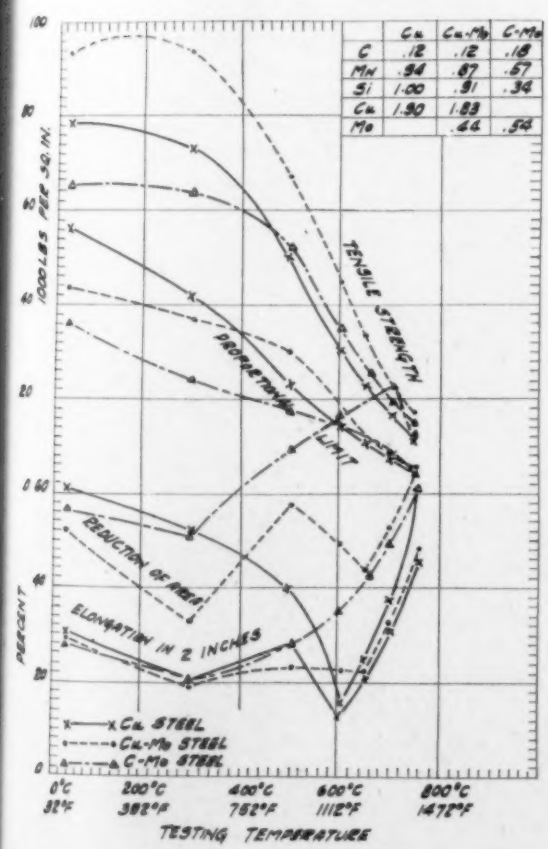


Fig. 16—Quick tensile properties of copper and copper-molybdenum steels after full anneal.

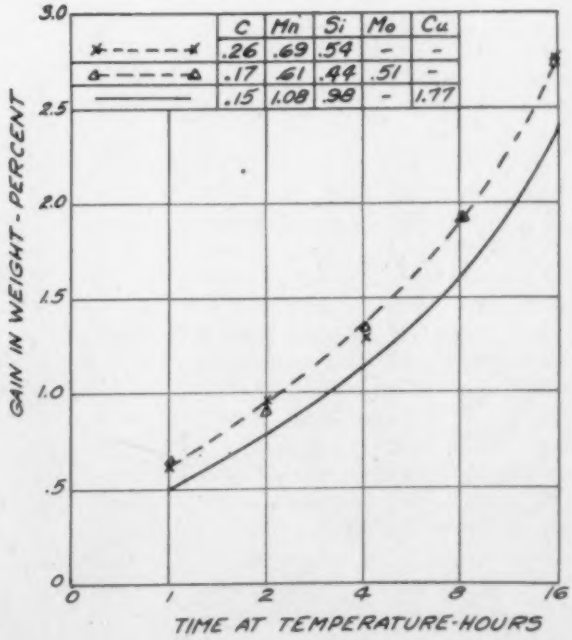


Fig. 17—Gain in weight of three steels due to oxidation in heat treat furnace at 1600° F.

obtain 1¾-in. strips and each strip was tested either in heat treated or "as-welded" condition, with no further machining. By comparing the properties of the welded specimens with those of the cast filleted specimens, the effect of welding on the base metal was determined. The welds were made with E 6010 type of electrode by using a current of 175 amps., an arc voltage of 26, and a speed of travel of 4 in. per min. All variables were automatically controlled.

Each 1¼-in. specimen was tested in the jig, as shown in Fig. 19. The specimen was held securely to prevent shifting and to cause bending to occur uniformly at each fillet as pressure was applied to the back of the specimen by means of the plunger operated by a hydraulic testing machine. The angle through which the specimen was bent at maximum load was measured. This angle is the supplement of the angle made by the legs of the tee.

The results of the tee-bend test are shown in Table 3. The angle at maximum load is generally accepted as a reasonable measure of weldability, and an angle of 45 degrees may be considered to be satisfactory. The results provide a comparison between plain carbon and copper steels.

Copper Steel Weldable

It is indicated that a copper steel containing 0.11 per cent carbon is quite weldable and can be used in the "as-welded" condition if necessary. This steel has an angle at maximum load in the "as-welded" condition of 55 degrees and, by tempering, the ductility of the specimen is raised to that of the annealed specimen with cast fillets.

Table 4

PHYSICAL PROPERTIES OF SOME LOW-CARBON COPPER STEELS IN NORMALIZED AND ANNEALED CONDITION

Heat No.	Composition, per cent C	Mn	Si	Cu	Heat Treatment	Yield Strength, psi.	Tensile Strength, psi.	Elongation, per cent in 2 in.	Reduction of Area, per cent
23	0.12	0.83	0.79	1.85	1706° F. (930° C.) 2 hr. Air Cool	68,500	84,500	26.1	59.6
24	0.12	0.83	1.48	1.85	1706° F. (930° C.) 2 hr. Air Cool	64,000	81,000	26.1	49.6
73	0.11	1.04	1.23	1.74	1742° F. (950°C.) 2 hr. Furnace Cool	64,000	81,000	31.0	56.0
33	0.08	0.77	1.00	1.95	1706° F. (930° C.) 2 hr. Furnace Cool	53,000	71,000	34.4	64.9

Table 5

PHYSICAL PROPERTIES OF WATER QUENCHED STEELS

Steel No.	Draw Temperature, °F.	°C.	Time, hr.	Yield Strength, psi.	Tensile Strength, psi.	Elongation, per cent in 2 in.	Reduction of Area, per cent	Izod, ft.-lb.	Brinell Hardness No.
1	1000	538	20	102,000	116,500	17.2	40.4	27	245
1	900	482	20	106,000	122,000	19.6	41.6	26	270
1	800	427	7	138,000	155,000	16.0	32.3	9	340
1	650	343	14	145,000	156,000	13.5	32.0	9	350
3	1200	650	7	117,000	127,000	18.5	49.2	32	255
3	1150	621	7	112,000	123,000	20.3	49.2	32	255
3	1075	580	7	151,000	158,000	16.4	43.0	20	—
3	1000	538	20	155,000	163,000	16.4	42.6	18	330
3	900	482	20	160,000	166,500	14.1	38.8	12	340
3	900	482	1	182,000	185,000	9.4	30.3	—	370
5	1075	580	7	111,000	126,000	17.6	39.5	29	255
5	1000	538	20	119,000	133,000	16.4	37.6	17	275
5	900	482	20	127,000	139,000	15.2	36.2	19	295
5	800	427	7	160,000	173,000	12.5	28.0	11	360

	Composition, per cent C	Mn	Si	Cu	Mo
1	0.17	0.92	1.14	1.94	—
3	0.15	0.90	1.13	1.75	0.54
5	0.25	0.91	1.19	1.71	—

Heat Treatment (all specimens)	Temperature, °F.	°C.	Time, hr.
Homogenize	2000	1093	7
Normalize	1700	930	2
Water Quench	1650	900	1

When the carbon in copper steels is 0.15 or 0.16 per cent, as in samples *D*, *H*, and *L*, it becomes necessary to stress relieve after welding in order to obtain satisfactory ductility. As welded, these steels have an angle at maximum load of about 30 degrees but, by stress relieving, it becomes approximately 60 degrees.

A 0.20 per cent carbon, copper steel can be restored to a ductility of 45 degrees bend by tempering at 1200° F. (650° C.) for one hr., but in the 0.24 per cent carbon steel the ductility of the welded plate was not restored to a suitable value even after tempering at 1200° F. (650° C.) for 4 hrs. The steels used in these tests contained more manganese and silicon than is normally used in cast steel, which would tend to impair the weldability of the steels due to their effect of increasing the hardenability.

However, it can be definitely concluded from these results, that copper steels containing less than 0.20 per cent carbon can be welded with no deleterious effect if given a stress relief treatment after welding.

Satisfactorily Welded

The fact that copper steels containing approximatly 0.10 per cent carbon can be welded satisfactorily without preheat or postheat becomes significant when it is also realized that remarkable physical properties are obtained from this alloy upon full-annealing or normalizing without any aging treatment whatsoever.

Thus, such a composition can be readily welded at tensile strengths and yield strengths above 70,000 and 50,000 psi., respectively. In Table 4, the physical properties of several steels which have approximately the foregoing composition are listed.

Water Quenched Copper Steels. Table 5 lists some properties of copper steels as quenched and tempered. All tensile blanks were 1x1x4-in. specimens and were heat treated in this form. An excellent combination of strength and ductility is obtained by water quenching and tempering copper and copper-molybdenum steels in one-in. sections, which indicates that these steels may be used in small gears and structural parts requiring yield strengths of over 100,000 psi.

Copper Steels Compared to Plain Carbon Steels. Fig. 20 shows the

range of properties that can be obtained in copper steels without liquid quenching. The lower limits of strength and the upper limits of ductility are the properties in the fully annealed condition; the other limits are the properties after aging to maximum hardness. In between the two extremes mentioned are the properties obtained at varying degrees of aging.

This graph (Fig. 20) indicates the extreme flexibility of copper steels and the excellent properties obtainable. The curves only extend to 0.25 per cent carbon, which is the maximum amount recommended for these steels. The optimum carbon is 0.10 to 0.20 per cent.

Obtainable Properties

Figure 21 is a graph taken from the *Steel Castings Handbook,* which shows the properties obtainable in plain carbon steels by quenching and tempering. The bands cover a tempering range from 1050-1300° F. (566-704° C.), and the range of carbon content of these steels is from 0.20 to 0.50 per cent. At the extreme left side of the chart the properties of copper steel containing 0.20 per cent carbon have been indicated in the normalized and aged condition.

The tensile strength of air cooled and aged copper steel is equal to that of the quenched and tempered 0.50 per cent carbon steel, and the yield point of the copper steel exceeds the yield point of the quenched and tempered carbon steel by more than 10,000 psi. The reduction of area and the elongation of the copper steel compare favorably with the corresponding properties of the carbon steels.

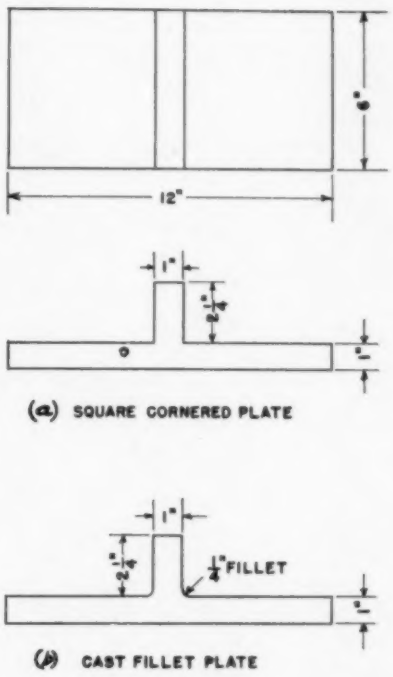
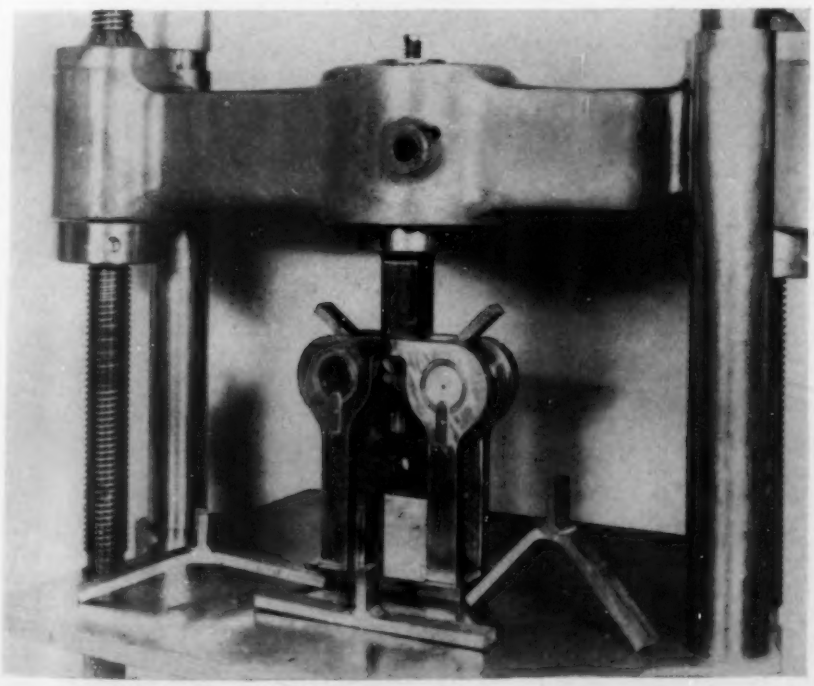


Fig. 18—Views of "Tee-bend" test plates—(A) square cornered plate and (B) cast fillet plate.

Fig. 19—Bending a "Tee-bend" specimen.



Conclusions

Steels containing 1.5 to 2.0 per cent copper, with silicon and manganese contents of approximately 1.0 per cent each, have excellent casting characteristics as proved by spiral fluidity tests and by production casting operations.

With copper steels, it is possible, by using a properly balanced composition and suitable heat treatment, to obtain strength values which can be attained in carbon steels only by raising the carbon content to the upper part of the medium carbon range and fully quenching and tempering.

Precipitation hardened copper cast steels have unusually high yield strength to tensile strength ratios. Such ratios are 0.8 or above as compared to normalized and tempered carbon steels at 0.6 or below, and normalized and tempered alloy steels at 0.6 to 0.7. Quenched and tempered steels have ratios of approximately 0.8, but such steels must be sufficiently alloyed to harden throughout the section being quenched.

By precipitation hardening, almost full strength is developed in copper steels at the center of fairly heavy sections. After normalizing and aging, the tensile strength at the center of an 8 in. diameter cylinder was only 15 per cent less than that at the center of a 2 in. diameter cylinder of the same material which was similarly treated. The corresponding decrease in yield strength was 12 per cent.

Heat Treating

Heat treating for precipitation hardening does not require critical time or temperature control. In the solution treatment, the copper is satisfactorily dissolved by any commercial austenitizing treatment such as is used for annealing, normalizing, or hardening, and is retained sufficiently at all rates of cooling equal to or faster than that which occurs upon air cooling of moderately heavy sections.

Aging for maximum strength requires somewhat closer control of time and temperature. For commercial practice, it is considered advisable to age at 900° F. (480° C.) for 5 hr. so that the danger of overaging is less than is encountered by attempting to age at higher tem-

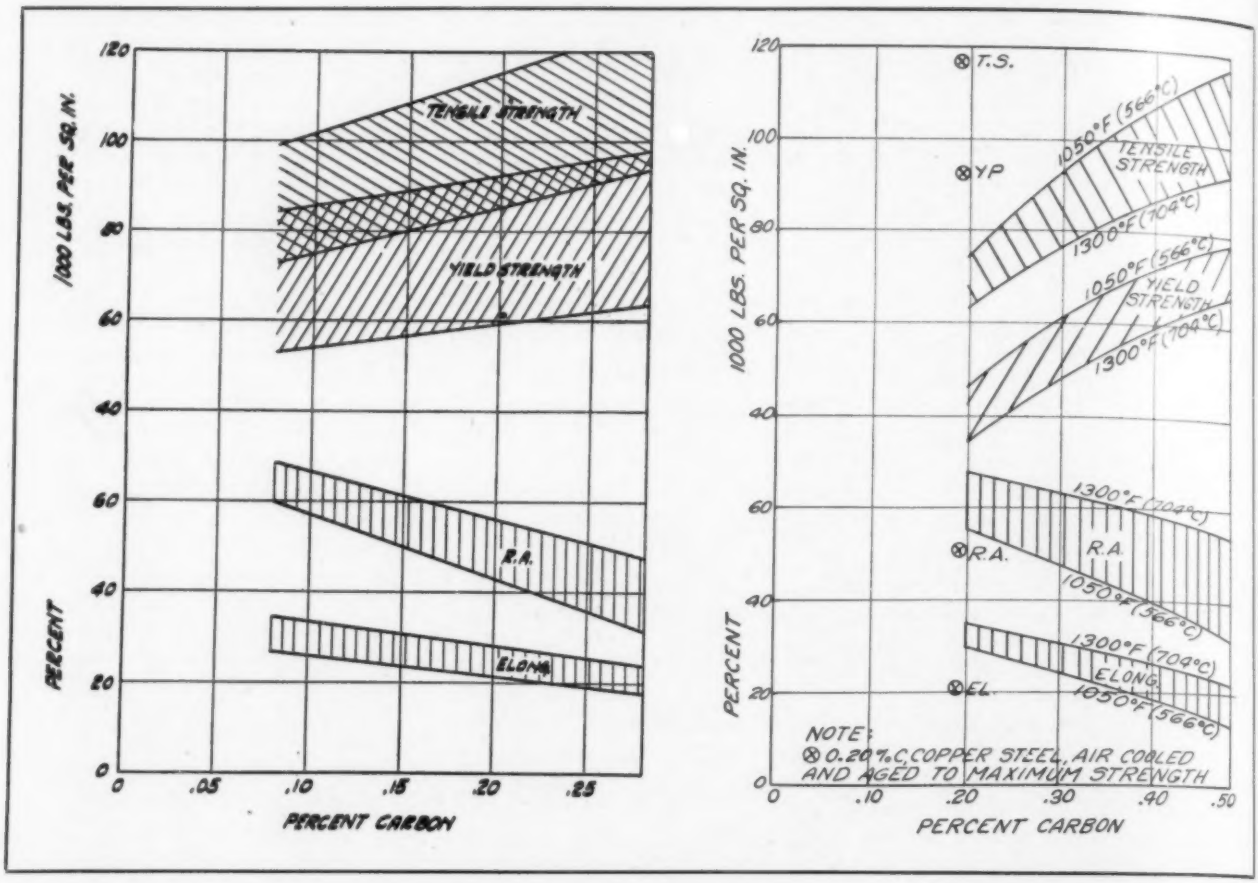


Fig. 20 (left)—Range of properties obtainable in copper steels without liquid quenching. Fig. 21 (right)—Range of properties obtainable in plain carbon steels in water quenched and tempered condition.

peratures for shorter periods of time.

Copper steels can be welded easily. When the carbon content is less than 0.15 per cent, weldments have sufficient ductility to be used in the "as-welded" condition; at approximately 0.10 per cent carbon, these steels have yield points above 50,000 psi. and can be welded without preheating or postheating. Above 0.15 per cent carbon, stress-relieving is necessary for satisfactory ductility.

For purposes of good weldability and development of optimum physical properties in copper cast steels, the carbon content should be low, preferably below 0.20 per cent.

Copper steels have excellent tensile properties at sub-zero temperatures in the normalized, normalized and aged, and normalized and tempered conditions. Likewise, they have less notch sensitivity at low temperatures in the normalized and tempered condition than do carbon steels of the same general composition and heat treatment condition.

It is indicated from short-time tensile tests that copper steels containing molybdenum are potentially useful for high temperature applications. Scaling tests show that copper steels are more resistant to oxidation at high temperatures than are carbon and carbon-molybdenum steels.

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ACCIDENT PREVENTION AS A FUNCTION OF MANAGEMENT

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This paper is fundamentally a review of some of the basic functions of management in the planning and execution of an effective accident prevention program. It is intended as an endeavor to assist in the strengthening and betterment of the foundry industry.

THOROUGH DISCUSSION of accident prevention, as a function of management, requires first the establishment of a broad and long range concept of its purpose and scope. If we deviate from the too commonly accepted notion of accident prevention as a minor undertaking concerned with the safeguarding of belts and gears, etc., and begin to consider it as a broader activity having to do with the welfare of workers, better utilization of manpower, improvement of industrial and public relations, we encounter some of the most challenging phases of management activity.

Accident prevention, planned and conducted on a broad scale, can contribute materially to the solution of many current managerial problems. It is the goal of all management to organize operations so that there is a minimum of injury to workers, interruption in production, wastage and misapplied effort. The significance and value of accident prevention, as a factor in creating and maintaining safe and efficient working conditions, has been fully demonstrated in recent years.

Merits of Plan

Many industrialists are squarely on record as to its merits and scores of efficient plants exist where an active, properly organized accident control program is a tangible factor in maintaining the smooth flow of production.

This paper was presented at a Joint Meeting, Plant Equipment and Safety and Hygiene Session at the Fiftieth Anniversary Convention of the American Foundrymen's Association in Cleveland, May 6, 1946.

All industrial accidents tend to do damage, disrupt, or destroy. There are, of course, variations in degree in the ultimate effects of accidents but basically they are all of negative value producing nothing of worth and invariably causing some type of harm, loss, or wastage.

Accident prevention can be applied to all types of industrial occupations when a recognizable hazard to life or property exists. Management has the prime responsibility in determining these hazards and in planning the corrective steps indicated. To obtain maximum accomplishment it, therefore, is necessary for management to establish a definite policy, to develop a program and to take an active part so that adequate and proper managerial stature is provided.

Success of Program

Management must encourage a broad conception of the benefits of accident prevention among supervisors and employees. The success of a program depends upon the understanding and cooperation of the persons involved. Simplified to an extreme degree, the prevention of accidents may be considered basically as an attitude of mind or a direction of interest. On this basis, it is necessary for management to take positive action in the creation of this attitude among all personnel.

Production in some form is the only purpose of any industrial undertaking. It is, therefore, essential to discuss accident prevention as it has a bearing on production. A safe place to work and a safe method of working is inherent in effectiveness. On this basis, management may well take the development of this relationship of accident prevention to production as a major project. As long as accident control is looked upon as a supplementary activity—an extra—to be super-imposed on production, it is cumbersome to apply and difficult to carry on successfully.

Safe Working Conditions

Integrated accident prevention has substance and practicality and it can prove its worth to supervisors and workers. It can be built into the layout or arrangement of a foundry, included in the design of a piece of shop equipment or made a part of a method. This approach brings it into contact with the problems of shop operations and provides a frame work on which to develop safe and efficient working conditions essential to the best interests of everyone concerned.

If this relationship is not accepted, worthwhile and continuing accident prevention is not likely to be achieved. In its place, a "hit or miss" program develops. Situations are apt to arise where there is false or seeming conflict between production and accident prevention requirements. In the majority of instances, it is not difficult to work out a best method from both production and safety standpoints. In some cases, however, a great deal of effort and ingenuity is required.

It is not idealistic to say that to an increasing degree the selection of a production method, where alternatives exist, is being made more commonly on the basis of the least hazardous. Management will be

called upon to make decisions of this kind and, in so doing, will give evidence in a very concrete manner of the significance placed on accident prevention.

Many near or minor type accidents often take place before the right combination of circumstances occur which set the stage for a serious accident. The correction of the cause of a near or minor type accident is, therefore, one of basic factors in accident prevention. The cause of a minor type accident will often be disclosed as a defect in a production method.

Illustration Cited

There are countless illustrations of this situation. Assume for example, that a group of workers are using wheelbarrows—moving loose dry material to a mixer. In one section of distance travelled there is a faulty floor condition. Workers after a few trial runs lay a loose plank over this section. The workers are obliged to exercise special effort to cross the plank.

In the course of time a wheelbarrow slips off the plank and is overturned. There is no injury to worker —no accident in one sense of the word—but a certain amount of productive effort was lost in reloading the spilled material and cleaning up. Several near slips take place and then after a time one worker stumbles, as his wheelbarrow gets out of control on the plank, and he wrenches a leg severely.

The faulty floor condition was as much a "production hazard" as it was an "accident hazard." Repairing the floor was no more an accident prevention step than it was one of production maintenance.

To carry on an accident prevention program, management will ordinarily delegate some portions of its responsibility and authority. This means that management must organize and make assignments in the same manner that supervision is planned for other shop procedures. The type of the organization depends primarily on the size of the foundry but, regardless of the size, there must be specific assignment of responsibility for safe conditions.

There must be manpower available to carry on the program. Each person involved should understand that his part in the program is a significant portion of his duties and will require a definite share of his time and effort.

Management's part in the plan of operation, as always, should be to furnish the overall guidance, stimulation and direction necessary to keep the program active and efficient.

As an outline to be used in evaluating an existing program or as a guide in developing a new one, the following items may be taken as covering the basic requirements of a complete program:

1. Sustained management support and direction.
2. Adequate safety organization.
3. Active participation by shop supervision.
4. Effective employee education.
5. Control of accident hazards:
   a. Safeguarding of machines and equipment;
   b. Establishment of safe methods and good practices;
   c. Control of occupational disease exposures—where they exist;
   d. Provision of personal protective devices—where no other form of safeguarding is practical.
6. Regular inspections—to detect unsatisfactory working conditions.
7. Investigation of accidents—to make certain that corrective measures are taken.
8. Proper first-aid facilities.

Minimum Requirements

These eight items are actually minimum requirements. There are supplementary features which have a direct or indirect bearing on accident prevention that might well be added. The eight items, however, have the most immediate bearing on the general problem and are the factors to be considered first and foremost. They provide a base on which to plan and build a sound program.

The American Foundrymen's Association, through the activity of its Industrial Hygiene Codes Committee, has developed several codes of recommended good practices which are comprehensive guides for the furthering of practices in the Foundry Industry. These codes have been carefully worked out to meet the actual requirements of the Foundry Industry and as codes are outstanding in their completeness.

They are more than codes in the conventional sense in that they contain definite recommendations for specific conditions and, as far as accident prevention is concerned, they can properly be taken as authoritative standards. The American Foundrymen's Association has performed an outstanding service to the Foundry Industry in the preparation and sponsorship of these codes.

In Section IV "Safety Practices" of the code, "Recommended Good Safety Practices for the Protection of Workers in Foundries," there is a complete outline for use of management to evaluate an existing accident prevention program or to develop a new one. The writer highly recommends wider use of A.F.A. Codes prepared by the Industrial Hygiene Codes Committee.

After management has adopted a policy and has established an adequate plan and organization for conducting an accident prevention program—it must continue to furnish guidance and stimulation.

Operating supervisors, foremen and sub-foremen should be frequently contacted by management concerning progress being made. It is essential that it be demonstrated regularly that management feels accountable for the success of the program. Shop supervisors are, in the long run, the key men in getting the specific details worked into shop operations. If there is slackening of management's concern, these men are the first to feel the effect.

In some foundries, safety departments or safety committees are doing worthwhile work. However, in many instances, these groups, because of a lack of authority, are only partially effective. Whenever a safety group is established, management should make certain that sufficient authority is assigned to carry out the work delegated to it.

Psychological Condition

Once such a group begins to fail in accomplishment, a poor psychological condition arises whereby its efforts are lessened and its standing or prestige among foremen and workers becomes of little consequence. Management must make sure that when it delegates responsibility in accident prevention, it also delegates sufficient authority to achieve results.

Management would be wise to plan in some real way to take part in the major routine accident prevention procedures which are put in operation. For example, it is important that management be present or be effectively represented at all meetings arranged as a part of the program. Such evidence of support and interest will go a long way in making these meetings worthwhile.

Accident Data Reviewed

Accident experience data must be reviewed frequently so that pertinent information can be passed along. It is essential that poor accident experience and reports or investigations of accidents be discussed thoroughly with supervision as current operating problems. When the record is good, it is equally important to recognize the good work of those responsible so that interest and cooperation will be kept alive and alert.

Reports of shop inspections, if made, should be reviewed and followed up by management. The prime purpose of inspections is to detect unsafe conditions so corrective measures can be undertaken. Prompt action, however, by management in remedying unsatisfactory situations disclosed by such inspections is another means whereby its interest in accident prevention is expressed in tangible form. Suggestions for the betterment of shop practices are oftentime encouraged by forward looking managers.

There are other ways, of course, which management can use definitely to associate itself with the actual working out of an accident prevention program—awards, contests, bonus plans and bulletins, but regardless of the method, if it serves to convince the shop personnel that management is in back of the program, it is bound to succeed.

A sizable percentage of reported foundry accidents are the most common types found in all industry—trips, falls, strains, bumps and bruises—resulting from hazards arising from poor housekeeping conditions and unsafe methods of handling materials.

As a sort of dual project it is recommended that good housekeeping and good materials handling methods be used by management as the framework on which to create improved foundries conditions. Poor housekeeping and inadequate handling methods invariably increase the number of exposures to accident hazards and the minor hindrances to production.

Housekeeping alone can accomplish a great deal in accident prevention in that it acts as a sort of catalyst. Once improved housekeeping is started, it stirs up interest in better maintenance and improved arrangement. An orderly foundry is well on the way to being a safe and efficient work place.

Management has a problem in planning for good housekeeping. There are several phases to be reviewed in setting up a program—storage facilities, maintenance, house cleaning procedures, waste disposal, etc. If such factors are not considered with the idea of simplifying housekeeping, housekeeping may become unnecessarily costly. As management takes steps to establish improved housekeeping methods, it becomes evident to all concerned that a real importance is attached to this part of shop practices.

Housekeeping Procedures

Many foundries today have established excellent housekeeping procedures. It would, no doubt, be found that it took considerable time to develop these programs. I believe it would also be found that as the programs were worked out the interest and cooperation of workers increased steadily. Housekeeping is an activity in which everyone working in a foundry can take part. It depends basically on selling the idea that orderliness produces benefits for all. Management will succeed in this task to the degree it demonstrates its faith in the worth of good housekeeping by providing and maintaining means and methods of carrying it on effectively.

The control of dust, fumes and smoke may be taken as a part of housekeeping although its significance is such that it is often very properly singled out for special treatment.

The methods of control required, of course, depend upon type of operation but, in any event, management is obligated to evaluate the significance of the exposure and to organize the control procedures. A comprehensive and adequate dust control setup in some foundries involves considerable effort and expense but it produces commensurate benefits in improved working conditions.

At this point it is probably in order to discuss the more or less common conception of a foundry as a "tough" place in which to work. There are, of course, legitimate methods of becoming dirty and tired in most industrial plants and a foundry is no exception.

In the past, in many foundries it was generally accepted that working conditions were on the rugged side and no real attempt was made to lighten things. The significant factor to consider is that in the "rugged" element there are some factors such as heavy manual lifting, pushing, and pulling which can be replaced by labor-saving methods or equipment with marked advantage from both production and accident prevention standpoints. Mechanization is not necessarily restricted to complete mold and sand handling units.

In certain shops some hard surfaced gangways and a lift truck would handle the bulk of the heaviest work. A few lengths of roller conveyor or a span or two of monorail in other instances would eliminate the worst strain hazards in the shop. In the foundries where mechanized handling units are not warranted or adaptable, management is faced with a challenge to utilize miscellaneous and partial approaches to reduce the amount of heavy manual labor required.

Materials Handling

From an accident control angle, materials handling is a major problem in many industrial undertakings. Where manual effort is required to handle heavy loads regularly, there is bound to be an outstanding hazard. Over a period of time the inherent hazards of such lifting are sure to result in accidents.

In some instances definite limitations exist with respect to the extent to which labor saving devices can be warranted based solely on accident prevention considerations. But here again, the close relationship of accident prevention and efficient production methods will oftentimes provide the solution. By studying the two factors together, the benefits of reduced handling costs and elim-

ination of accident hazard will be found adequate to justify the indicated action.

The following items describe situations where investigations of materials handling methods originating from accident prevention approach are indicated:

1. When an analysis of accident occurrence shows that there is a *high frequency of accidents arising from handling of a particular type* of material or object (kegs, bags, flasks, patterns, pig iron, scrap, castings).

2. When *more than one person is regularly required to lift* or move an object and there is severe potential hazard (of strains, bruises, burns) which might result in an accident owing to human failure, unsafe practice or improper working conditions.

3. Where *one person is required to lift* or move regularly a load that is in excess of 70 lb. for a man or 25 lb. for a woman. (These figures of 70 lb. and 25 lb. are not intended to have any authoritative significance as to what constitutes a safe maximum weight for anyone to lift. The weights were taken only as a starting point with the idea that some top limit be established as a point at which on regular lifts improved methods would be considered. The speed or frequency, distance of transport, shape of object, position and reach height of lift are all factors which must be taken into consideration in analyzing the hazards accompanying a particular lift.)

Material in Process

4. Where *congestion of material in process* is causing unsafe working conditions with accompanying accident hazards. It is often possible to clear up cluttered areas by providing equipment which makes it possible to use remote storage areas, to handle material in process in smaller lots with less floor area consumed at work stations for temporary storage, in disposal of scrap and waste, etc. Improved housekeeping often may be obtained only by better materials handling procedures.

5. Whenever there is excessive *manual rehandling* of an object between process steps. As the amount of handling is reduced, the exposure to accidents is cut down. The simple ideas should be studied first—use of platforms and benches so that material can be kept off the floor, platform skids (for hand or power lift trucks) so units of material can be moved about easily and kept ready for transport between process operations, etc.

This procedure of "spotting" accident hazards in materials handling is similar in approach to that of job analysis, methods planning, and motion study in straight production work. When it is adopted and used, hazards are cleared out along with certain production interferences. In many foundries, management will find that improved housekeeping coupled with application of better materials handling will turn the accident frequency curve downward.

Accident Probability

The probability of an accident arising from some particular operation conceivably could be eliminated entirely by 100 per cent safeguarding, in other words, as the saying goes, the job made "fool-proof." However, the bulk of the hazards in foundries cannot be handled in this manner. The complete safeguarding of workers depends rather on a combination of safe conditions and safe practices. Safe working conditions are primarily concerned with tangible physical conditions which can be approached as an engineering proposition. This is a relatively simple problem with a definite beginning and an end.

The establishment of safe practices is more involved. It might be described as an undertaking in human engineering. While one of the inherent instincts of man is that of self-preservation, it appears that under certain industrial activity, an indifference to certain type hazard develops which ultimately is the fundamental cause of an accident. Safe practices, therefore, fundamentally are concerned with the creation of patterns of thinking to build up defenses against this indifference and lack of understanding.

In any apprentice or training program, therefore, it is essential that accident prevention be stressed as fundamental factor in becoming an efficient foundryman. The hazards of the business must be pointed out and, at the same time, the safeguards necessary for the worker and his fellow workers must be explained and sold as the right way of working. If a hazard is known and can be anticipated, it can be controlled and accidents prevented.

Management must, therefore, exercise a continuous scrutiny of shop practices to detect unsafe methods of working. Supervision has the front line responsibility. Shop foremen must be on the alert for unauthorized short cuts, improvisations and failures to follow approved procedures. It is often improper to make a charge of carelessness when a man is injured because of a failure to follow good practice.

There is invariably a much sounder reason—lack of understanding of hazard, lack of instruction or a misconception of the ultimate benefit to himself and to his employer in accident prevention effort. Management, as a step to foster safe practices should, therefore, be on record with approved methods of carrying on hazardous operations—shop rules for the guidance of all supervisors and workers.

Another major responsibility of management in the control of accidents is the proper placement of individual workers. There are usually basic physical requirements for a job or an occupation and if qualified persons are assigned to these jobs, the probability of accident is reduced. A definite review of placement routine is in order in any size foundry. Physical capabilities should be matched up with job demands.

Job Analysis

To obtain maximum benefits, it is often desirable to develop a procedure of physical demand job analysis. With such analyses available and a method of utilizing them, it is possible to place workers with a degree of certainty that there will be a minimum of misplacement and accompanying liability to accident.

The foundry industry looks forward to a future in which technical developments in the casting of metals will make possible an outstanding contribution by the industry to the industrial growth of the nation. This result, however, is dependent upon manpower for actual fulfillment.

The industry must, therefore, interest and attract men to carry on the work to be done. This is a real and tangible problem which is being

given an increasing amount of attention at present. It is a responsibility of management to take action to make certain that the foundry industry becomes better known for what it can be, an interesting, profitable and satisfactory place in which to earn a livelihood.

Accident prevention, in that it is a major factor in the creation and maintenance of good working conditions, can be made to play an important role in this undertaking. The benefits in employee welfare will tend to effect a new public notion of foundry work. An "open house" such as some foundries have held, provides a splendid opportunity to acquaint potential employees and other interested people with modern practices.

Accident prevention is not a static or restricted activity if considered in light of its potentialities. New approaches sometimes carry accident prevention a considerable way beyond the conventional acceptance of safety. For example, a program involving industrial preventive medicine offers much. Workers in good physical condition are better prepared to work safely and effectively.

Placement procedures, taking into account the physical conditions of workers, can be expected to produce worthwhile results in the control of accidents. Health maintenance, instructions in nutrition, and after work safety are factors in the handling of absenteeism. These and similar projects are all of long range concern to management in assuring the safety and well being of those employed in our foundries.

In conclusion, it can be said that accident prevention has proven its worth and is generally recognized as a function of management. To be effective, it must be based on acceptance of its real significance and organized as an integrated part of shop operations. It requires the understanding and participation of all shop personnel. Successful accident prevention conserves human life and skills, expedites production and promotes good working conditions. It tends to create good morale among workers and in so doing is of interest and concern to the public.

For the ultimate good of the foundry industry and the welfare of those engaged in it, management may properly expect to achieve lasting gains through its efforts in the prevention of accidents.

DISCUSSION

Chairman: JAMES THOMSON, Continental Foundry & Machine Co., East Chicago, Ind.

Co-Chairman: C. P. GUION, W. W. Sly Mfg. Co., Chicago.

D. TAMOR[1]: While it seems that in most cases management is definitely sold on safety, what means would you adopt to combat human failure on the part of workers and the general violation of safety rules?

MR. MEIGS: As I get it, management is sold but they cannot sell the employees.

MR. TAMOR: Not quite that. In 99 per cent of the cases, management is sold, but you do have human failures. You might set up a rule on the use of leggings, spats, or goggles and nine men wear them and one man will not and that one man who will not wear them demoralizes the other nine. How can you get away with a selling program? What would you do if you happened to be Assistant Foreman or Foreman of a Pouring Gang?

MR. MEIGS: The whole point is that if management really accepts safety, they must work to get it across. For years it has been recognized that an educational program, a program of stimulation, a program of educating people in safety, was a part of a complete safety program. Now if you have the situation where one man does not wear leggings and nine do, I would say you had an easy problem. The most common situation would be where you had much less cooperation than that. You have the thing boiled down so it is a small problem to work on that one man.

MR. TAMOR: What do you do with that one man?

MR. MEIGS: Without getting mixed up in labor relationships, I know of places where they would not let him work. It depends upon how important safety is considered. If you do not want to let him disrupt your whole program, you must stop him.

MR. TAMOR: But you may have a labor relations problem there.

MR. MEIGS: You may have to switch around to a program of leading them up the sawdust trail. I do not think that situation is going to last very long. I think as safety becomes more generally accepted, in the shops where they have put on an earnest program they have been astounded by the success they have had.

MEMBER: Where you have 99 per cent of the people observing safety rules and one man is careless, I have found it effective to take that one person off to the side and have a heart to heart talk with him. That way you reduce that carelessness. I have found that procedure very effective.

G. R. CARPENTER[2]: It might help to get the unions to endorse the safety program. You can put more pressure on the fellow who does not cooperate. In many cases, the union will endorse the program or will go along on the general principle, and in that event the non-cooperator has no backing for not supporting it.

CHAIRMAN QUION: In fact, the unions are more or less insisting on safety being a part of the labor contracts as I understand it.

MR. TAMOR: Both expedients have been tried. That is, the man has been given a heart-to-heart talk and, in addition, we have had the cooperation of the union. The fact that 90 per cent of the men do abide by the safety rules and regulations is enough evidence that the union is back of us. But the I. Q. of the men in certain divisions of the foundry is so low that it is difficult to give them a sales talk. We actually have a replica of the eye up on the display board, with all its functions, and we try to give the man a story on how important it is to keep his eyes, but the problem still does exist.

The problem also exists that experienced pourers or weight shifters are difficult to get. If you in your foundry do not have that problem, you are fortunate; and if you have a good selling campaign, you are fortunate, too. At one of our plants, even a salesman could not get in without glasses. I was there recently and one man in the foundry grinding room did not wear goggles and I heard the foreman call up the safety director and ask him how far you can go with the rule. So you do have a labor relations problem there. You can not upset an organization and yet you can not stand the chance of disruption of a safety program.

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[2]Minneapolis-Moline Power Implement Co., Minneapolis.

EFFECT OF HEAT TREATMENT ON THE ENDURANCE LIMIT OF ALLOYED GRAY CAST IRON

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AS far as the author can determine, Moore and Picco[1] are the only ones who have reported any results of the effect of heat treatment on an endurance limit. They ran endurance limits both in reverse bending and in torsion on specimens taken from a 7-in. I.D. cylinder with a 1¼-in. wall thickness, both in the "as-cast" and in the heat-treated condition on a gray cast iron of the following analysis: total carbon, 3.07; silicon, 1.26; manganese, 0.90; sulphur, 0.08; and phosphorus, 0.15 per cent. The results that they obtained are shown in Table 1.

It will be seen that although they obtained a 58 per cent increase in tensile strength, there is only a 19 per cent increase in the endurance limit in reverse bending and a 30 per cent increase in torsion. The endurance ratio (endurance limit divided by tensile strength) is actually reduced both in bending and in torsion.

In the development of materials which would serve for cast crankshafts, the possibilities of obtaining high endurance limits by heat treatment were investigated. This paper covers this investigation.

Two crankshafts were cast of a processed cupola iron* of the following analysis: total carbon, 2.84; silicon, 1.40; manganese, 0.67; phosphorus, 0.11; sulphur, 0.059; nickel, 1.69; chromium, 0.16; molybdenum, 0.46; and copper, 0.15 per cent. (The chromium and the copper were not added but were residuals in the scrap. The nickel and molybdenum were added to the charge in the cupola and not to the ladle.)

One of these crankshafts was allowed to cool to room temperature in the mold. The other was shaken out of the mold when it reached 1600° F. and allowed to cool in still air to room temperature. The temperature at which the shaking out was done was determined by a thermocouple inserted in the mold cavity before pouring.

The crankshafts cast were of such a design that a number of test sections 6¾-in. O.D. by 3-in. I.D. by 9½-in. long could be cut. The 3-in. I.D. was a cored hole, which gave a metal section of 1⅞-in. thickness.

A sufficient number of test specimens were obtained from the crankshaft that cooled in the mold to carry on the various heat treatments given in the following. The shaking-out of the crankshaft at 1600° F. and allowing it to cool in air is considered as one type of heat treatment, so the results are included.

Table 1

ENDURANCE LIMITS FOR IRONS TESTED

Condition of Specimen	Static Tensile Strength, psi.	Endurance Limits, psi. — Bending**	Endurance Limits, psi. — Torsion**	Endurance Ratios* — Bending	Endurance Ratios* — Torsion
As cast	48,000	21,000	16,500	0.44	0.34
Heat treated†	76,500	25,000	22,000	0.33	0.29

*Ratio of endurance limit to static tensile strength.
**Completely reversed cycles of stress for unnotched specimens.
†Preheated slowly to 1000° F., then heated to 1600° F., quenched in oil and reheated to 1000° F.

Considerable information is available on the heat treating of gray cast iron to increase the tensile strength. However, the effect of this heat treatment on the endurance limit has been more or less a neglected subject. Furthermore, the effect of size of section and original microstructure on the final heat treated tensile strength and endurance limit is apparently an open field for investigation.

*All of the material used for this investigation was manufactured under license of The Meehanite Metal Corp.

This paper was presented at a Gray Iron Session of the Fiftieth Annual Meeting, American Foundrymen's Association, at Cleveland, May 10, 1946.

The standard practice in this foundry is to give a stress-relieving treatment at 1050° F. to all cast crankshafts. Both shafts, therefore, were given this treatment before the test specimens were cut from them. The "as-cast" condition used in this paper means that the pieces were stress relieved.

The following heat treatments, with the exception of heat treatment *D*, were given to the test sections obtained from the cast crankshaft.

Heat Treatment A. Heat slowly to 1600° F., hold 2 hr. Quench in oil. Reheat to 1050° F.—2 hrs. Cool in air.

Heat Treatment B. Heat slowly to 1600° F., hold 2 hr. Oil quench to 900° F. Hold at 900° F.—16 hr. Cool in air.

Heat Treatment C. Heat slowly to 1600° F., hold 2 hr. Oil quench to 650° F. Hold at 650° F.—16 hr. Cool in air.

Heat Treatment D. Cool in mold to 1600° F. Then strip from mold and air cool, then stress relieve at 1050° F.

Heat treatment *A* is the usual "harden and temper" treatment used to produce high tensile strength and high hardness. The tempering temperature of 1050° F. was given to approach the maximum hardness that could be machined.

Flinn and Reese[2], working with nickel molybdenum alloyed gray irons with total carbon of 2.50 per cent and silicon of 2.50 per cent, have shown that extremely high physical properties may be obtained if the matrix structure of the iron is acicular. They have also shown that this acicular structure is an austenite transformation product which is formed at temperatures of from 900° F. to 500° F. provided that the material is austenitic when it reaches this temperature range.

Some time must be spent in this temperature range to produce the desired structure. In other words, if the rate of cooling of the material is sufficiently fast to bring it past the nose of its isothermal transformation curve (S curve) and then held in the range of 500-900° F., an acicular structure will be formed.

Mold cooling is so slow that it is practically impossible to form an acicular structure in unalloyed gray iron. The addition of nickel and molybdenum shifts the nose of the curve to the right, giving a longer nose time. This allows the retention of the austenite down into the range of 500-900° F. where, if the mold cooling rate is satisfactory, the acicular structure will be formed.

Flinn and Reese[2] also have pointed out that the combined carbon in the matrix must be eutectoid composition to form a truly acicular structure. This range of carbon is 0.70-0.90 per cent. In order to produce this eutectoid composition in any given section, a proper balance between the total carbon and silicon must be maintained.

Heat treatments *B*, *C* and *D* are attempts to produce the acicular structure in the section cast and with the analysis previously stated, which by microscopic examination was proved to be almost eutectoid in composition. Since the formation of the acicular structure is really a function of the cooling rate, it was believed that this cooling rate could be controlled by heat treatment, and thus the acicular structure could be produced.

As no "S curves" were available for the composition of metal involved, oil quenching was considered to be sufficiently fast to cool past the nose of the curve. No method was available for determining the temperature at which to remove the test section from the oil and place it in the furnace for holding.

This temperature had to be estimated, so it is very possible that it was above or below the holding temperatures of 900° F. and 650° F. used. All that is certain is that both test sections were below 1000° F. when taken out of the oil.

The holding temperatures of

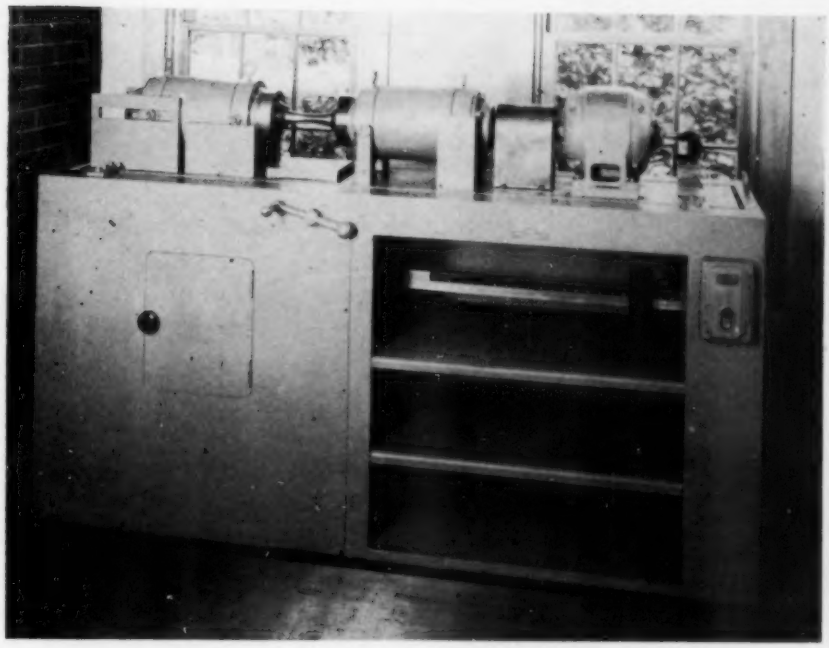


Fig. 2—Endurance testing machine capable of handling 1-in. diameter specimens.

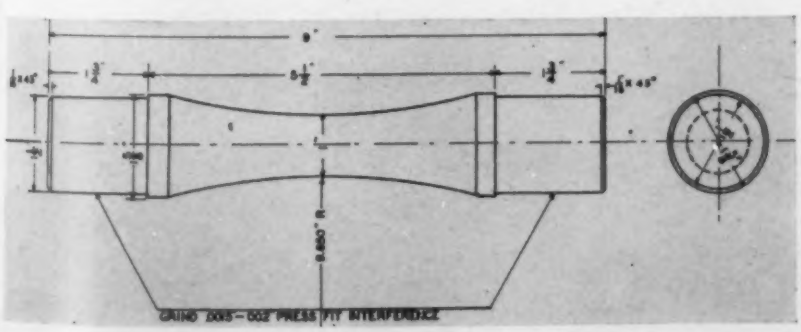


Fig. 1—Sketch showing design of endurance test bar.

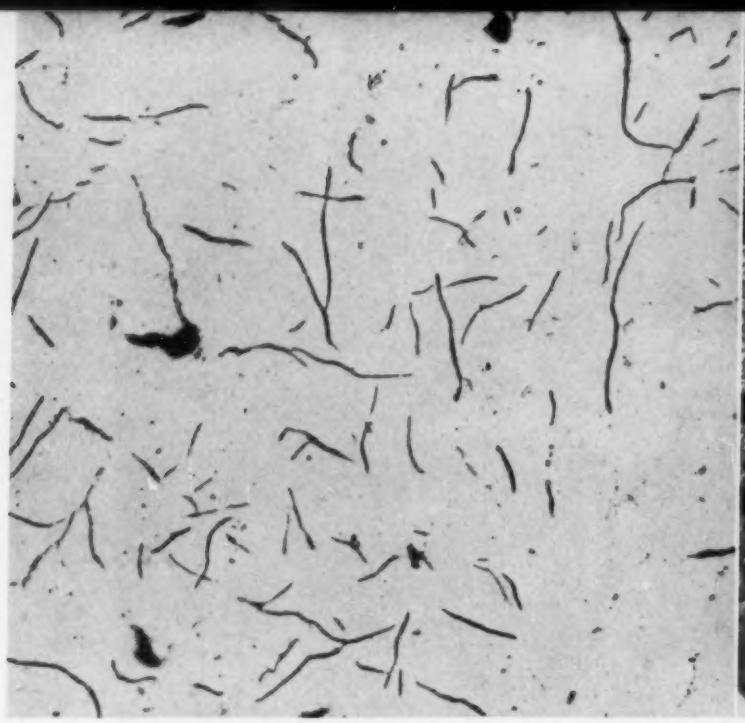


Fig. 3—Photomicrograph showing graphite size and distribution, Heat No. 4615. Unetched. 100X.



Fig. 4—As-cast microstructure of Heat No. 4615. Etched, hydrochloric and picric acids in alcohol. 1000X.

900° F. and 650° F. were arbitrarily chosen. Other temperatures would have been tried if there had been sufficient material to do so. The air-cooling treatment was one that had produced promising results on previous experiments, hence it also was tried.

Sufficient material was available in each test section to obtain tensile tests, impact test and endurance limit bars. Microstructure specimens were cut from the threaded end of the tensile test bars. Standard ASTM tensile bars of 0.800-in. diameter were cut from the test sections so that the center of the test bar was at the center of the section wall.

The endurance limit bars and impact test bars were also located in the same position. Therefore, all testing was done on approximately the metal which was located at the same position in the section thickness of the test section.

The endurance test pieces taken from the test sections were machined to the dimensions shown in Fig. 1. It will be noted that the diameter of the section tested is 1.000 in. instead of the smaller test pieces which are usually around 0.300 in.

All of the testing was done on a Baldwin Southwark endurance testing machine capable of handling the one-in. diameter specimens and testing in reverse bending. This machine is illustrated in Fig. 2. The ends of the specimen have collars shrunk onto them, which are, in

Fig. 5—Same as Fig. 4 except given heat treatment *A*.

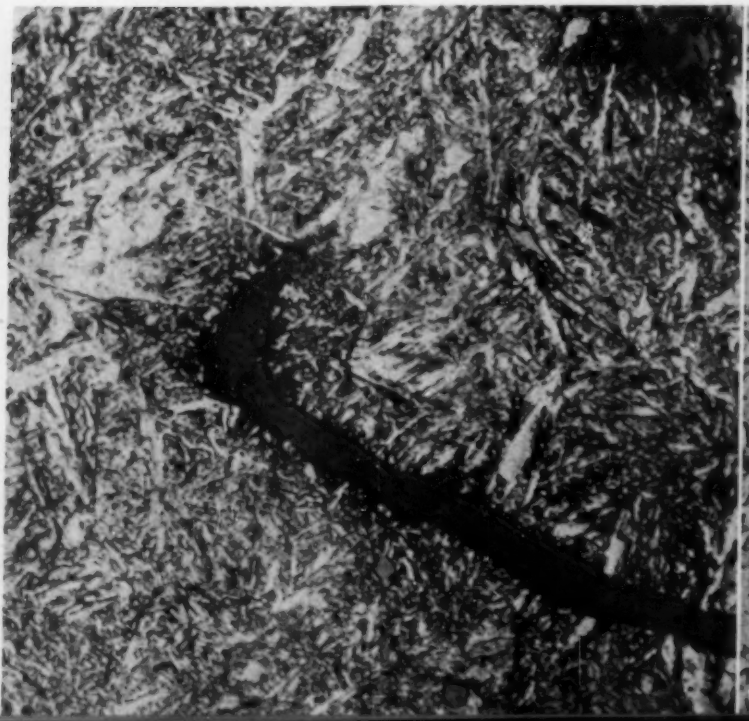
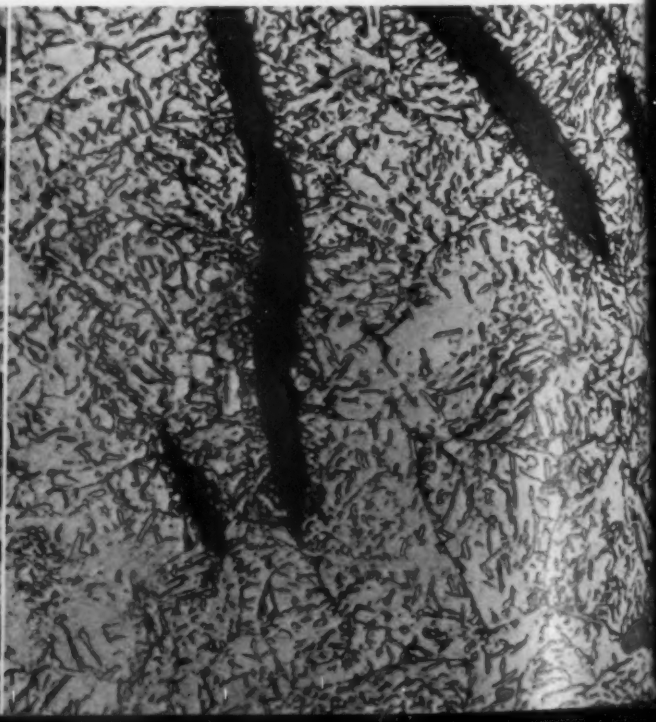


Fig. 6—Same as Fig. 4 except given heat treatment *B*.



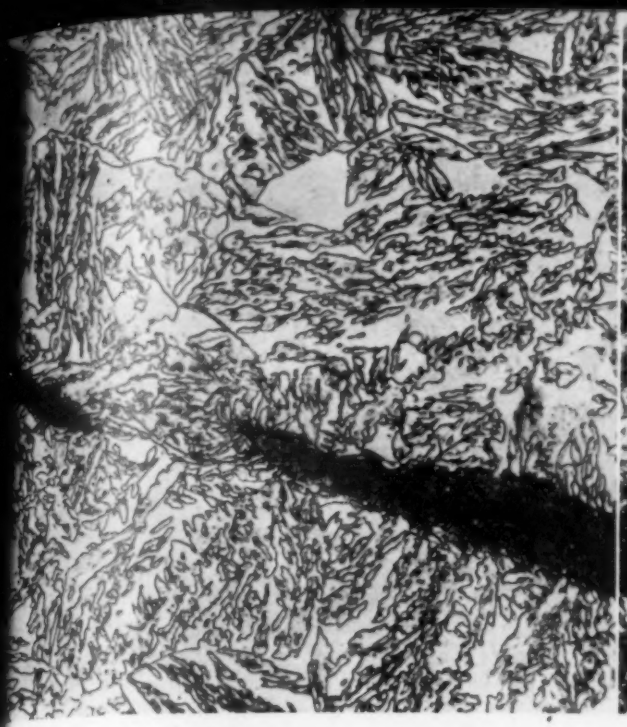


Fig. 7—Same as Fig. 4 except given heat treatment *C*.



Fig. 8—Same as Fig. 4 except given heat treatment *D*.

Table 2

RESULTS OF PHYSICAL PROPERTY TESTS IN HEAT NO. 4615*

Heat Treatment	*Tensile Strength, psi.*	*Brinell Hardness*	*Izod, ft.-lb.*	*Endurance Limit, psi.*	*Ratio, Endurance Limit: Tensile Strength*	*Tensile Strength Increase by Heat Treatment, per cent*	*Endurance Limit Increase by Heat Treatment, per cent*
As cast	50,500	217	18	17,500	0.35	.....	.....
A	66,600	311	16	18,000	0.270	31.9	2.9
B	63,300	248	15	17,500	0.28	25.4	None
C	71,000	255	21	22,500	0.32	40.6	28.6
D	66,000	248	18	21,000	0.32	30.8	20.0

*Chemical Analysis—T.C., 2.84; Si, 1.40; Mn, 0.67; P, 0.11; S, 0.059; Ni, 1.68; Cr, 0.16; Mo, 0.46; Cu, 0.15 per cent.

Table 3

ENDURANCE LIMIT TEST RESULTS ON HEAT NO. 4615

Heat Treatment	*Bar No.*	*Stress, psi.*	*Cycles for Fracture*	*Remarks*
As Cast	1	21,000	420,000	Broken
	2	18,000	652,000	Broken
	5	17,500	10,102,000	Not broken
	4	17,000	13,713,000	Not broken
	3	16,000	10,725,000	Not broken
		Endurance Limit—17,500 psi.		
Heat Treatment	1	18,500	1,387,000	Broken
A	5	18,500	1,312,000	Broken
	4	18,000	12,296,000	Not broken
	3	17,000	10,191,000	Not broken
	2	16,000	10,218,000	Not broken
		Endurance Limit—18,000 psi.		
Heat Treatment	1	20,000	829,000	Broken
B	2	18,000	3,172,000	Broken
	3	17,500	17,466,000	Not broken
		Endurance Limit—17,500 psi.		
Heat Treatment	2	25,000	375,000	Broken
C	4	23,000	619,000	Broken
	5	22,500	13,966,000	Not broken
	3	22,000	10,160,000	Not broken
		Endurance Limit—22,500 psi.		
Heat Treatment	3	22,000	586,000	Broken
D	5	21,500	1,830,000	Broken
	4	21,000	10,287,000	Not broken
	2	20,000	18,184,000	Not broken
		Endurance Limit—21,000 psi.		

turn, bolted to the spindles of the machine. The speed of operation of the machine is 3475 rpm.

The Izod impact used is one that has been recommended in England and employs an unnotched specimen of 0.798-in. diameter so held that it projects 1⅛-in. above the holding jaws. The standard 120 ft.-lb. Izod tup was used to break the specimen (see references 3 and 4 for a complete description with dimensions of this test).

This type of impact test has been temporarily adopted in the author's laboratory. Sufficient information for the evaluation of this test is not as yet available; however, the results obtained are included because it is felt that they at least indicate relative trends.

A complete micrographic study was made of all the sections tested. Only pertinent photomicrographs are reproduced in this report.

The results of the physical property tests, including the endurance limit obtained on the various test sections, are shown in Table 2. The endurance test results are shown in Table 3. The ratio of endurance limit to tensile strength has been calculated, as well as the percentage increases in tensile strength and endurance limit caused by heat treatment (also shown in Table 2).

The photomicrographs of the as-cast structure are shown in Figs. 3 and 4. Those obtained in the heat-

Table 4

TEST RESULTS FOR HEAVY SECTIONS (2¼-IN. THICKNESS) NI-CR-MO ALLOYED GRAY IRON

Heat No.	Chemical Analysis, per cent T.C.	Si	Mn	P	S	Ni	Cr	Cu	Mo	Condition	Tensile Strength, psi.	Endurance Limit, psi.	Brinell Hardness	Ratio, Endurance Limit: Tensile Strength	Increased by Heat Treatment, per cent Tensile Strength	Endurance Limit
3364	3.06	1.40	0.91	0.11	0.071	1.54	0.30	.....	0.49	As cast	46,800	16,000	197	0.34	.....	.....
3364	3.06	1.40	0.91	0.11	0.071	1.54	0.30	.....	0.49	1600° F.—2 Hr. (Oil) 1130° F.—2 Hr. (Air)	65,400	15,000	255	0.23	39.8	—6.2
3397	3.09	1.74	0.93	0.068	0.051	1.40	0.28	.....	0.58	As cast	52,000	17,000	207	0.33	.....	.....
3397	3.09	1.74	0.93	0.068	0.051	1.40	0.28	.....	0.58	1600° F.—2 Hr. (Oil) 1025° F.—2 Hr. (Air)	79,750	15,500	293	0.20	53.3	—8.8
3913	2.94	1.32	0.99	0.089	0.055	1.39	0.30	.....	0.50	As cast	49,000	16,500	217	0.34	.....	.....
3913	2.94	1.32	0.99	0.089	0.055	1.39	0.30	.....	0.50	*	58,500	16,500	277	0.28	19.5	None

*1600° F.—10 hr., quenched in oil to 900° F. Held at 900° F.—16 hr. Air cooled.

treated condition are shown in Figs. 5 to 8, inclusive.

The as-cast condition of the metal was given a thorough investigation. The graphite size and distribution, Fig. 3, is a random flake (A.F.A. GRAPHITE CHART, Type A, Size 4). The matrix structure is almost pure pearlite, Fig. 4, but an occasional spot of ferrite could be detected and one such spot is shown in the photomicrograph.

Examination at lower diameters showed the matrix to have a slight amount of primary cementite. The photomicrograph of this is not shown in this paper. This structure was considered to be almost truly pearlitic.

The tensile strength of the as-cast billet was 50,500 psi. and the endurance limit was 17,500 psi., which gave a ratio of endurance limit to tensile strength of 0.347. This ratio is somewhat lower than usually shown for gray iron, but it is within the range the author has usually experienced with sections of this size.

The normally recommended treatment for gray iron to improve the tensile strength is to oil harden and temper. That was done, of course, in heat treatment *A*. It will be noted that this treatment increased the tensile strength from 50,500 psi. to 66,600 psi., an increase of 32.2 per cent.

However, this increase is not reflected in the endurance limit, which in this case was raised from 17,500 psi. to 18,000 psi., or 2.9 per cent. An increase is also noted in Brinell hardness. The impact resistance is lowered from 18 to 16 ft.-lb.

The microstructure obtained by this heat treatment *A*, as shown in Fig. 5, is what may be considered sorbo-martensite and is typical of the structure obtained by such a heat treatment of an alloyed gray iron. Higher tempering temperatures would change this structure to sorbite with a corresponding drop in hardness and tensile strength. However, no further testing was done along this line because of lack of material.

Heat treatment *B*, which is one attempt to obtain an acicular structure by allowing the transformation to take place at 900° F., increased the tensile strength to 63,300 psi., an increase of 25.4 per cent; however, the endurance limit remained the same, 17,500 psi.

The microstructure for heat treatment *B* is shown in Fig. 6. This structure is what Flinn and Reese[3] call acicular ferrite. The impact resistance is lowered by this treatment. In fact, this was the lowest impact value obtained in any of the heat-treatments.

Heat treatment *C*, where the

Table 5

ENDURANCE LIMIT TEST RESULTS ON HEATS NO. 3364, 3397 AND 3913

Heat No.	Heat Treatment	Bar No.	Stress, psi.	Cycles for Fracture	Remarks
3364	As Cast	2	18,000	843,000	Broken
		3	17,000	1,187,000	Broken
		4	16,500	5,507,000	Broken
		6	16,250	4,805,000	Broken
		1	16,000	10,101,000	Not broken
			Endurance Limit—16,000 psi.		
3364	1600° F.—2 hr.—Oil 1130° F.—2 hr.—Air	1	20,000	380,000	Broken
		2	19,000	370,000	Broken
		3	17,000	1,272,000	Broken
		5	16,000	1,685,000	Broken
		6	15,500	1,541,000	Broken
		4	15,000	10,532,000	Not broken
			Endurance Limit—15,000 psi.		
3397	As Cast	2	20,000	1,175,000	Broken
		3	18,000	3,378,000	Broken
		6	17,500	1,663,000	Broken
		5	17,000	11,439,000	Not broken
		1	16,000	10,249,000	Not broken
			Endurance Limit—17,000 psi.		
3397	1600° F.—2 hr.—Oil 1025° F.—2 hr.—Air	1	18,000	802,000	Broken
		2	16,000	3,039,000	Broken
		5	15,500	10,003,000	Not broken
		4	15,000	10,070,000	Not broken
		3	14,000	10,342,000	Not broken
			Endurance Limit—15,500 psi.		
3913	As Cast	1	20,000	726,000	Broken
		2	18,000	1,317,000	Broken
		4	17,000	2,634,000	Broken
		5	16,500	10,086,000	Not broken
		3	16,000	10,213,000	Not broken
			Endurance Limit—16,500 psi.		
3913	1600° F.—10 hr.—Oil cooled to 900° F., held 16 hr.—cooled in air.	1	20,000	696,000	Broken
		2	18,000	1,110,000	Broken
		4	17,000	2,666,000	Broken
		5	16,500	10,006,000	Not broken
		3	16,000	10,156,000	Not broken
			Endurance Limit—16,500 psi.		



Fig. 9—Photomicrograph showing graphite size and distribution, Heat No. 3364. Unetched, 100X.

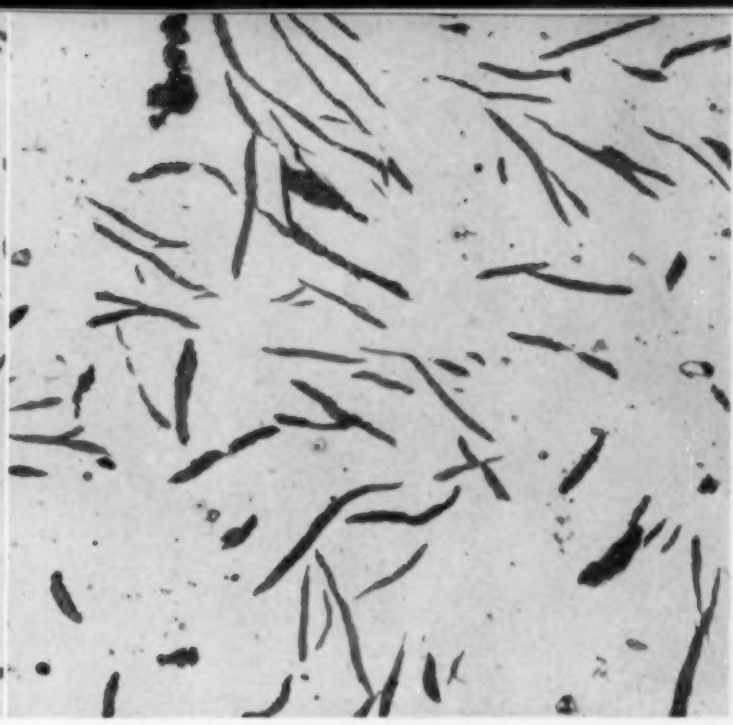


Fig. 10—Photomicrograph showing graphite size and distribution, Heat No. 3397. Unetched, 100X.

transformation was allowed to take place at 650° F., gave a tensile strength of 71,000 psi., an increase of 41 per cent. The endurance limit was 22,500 psi., an increase of 28.6 per cent. The Brinell hardness was 255, which is lower than the hardened and tempered material. An impact value of 21 ft.-lb. is the highest impact value obtained.

The microstructure for heat treatment *C* is shown in Fig. 7. This structure is not purely acicular but shows patches of free ferrite present.

Heat treatment *D*, which is an air cool from 1600° F., tempered at 1050° F., gave 66,000 psi. tensile strength; an increase of 30.8 per cent. The endurance limit was increased to 21,000 psi., an increase of 20.0 per cent. The Brinell hardness was greater, being 248, but the impact resistance was exactly the same as in the as-cast condition.

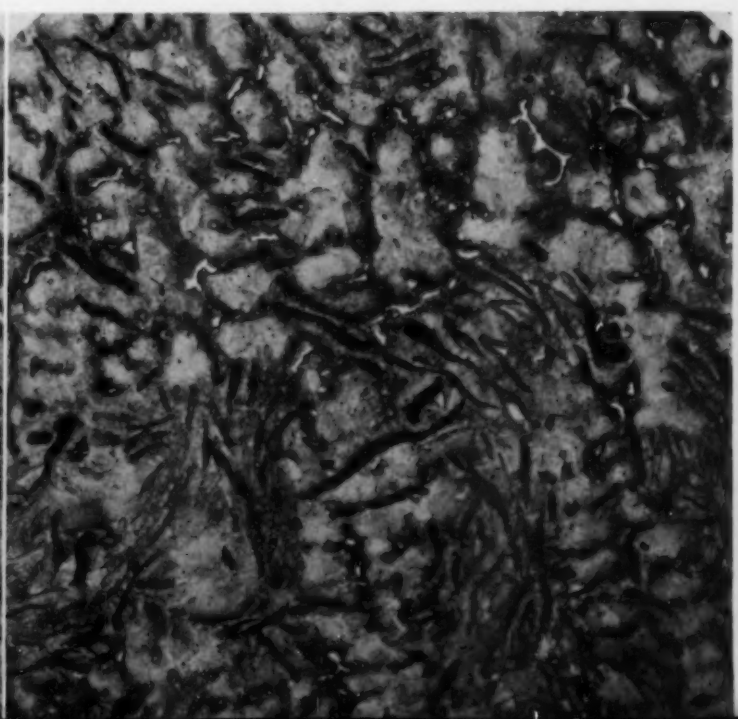
The microstructure obtained for this heat treatment is shown in Fig. 8. It will be seen that this structure is pearlitic with some patches of the acicular structure. However, the pearlite is extremely fine, hardly being resolved by the microscope. It resembles an as-cast structure that is frequently observed in sections much thinner than those tested.

All of the foregoing testing was done with one analysis of material with one basic (as-cast) microstructure. This material was chosen for the test because the microstructure was pearlitic. However, it must be acknowledged that conclusions based on such a test may be misleading

Fig. 11—As-cast microstructure, Heat No. 3364. Etched, hydrochloric and picric acids in alcohol. 50X.



Fig. 12—Heat No. 3397—Structure of as-cast material. Etched, hydrochloric and picric acids in alcohol. 50X.



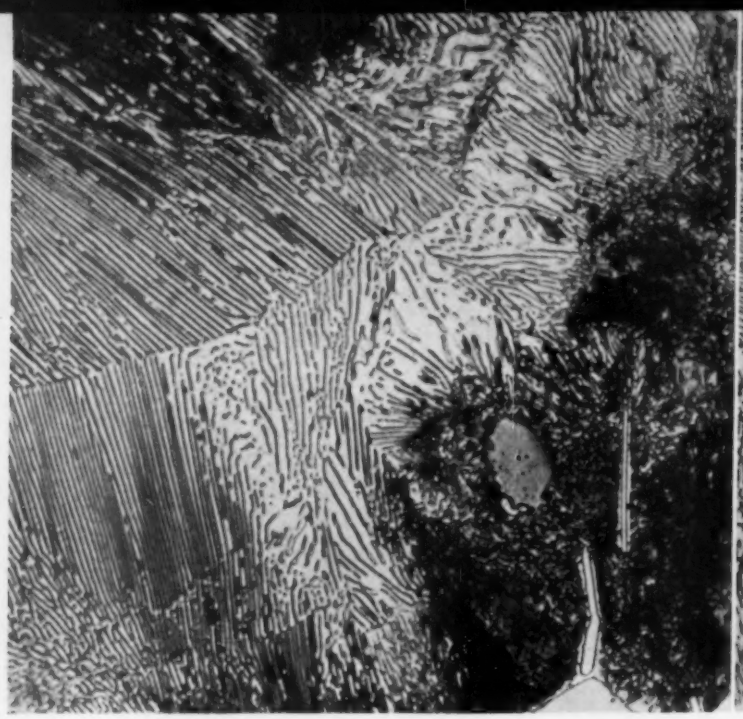


Fig. 13—Heat No. 3364—Same as Fig. 11 except 1000X.



Fig. 14—Heat No. 3397—Same as Fig. 12 except 1000X.

ing. With this in mind, further testing was carried on using material not so ideal in basic structure.

Three test sections were obtained from three experimental crankshafts cast from three different heats of a nickel, chrome, molybdenum alloyed gray iron. The size of these test sections was 8-in. O.D. by 3½-in. I.D. (cored hole) by 9½-in. long. The section thickness was therefore 2¼-in. The castings were mold cooled and, because they were cast for other experimental work, they were not treated to relieve stresses.

The heat treatment, the analysis, tensile strength, endurance limit and Brinell hardness of each of these test sections is shown in Table 4. The endurance test results are shown in Table 5. It will be noted that testing was done in both the as-cast and in the heat-treated condition. The testing procedure previously described was also used for this test, including a complete metallographic study.

It will be noted that two of the test sections, Heats No. 3364 and 3397, were given the normal harden and temper treatment, the difference being in the tempering temperature used. The special heat treatment given to Heat No. 3913 will be discussed separately.

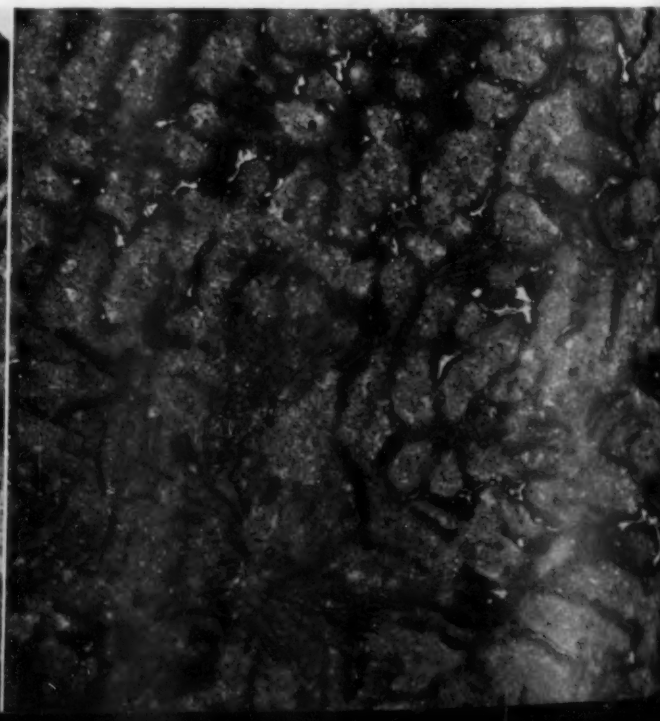
It will be noted from Table 4 that heat treatment increases the tensile strength of Heat No. 3364 from 46,800 psi. to 65,400 psi., an increase of 39.8 per cent, but the endurance limit has been reduced from 16,000 psi. to 15,000 psi., a reduction of 6.2 per cent.

In the case of Heat No. 3397, heat treatment raised the tensile

Fig. 15—Heat No. 3364—Structure of heat treated material. Etched, hydrochloric and picric acids in alcohol, 50X.

Fig. 16—Heat No. 3397—Structure of heat treated material. Etched, hydrochloric and picric acids in alcohol, 50X.





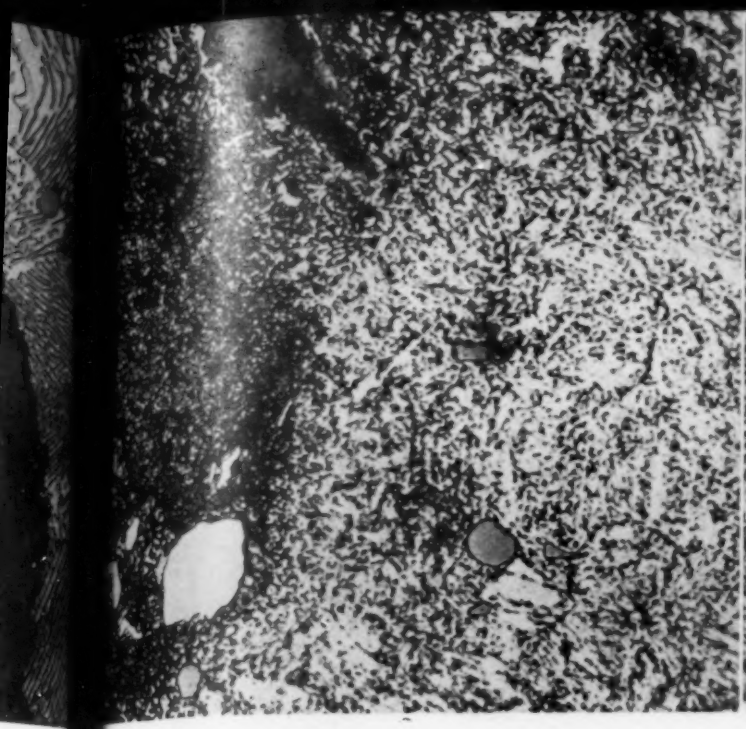


Fig. 17—Same as Fig. 15 except 1000X.

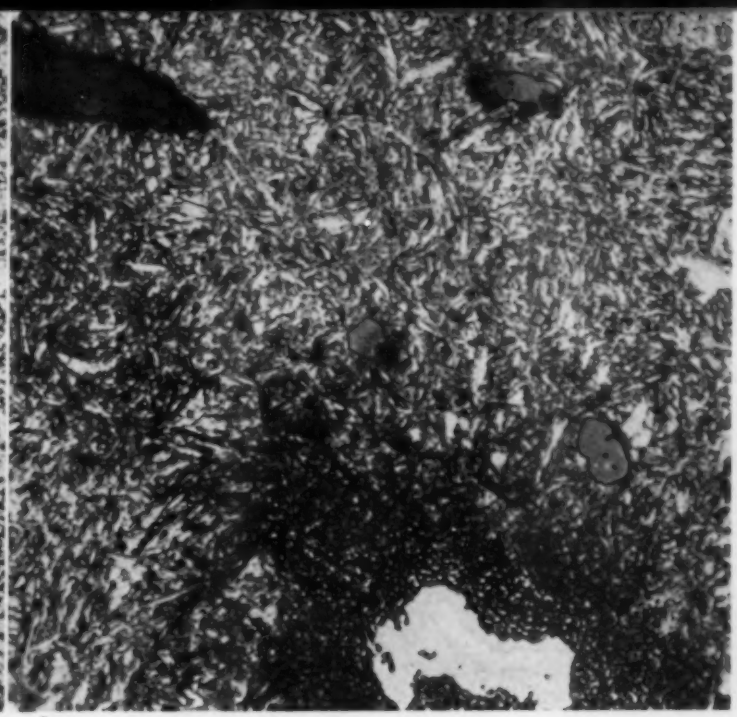


Fig. 18—Same as Fig. 16 except 1000X.

strength from 52,000 psi. to 79,750 psi., an increase of 55.5 per cent; but the endurance limit is reduced from 17,000 psi. to 15,500 psi., a reduction of 8.8 per cent. In these two cases, hardening and tempering gave a very substantial increase in tensile strength; but there is an actual reduction in endurance limit.

In the case of Heat No. 4615, previously described, the harden and temper treatment did not reduce the endurance limit; instead, a very slight increase was experienced.

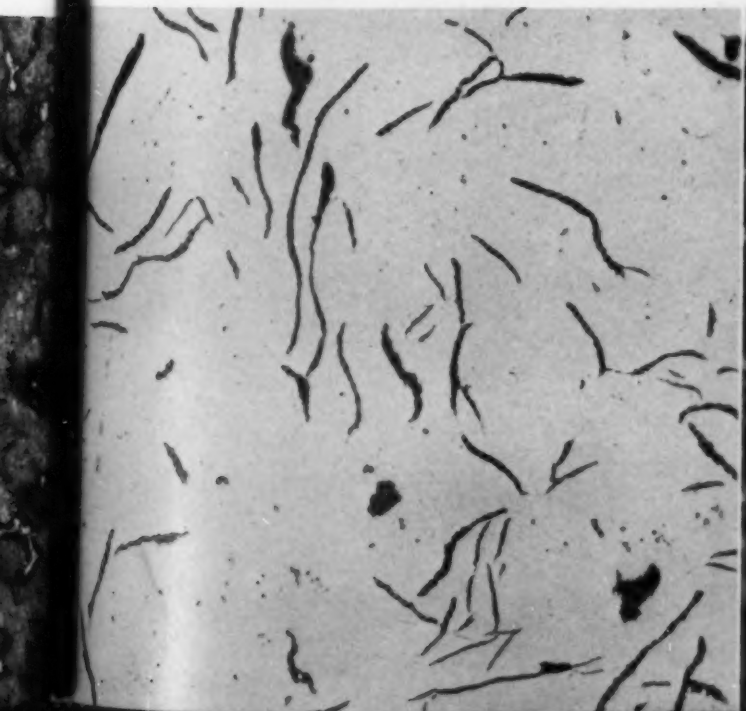
A complete metallographic study was made of these two heats. Figure 9 shows the graphite distribution of Heat No. 3364 and Fig. 10 shows that of Heat No. 3397. The graphite distribution is random flake (Type A) in both heats. Heat No. 3364 shows approximately 60 per cent Size 3 and 40 per cent Size 4, and Heat No. 3397 shows 40 per cent Size 3 and 60 per cent Size 4 when compared to the A.F.A. GRAPHITE CHART.

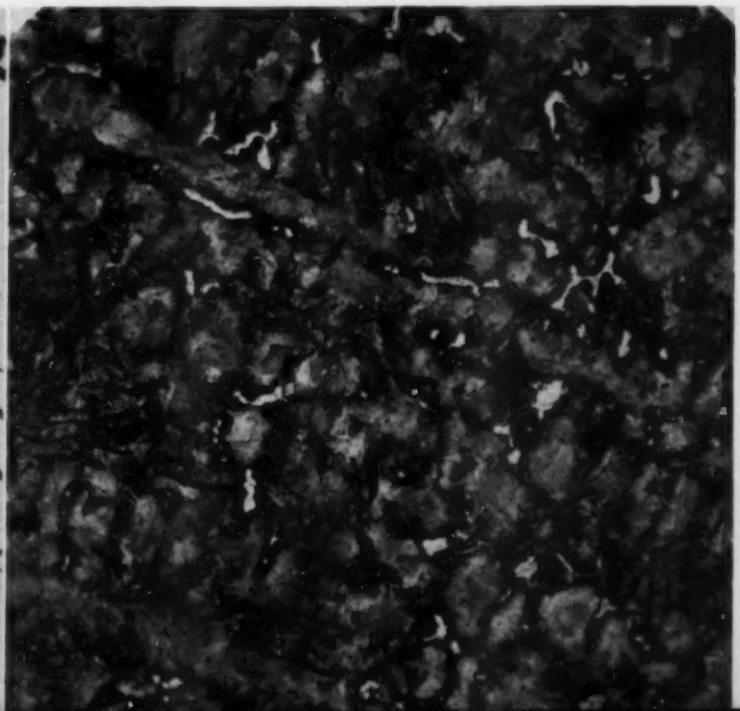
Figures 11 and 12 show the etched structures of these two heats at 50 diameters. It will be noted that both of them show considerable amounts of primary cementite. It also will be noted that the matrix shows what, for want of a better name, may be called a pseudo-dendritic pattern.

It is believed that this pseudo-dendritic pattern is an indication of combined carbon segregation; that is, areas higher in combined carbon etch darker than those of lower carbon content, thus revealing the original dendritic pattern of the metal. When carbon is diffused more evenly throughout the

Fig. 19—Photomicrograph showing graphite size and distribution. Heat No. 3913. Unetched. 100X.

Fig. 20—Heat No. 3913—Structure of as-cast material. Etched, hydrochloric and picric acids in alcohol, 50X.





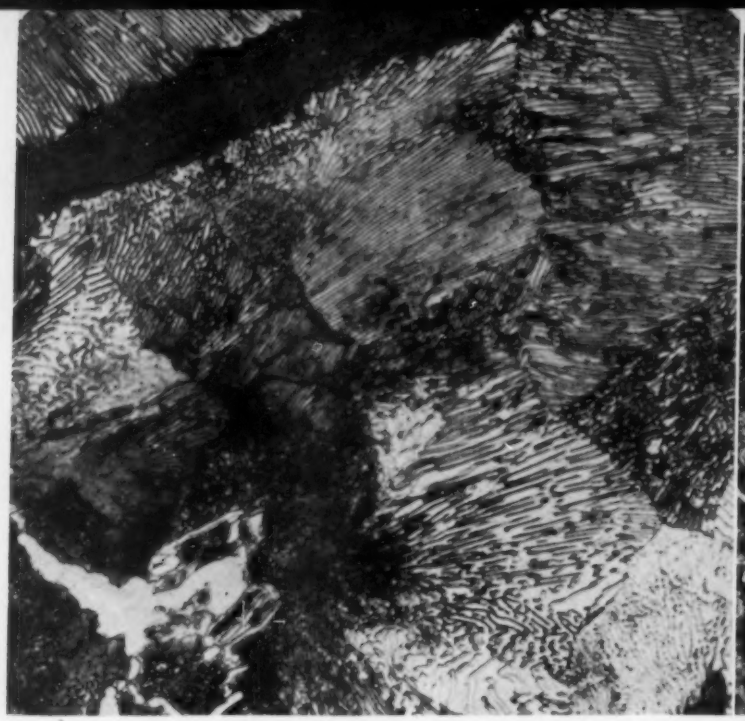


Fig. 21—Heat No. 3913—Structure of as-cast material. Etched, hydrochloric and picric acids in alcohol, 1000X.

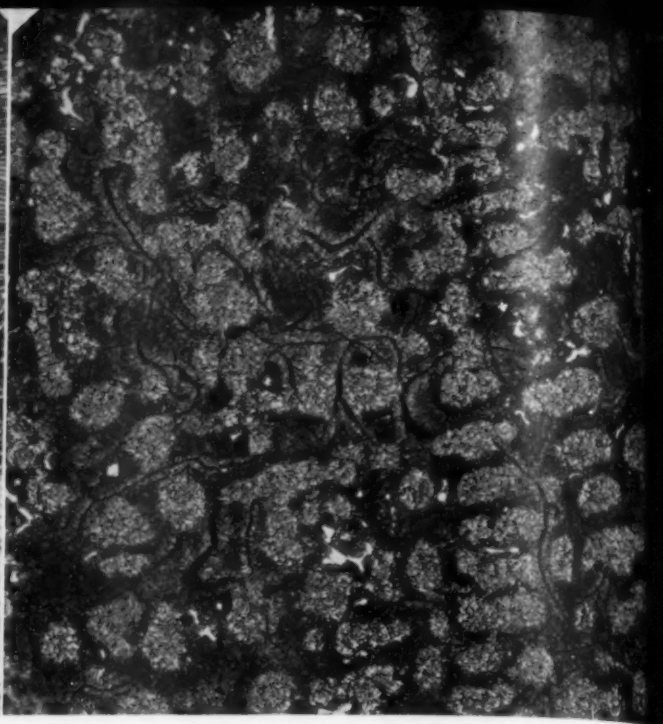


Fig. 22—Heat No. 3913—Structure of heat-treated material. Etched, hydrochloric and picric acids in alcohol, 50X.

matrix, no such pattern can be seen.

Figures 13 and 14 show the structure of the matrix at 1000 diameters. It will be noted that the matrix is pearlitic, but in the case of Heat No. 3397 the pearlite has begun to spheroidize. Attention is called to the structure around the primary cementite particles, which is pearlitic, but etches darker than the rest of the matrix. This is believed to be an indication of carbon segregation.

The structure of the matrix at 50 diameters after heat treatment is shown in Fig. 15 for Heat No. 3364; and in Fig. 16 for Heat No. 3397. It will be noted that the pseudo-dendritic pattern has been accentuated by the heat treatment.

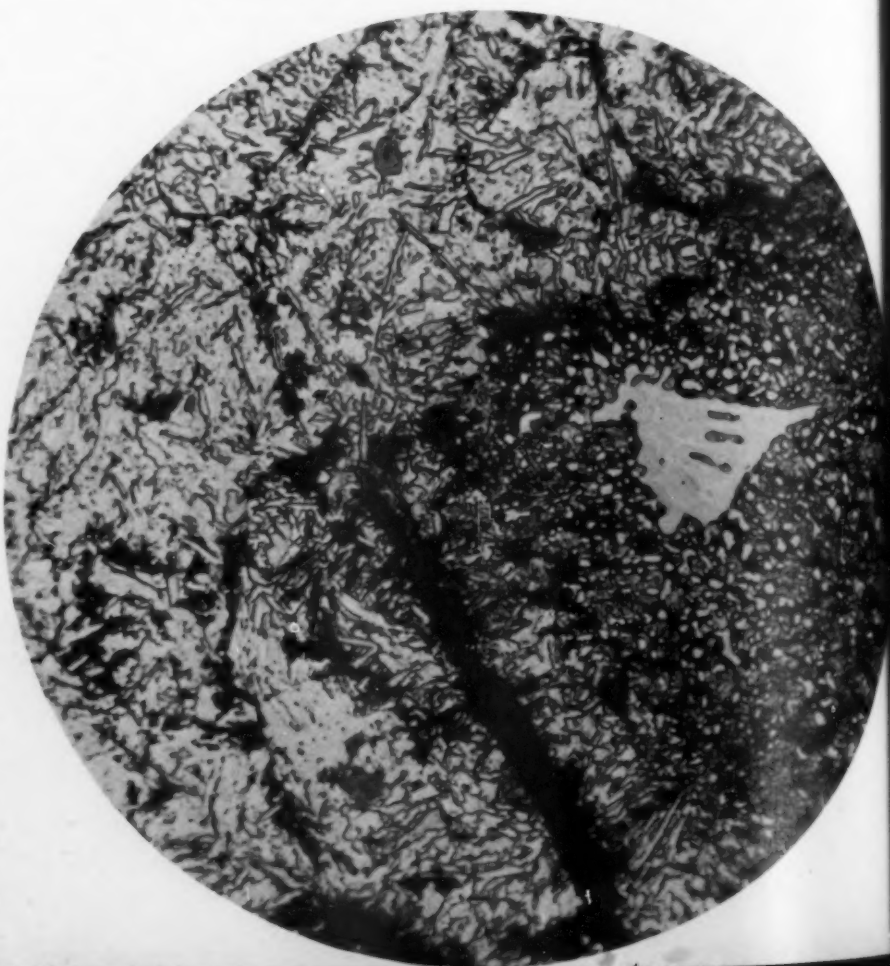
The structure of the matrix at 1000 diameters for Heat No. 3364 is shown in Fig. 17, and for Heat No. 3397 in Fig. 18. After heat treatment, Heat No. 3364 is sorbitic in structure; while Heat No. 3397 may be called sorbo-martensite, somewhat like the structure of Heat No. 4615 (Fig. 5).

Attention is again called to the difference in structure around the primary cementite particles. This structure is entirely different from the rest of the matrix and is believed to be due to the carbon segregation around the cementite particles.

The heat treatment given to Heat No. 3913 was an attempt to break up the cementite particles and allow a better diffusion of the carbon throughout the matrix. Holding gray iron at 1600° F. for 10 hr. is not normally recommended, as graphitization of the matrix usually takes place, lowering the combined carbon content, with subsequent lower physical properties.

The holding at 900° F. was an attempt to form the acicular structure. The results obtained were believed to be of sufficient interest to include them in this report of tests.

Fig. 23—Heat No. 3913—Same as Fig. 22 except 1000X.



It will be observed that the tensile strength has been increased from 49,000 to 58,500 psi., an increase of 19.5 per cent, but there is no increase in endurance limit. The Brinell hardness has been increased from 217 to 277.

The microstructures obtained are of interest. Figure 19 shows the graphite distribution which is approximately the same as that of other materials used for this later test. Figure 20 shows the structure of the as-cast material at 50 diameters. This again shows primary cementite particles and the pseudo-dendritic structure.

Figure 21 shows the as-cast structure at 1000 diameters. Again it will be seen to be pearlitic, with darkened areas around the primary cementite particles. Figure 22 shows the heat-treated structure at 50 diameters. It will be observed that the long holding time at 1600° F. has not completely broken down the cementite particles.

The pseudo-dendritic pattern has been broadened, which is believed to be caused by the diffusion of the carbon through the matrix. The light areas are acicular ferrite, which can be seen better at a higher magnification, Fig. 23.

Structure Gradations

This photomicrograph shows an interesting transition; the outer matrix is acicular ferrite, but there are several gradations of this through the truly acicular to an almost martensitic structure around the cementite particles. It will also be noted that there is a considerable amount of spheroidization of the primary cementite.

Any conclusions that may be drawn from the data offered in this paper must necessarily be confined to the nickel-molybdenum and nickel-chrome-molybdenum alloyed gray cast irons. They must also be confined to section sizes which have cooling rates analogous to those used for the test.

Furthermore, because of the many types and methods of heat treatment available today, conclusions must be confined to the heat treatments used. It also is realized that seven endurance tests are not necessarily sufficient to definitely prove the points offered. Nevertheless, it is believed that trends may be indicated. With this in mind the following is offered:

1. That, in general, heat treatment to obtain high tensile strength does not give a proportionate increase in endurance limit. This is shown by Moore & Picco[1] and confirmed by the work offered here.

2. That the normal harden and temper heat treatment to attain high tensile strength does not necessarily produce any increase in endurance limit. Under certain conditions the endurance limit may be lowered by such a treatment.

3. That the original as-cast structure has a profound influence on the endurance limit obtained after heat treatment. Unfavorable structures in which the combined carbon is segregated, apparently can give a lower endurance limit than the original as-cast condition.

4. That the nickel-molybdenum and nickel-chromium-molybdenum alloyed gray irons in which no great amount of free cementite is present, can be heat treated to the acicular structure with an increase in endurance limit; however, the increase is not in proportion to the increase in tensile strength.

Acknowledgment

The author is indebted to George Johnstone, foundry superintendent, for providing the material to carry on this investigation; to W. R. McCrackin for the photomicrographic work; and to C. F. Moore for his work in conducting the endurance tests.

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DISCUSSION

Chairman: H. BORNSTEIN, Deere & Co., Moline, Illinois.

Co-Chairman: V. A. CROSBY, Climax Molybdenum Co., Detroit.

DR. R. A. FLINN[1] *(written discussion):* Mr. Eagan's paper provides an important correlation of fatigue properties with heat treatment and tensile strength. It indicates strongly that increase in tensile strength by heat treatment does not necessarily improve fatigue or impact properties.

In the first set of experiments, higher endurance limit was *only* obtained when the heat treatment was adjusted to provide fine acicular structures at the 650° F. level or by accelerated cooling of the casting. Tempered martensitic structures, as obtained by the conventional quench and draw, or coarse acicular structures of the 900° F. level showed negligible improvement.

The experiments do not include the fatigue properties to be expected from as cast, fully acicular specimens or tempered, cast acicular structures. We realize that the alloy content required to produce a fully acicular structure in the heavy castings of this experiment might be an objectionable expense but we would like to indicate that better properties might be obtained. In lighter sections, the use of alloy in place of high temperature heat treatment is generally preferable.

Some improvement might have been obtained if the acicular structures were tempered after air cooling to room temperature in treatments BC. The groundmass of the acicular structure always has some retained austenite which is best removed by tempering after cooling to room temperature. We assume the reheating to 1050° F. was considered necessary in A and D; a lower draw temperature might have given better properties in general.

The second set of experiments using 0.30 per cent chromium iron with nickel and molybdenum indicate the deleterious effect of free carbide upon fatigue strength after heat treatment. Since the material of the first experiments contained 0.16 per cent chromium and some free carbide could this small amount also be operative in reducing heat treated fatigue strength previously discussed? An alternative solution is that the particles may be the complex molybdenum phosphide discovered by Rote.

As matters of general information we would like to know the reason for the selection of the large fatigue specimens, the correlation of results between large and small specimens, and any available combined carbon analyses of the specimens.

J. E. COLLINS[2]: I would like to have more detailed explanation on the surface finish of the fatigue specimens. Were they simply machined surfaces? There is some thought that is not necessary to polish cast iron.

MR. EAGAN: As to the finish of the specimens, they were all polished. That is true that a lot of people think it is not necessary to finish a cast iron specimen because it is not notch sensitive but these specimens were polished. They were not longitudenally polished. We used 2/0 metallographic paper with oil. The finish was considered good.

J. H. SCHAUM[3]: You place much blame on this dendritic structure for not being susceptible to increases in endur-

ance limit when heat treated. I do not believe that in your heat treatment you obtained a complete acicular structure. Is it possible to get a completely acicular structure in this dendritic type iron? If you did get it, would it not show a corresponding increase?

CO-CHAIRMAN CROSBY: I wish to congratulate Mr. Eagan on a well prepared paper on a subject that adds materially to our fund of information concerning gray cast iron.

I was somewhat of the opinion that better ratios (tensile to endurance) could have been obtained with an iron which was more homogeneous in structure, i.e., acicular plus graphite, instead of acicular plus carbide plus ferrite plus graphite. It has been our experience that structures exhibiting such heterogeneity are usually disappointing in physical properties.

Mr. Eagan's observations that it is not easy to produce castings with 100 per cent acicular structure appears to be sound and the further we depart from conventional test bars the more difficult the problem becomes. The formation of an acicular structure (bainite) is the result of the transformation of austenite at temperature ranges of 900° F. to 600° F., and in the areas adjacent to the gates or risers, the cooling rates are very slow and we have some high temperature transformation products of austenite (pearlite or possibly ferrite) taking place. The structure may well be acicular in sections removed from the gates or risers where the cooling rate is faster.

In regard to the apparent inability of Mr. Eagan to show a corresponding increase in the endurance limit by heat treatment which increased tensile strength, I am wondering if there might be some correlation between his findings on ratio of tensile to endurance and the fact that deflection and impact properties are not increased by a quench and temper treatment. This latter investigation was made by G. A. Timmons and reported in a paper entitled "Some Factors Involved in Hardening and Tempering Gray Cast Iron" by G. A. Timmons, V. A. Crosby and A. J. Herzig appearing in TRANSACTIONS, American Foundrymen's Association, vol. 47, p. 397 (1939).

Comparing the properties after the quench and draw treatment producing the maximum strength with the original as cast properties, the following are found: (a) higher strength, (b) higher hardness, (c) same transverse strength, (d) reduced deflection, and (e) reduced impact. Comparing the quenched and drawn irons of the same hardness as the "as cast" iron the heat treated irons will be found to possess: (a) higher strength, (b) lower transverse, (c) lower deflection, and (d) lower impact.

MR. EAGAN (author's closure): I definitely blame the pseudo-dendritic structures that were in heat numbers 3364, 3397, 3913 for the poor results we obtained as brought out by Mr. Schaum. Heat treatment apparently does not allow sufficient time for the combined carbon to migrate throughout the matrix. Thus, instead of having a uniform structure we have a great number of structures depending on the combined carbon at any one spot.

The tensile test on such structures indicates an increase of considerable proportions; however, the endurance limit is apparently that of the weakest part of the structure and no doubt is affected by the notch effect condition set up by these various structures. The heat treatments given heat numbers 3364 and 3397 are those usually recommended for commercial castings. The one given heat number 3913 was an attempt to allow a redistribution of the combined carbon so as to eliminate the pseudo-dendritic structure and then to attempt to produce an acicular structure. It of course was unsuccessful.

I hold no brief for the original as-cast structures of these three heats; however, they are what may be expected when we alloy gray iron in an attempt to produce high physical properties in heavy sections without taking into account the proper carbon and silicon balance. Such structures are not uncommon.

The only attempt at producing an acicular structure was with heat No. 4615. This heat was chosen because it had a matrix which was almost 100 per cent pearlitic, free of excess carbides, and had no combined carbon segregation.

In answer to Mr. Crosby's discussion, heat No. 4615 was in our opinion a pearlitic matrix iron produced in a commercial casting of relatively heavy section. What I wished to show with this heat was that with the normal oil quench and temper treatment you can obtain a good increase in tensile strength but a poor increase in endurance limit. The best increase in endurance limit is that obtained by producing an acicular structure which was almost obtained in heat treatment "C".

By shaking out the casting hot you can also produce a relatively high endurance limit (heat treatment D). However, designers usually use the tensile strength as the criterion for endurance limit and if they use the usual 0.5 ratio they would surely not be on the safe side for heavy sectioned castings.

In answer to Dr. Flinn's discussion, we have done considerable work on producing the cast acicular structure in the same sections as reported in this paper. The best we have been able to do is a 95 per cent acicularity. The ratio of endurance limit to tensile strength is approximately the same as that produced by heat treatment but the endurance limit has varied considerably perhaps due to other things than the structure such as micro-shrinkage. I hope to report these results later on.

We did not attempt to temper the acicular structures at 500° F. because I felt that the long hold at 650° F. was sufficient to allow the transformation to go to completion. The tempering temperature of 1050° F. is normal for commercial heat treatment to produce a machinable hardness so it was chosen. Obviously to do endurance limit work properly, a number of different tempering temperatures would be desirable but expensive. We were limited as to the amount of material available.

Both Mr. Crosby and Dr. Flinn asked about the reason for selecting the large fatigue specimen. We have for a long time believed that the use of the smaller diameter specimen for testing gray iron was undesirable. Our reasoning behind this was that with such a heterogeneous material as cast iron any small, normally insignificant flaw which could be in the iron would affect the results obtained far more than it should. Even large graphite flakes may do this.

Secondly, when testing heavy sections it is not possible to have the full section in the endurance test bar, hence in most cases the section actually tested is in the center of the cast section. Small diameter test bars therefore do not represent sufficient material, whereas the larger diameter bars take in more territory and should give a better representative test.

Furthermore, high tensile strength gray iron is difficult to machine, thus when machining the small diameter specimens there is some chance of flexing the specimen which of course may affect the results obtained. The chance of damaging the larger diameter specimen is less.

As to the correlation of the small diameter and the large diameter test bars, we have run one test and found that with the small material we obtained an endurance limit of 25,000 psi. with the small specimen and 21,500 psi. with the large specimen. One test is not a good criterion for judging; however, we felt that at least it was on the safe side.

We do not have the combined carbon analysis on any of the heats reported in this paper.

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[1]American Brake Shoe Co., Mahwah, N. J.
[2]Du Pont de Nemours & Co., Inc., Wilmington, Del.
[3]Naval Research Laboratory, Washington, D. C.

STEEL CASTINGS AND WELDMENTS RESIDUAL STRESS RELIEF

C. R. Jelm
and
S. A. Herres

Available literature and data on the relief of residual stresses by thermal treatment have been reviewed and analyzed. Three basic methods have been utilized in determining the effects of stress relief annealing: (1) determination of the residual stresses or movement upon machining an actual part or structure before and after annealing; (2) determination of the remnant elastic extension or bend after annealing suitable specimens prestressed in a jig, such as a bolt fixture, or a strip specimen bent to a predetermined radius; (3) determination of a stress-time curve by loading a tensile specimen at an elevated temperature and reducing the load to maintain the gage length constant at the initial extension. The relief of stresses during annealing is due to the combined action of the reduction in yield strength of the material on heating to elevated temperatures and creep under decreasing stress. This latter effect decreases at temperature with decreasing stress level.

Analysis of the available stress relief data indicates that at the present time (i.e., within accuracy of present test methods) the percentage relief of residual stresses, brought about by stress relieving in the temperature range of 750° to 1300° F., appears independent of steel type, composition, or yield strength. However, the higher the original residual stresses the higher will be the remnant stresses after a given thermal stress relief treatment.

RESIDUAL STRESSES are elastic stresses which exist within a body as a result of non-uniform plastic deformation which may have been caused by thermal gradients, phase transformation gradients, and hot or cold working during fabrication or in service. They are present in all metal structures regardless of how they are fabricated, but their intensity may vary and may at times be controlled from immeasurably small to extremely high values depending upon the material, the fabrication process, and the prior service.

Residual stresses are known to cause undesirable dimensional changes in fabrication and service and at times to affect adversely the service properties of structures. It is therefore important to study the methods of controlling and reducing residual stresses.

Residual stresses may be relieved by: (1) removal of stressed metal which usually results in deformations due to the release and rebalancing of the residual stresses (the cause of machining instability); (2) conversion of the elastic stress to plastic strain which can be accomplished by appropriate deformation under external loading and by thermal treatment.

The first method has no important commercial application as a means of relieving stresses. Shot blasting, tumbling, peening, hydrostatic and proof loading, or limited cold work are utilized in certain special applications. The most important and the most widely applied means of relieving residual stresses is by the thermal treatment.

The purpose of this report is to review and correlate test methods and available data on the thermal relief of internal stresses and to indicate the factors involved in selecting the most practical thermal stress relieving cycles.

It has been known many years that internal stresses did exist and did affect the fabrication and service characteristics of metals[1, 2, 3, 4] and that their elimination was in many cases essential to secure machining stability and to prevent warpage or distortion and premature failures in service, but no reliable quantitative data on their magnitude or relief were developed until comparatively recent times.

Barrett[12] in a review of the methods of internal stress measurement and their history reported that Kalakuzky[5] in 1889 made the first quantitative determinations of internal stresses in iron cylinders; somewhat later Heyn and associates[4, 6] and Howard[7] in the period 1910-1915 developed similar methods of calculating the stresses in simple objects by measuring the dimensional changes on machining.

However, work by Sachs[8] in 1927 showed that the calculations of these investigators and others who used their systems were considerably in error due to certain erroneous assumptions. It was, nevertheless, be-

NOTE: The statements or opinions expressed herein are those of the authors, C. R. Jelm, formerly associate metallurgist, Watertown Arsenal, Watertown, Mass., and S. A. Herres, Captain, Ordnance Department, Watertown Arsenal, Watertown, Mass., and do not necessarily express the views of the Ordnance Department. Presented at a Steel Session of the Fiftieth Annual Meeting, American Foundrymen's Association at Cleveland, May 8, 1946.

fore this time that the relation between internal stresses and "season cracking" of brass was determined and a method for its prevention by thermal treatment developed[9, 10, 11].

Several papers on the thermal relief of residual stresses were published prior to 1930[2, 13, 14, 15, 16, 17], but for the most part these qualitative discussions failed to amplify or clarify existing concepts on the relief of stresses in ferrous materials which were established prior to the turn of the century. Quantitative data when presented were likely to be misleading. The results of several quantitative investigations and reviews[18-26] have been presented in the past few years but the need for more and accurate data on the thermal relief of stresses is apparent from the paucity of the existing reliable data.

One of the chief difficulties in the study of the relief of stresses has been the development of suitable test methods which would permit simple accurate determination and control of the initial and final residual stresses and a comparison of the efficiencies of various stress relieving cycles which could be translated into the service or fabrication requirements of the individual application.

Structure Tests

There are three basic methods which have been used and are being used at the present time. For lack of better annotations these will be termed for the purpose of discussion "structure tests," "simulated structure tests," and "constant gage length tests"; they are discussed in detail in the subsequent paragraphs.

Structure Tests. Perhaps one of the oldest and most common methods of obtaining data on the relief of internal stresses and the one which is most readily applicable to specific individual problems is the "structure test." This test method consists of subjecting an "as stressed" structure, such as quenched or cold drawn or rounds, tubing, castings, simple welded structures, etc., to desired thermal treatments and subsequently determining the dimensional changes upon machining in an appropriate manner. By comparing these measurements, or the calculated stresses, to those of the "stressed" structure the efficiencies of the different cycles can be determined.

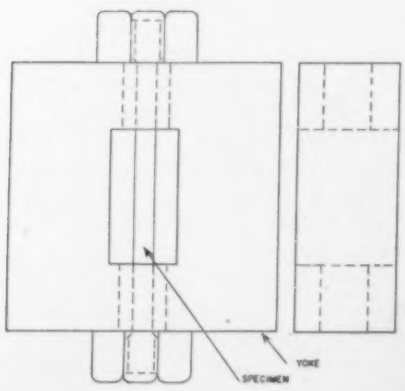


Fig. 1—Schematic diagram of stressing jig (Stewart Method).

Stress Calculations

Because of the relative simplicity of the stress calculations in rounds and cylinders, the use of this type of structure has been popular. Heyn[4] in 1912 obtained quantitative data on the thermal relief of stresses in cold drawn rounds by this method, and it has since been employed and refined by other investigators[27-33]. Cold drawn and heat-treated tubing[34-36] have been used, as have ring, hoop, wheel and hub and other specifically designed castings[37-42]. Simple welded structures have been utilized in some instances[20, 43]. X-ray determination of surface stresses in cold drawn and annealed wire was conducted by Faggiani[44]; this method of stress determination has been discussed by others[12, 45].

Except for parts identical to those under investigation, the relief data obtained by this method are not generally applicable unless special precautions are taken. In the first place, the determinations of the actual stress values are difficult and tedious to perform in other than the most simple shapes[22, 36] and are very likely to be only approximate.

In several of the investigations only single deflection measurements were made, such as in the slitting of rings[37-38], tubing[34], spokes of hub and wheel castings[41-42] or other similar structures[40, 43]. Stress calculations in such cases are very likely to be in error and may be entirely misleading.

When the percentage reduction in movement upon sectioning is determined as a function of annealing temperature and time, the results are probably an approximation of the relief of stresses in that structure, but they cannot be interpolated to other structures unless the actual stress magnitudes are known, both before and after annealing.

Not only is the determination of stresses in structures difficult, but control of the magnitude and distribution is virtually impossible. They must generally be accepted as they develop and exist. As the relief of residual stresses depends to a certain extent upon their severity, as shown later, the data obtained with one initial stress magnitude and distribution cannot be readily applied to another structure where the conditions are not the same. It is, therefore, evident that the "structure test" is difficult to perform and the data difficult to apply to other than specific applications.

Simulated Structure Tests. Owing to the difficulties and disadvantages of the structure tests, other test methods have been developed which afford greater control of the original stresses and an easy determination of the residual stress values. Benson and Allison[47], Moore and Beckinsale[9] and others[46-53] have developed such methods.

Naval Research Method

The method developed by Stewart[49] at the Naval Research Laboratory, and later applied by Beyma[51] and Cunnick and McDowell[19], is probably the most convenient and readily interpretable of the group. This consists of a simple yoke jig and an ordinary tensile specimen as shown in Fig. 1. By tightening the nuts at either end of the specimen any desired tensile strain can be applied (it may be desirable to key the specimen to prevent the establishment of torsional stresses).

The assembled jig is then placed in a furnace and subjected to a stress relieving cycle. After allowing the assembly to cool completely to room temperature, the nuts are loosened and the amount of residual elastic strain is determined. Since the initial and final elastic strains are known the percentage relief can be easily calculated and the actual stress values can be determined by multiplying by the elastic modulus.

Slightly modified forms of this test have been used by other investigators[47, 52]. Benson and Allison[47] applied the initial stress by shrinking the specimen from an elevated temperature over a closely machined two-section sleeve. Mochel[52] used a

sleeve instead of the large yoke to support the specimen.

In using this test method certain simple precautions are necessary to obtain reliable results. First, the yoke and the specimen should be constructed from the same materials in order to avoid differences in the thermal coefficients of expansion; a difference as small as 0.5×10^{-6} in. per in. per °F., such as can easily exist between two types of steel, may result in an error of the true stress value of 15,000 psi. after cooling from 1000° F.

Second, the heating and cooling rates of the assembly must be slow enough that no significant temperature differences develop between the yoke and the specimen.

Third, the ratios of the cross-sectional areas of the specimen and yoke should be properly adjusted—two arrangements are possible which offer the least chance of introducing an additional variable due to the different rates of relief of the different stresses in the two sections: (a) the areas of the specimen and yoke approximately equal, and (b) the area of the yoke many times larger than that of the specimen. The latter condition approaches the stress distribution in welds where high tensile stresses exist in and along weld, balanced by comparatively low compression stresses outside weld zone.

Other simulated structure tests consisted chiefly of strip specimens elastically bent to a desired radius, secured, and subjected to a thermal cycle after which the remaining elastic bend, or "spring back," was determined on unclamping. The maximum initial and final stresses were calculated from the following beam formulae:

$$S = \frac{Et}{2r} \qquad S^1 = \frac{Et}{2\,(r^1 - r)/r \times r^1}$$

where S and S^1 are the initial and final stresses, t the thickness of the strip, and r and r^1 the radii of the strip as bent and upon unclamping. This type of test with modifying features has been used by many investigators[9, 46, 50, 53].

This latter simulated structure test does not appear as simple as the Stewart method, but may be more convenient in special cases. Similar precautions have to be taken in the choice of clamping materials to avoid differences in the thermal coefficient of expansion, and the thickness and radii of the strip have to be closely determined. The formulae for the calculation of the initial and final stresses assume that the stress distribution within the strip varies linearly from maximum tension at the outer fibres to maximum compression at the inner fibres.

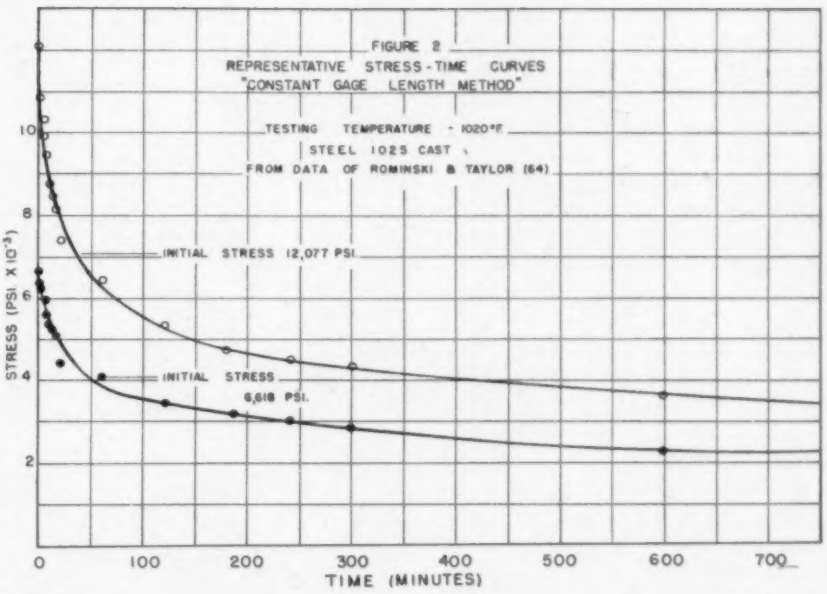


Fig. 2—Representative stress-time curves (Constant Gage Length Method).

Stress Relieving

If the initial applied stress exceeds the elastic limit, the calculations are no longer valid; and, in any event, after stress relieving, the assumed condition is no longer present as the higher surface stresses are relieved more rapidly than the lower stresses in the center of the strip destroying the linear distribution as pointed out by Scott[50].

Constant Gage Length Tests. As the structure and simulated structure tests require an extensive number of tests to investigate the effect of holding time on the relief of residual stresses and involves a variable factor in the heating and cooling rates, a third test method has been developed. Briefly, this method of testing consists of loading a suitable tensile specimen to a desired stress (or extension) at an elevated temperature and subsequently reducing the applied load to maintain the gage length constant by reducing the elastic strain to compensate for the plastic extension due to creep. Two stress-time curves are shown in Fig. 2.

This type of test was conducted by Mailander[54-55] in 1931 in studies on two Cr-Ni steels and has since been used extensively in the study of the relaxation of stress in bolts[56-63, 65-67] and occasionally in the study of stress relief[61, 64]. Detailed descriptions of the apparatus used in this test are given by Boyd[58] and Nadai and Boyd[63]. A modified form of the test was used by Betty, MacQueen and Rolle[68] in the study of nickel alloy springs.

The advantages of this method are evident; it offers close control over the applied stress and the residual stresses are simply and continuously determined with appropriate apparatus. Only a single test need be run at one temperature and stress level to determine the effect of time, whereas several tests of either the structure or simulated structure types would be required. Finally, the data are more readily interpreted quantitatively, as will be discussed later.

The disadvantages are that in order to translate the applied and residual stresses at the elevated temperature to similar room temperature conditions, the modulus of elasticity of the material at the testing temperature should be known; this property is difficult to determine at high temperatures. Secondly, changes in the density or specific volume of the material may introduce a considerable error in the results as related to the stress relief of a structure; a change of as little as 0.0002 cm.^{3} per gm. in the specific volume during the test will cause a change of about 6,000 psi. in the remnant

stress. Similarly, small changes in the modulus of elasticity will alter the results.

Changes in the properties of the steel during the test may have an effect on the relief of the stresses, and their determination may be necessary for complete interpretation of the data. Finally, in the constant gage length tests all relief occurs at the testing temperature while in actual stress relieving cycles a great portion of the relief occurs over a range of temperatures during heating; to apply the constant gage length test data to the specification of thermal cycles it must be assumed that the temperature of relief on heating does not effect the remnant stress values as determined by this method.

Data Compilation and Correlation

Data on the effect of annealing on the residual stresses in steel are surprisingly meager; Figs. 3, 4 and 5 represent a compilation of most of the existing data on thermal stress relief. The soaking times of 1, 4 and 8 hr. were chosen because more data were available for these holding times than for others. Data existing for other miscellaneous soaking periods would have added little to this presentation; reference has been made to these investigations previously and they are noted in the bibliography.

There is considerable scatter in the plotted data. Most of this can be accounted for by variations in the heating and cooling rates among different investigators, discrepancies between actual and recorded temperatures and times at temperature, errors in strain measurements and stress calculations, etc.

As the plotted data include the results of many different types of steel varying in alloy composition and strength, part of the variation may be due to differences in the steels, as well as to the different types of stress measurements and methods of investigation. However, analysis has indicated that both the mild steel and the alloy steel data are distributed approximately equally between the extreme values of the data. All the values fall within a range of ± 15 per cent maximum to ± 5 per cent at the higher stress relieving temperatures and longer holding times.

Both the meagerness and the scatter of the available data prevent the establishment of any reliable conclusions on the effect of alloy content on the thermal relief of residual stresses in steel at the present time. The need for additional accurate data obtained under controlled and reproducible conditions is apparent. For these reasons it is believed that at the present time, the most useful summation of the data is a plot of the mean paths of the data of Figs. 3, 4, and 5 as shown in Fig. 6. On the basis of the present knowledge, this figure represents an approximate working basis for the specification of thermal stress relieving cycles for most structural ferrous materials.

Different materials may behave somewhat differently, but lacking specific information on the steel and structure under consideration, Fig. 6 can be used as a guide with assurance of its accuracy within the limits cited previously. Most of the

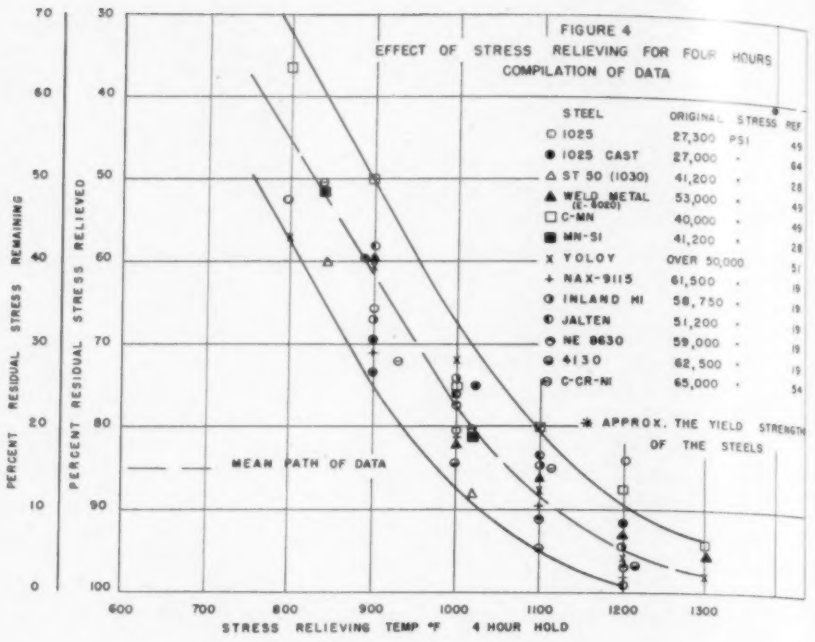


Fig. 4—Effect of stress relieving for 4 hr.

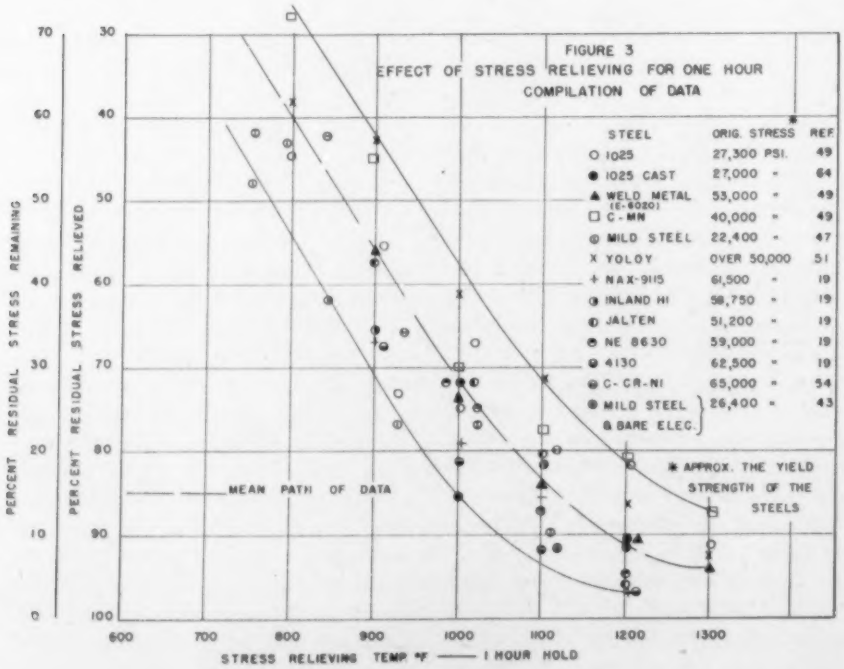


Fig. 3—Effect of stress relieving for 1 hr.

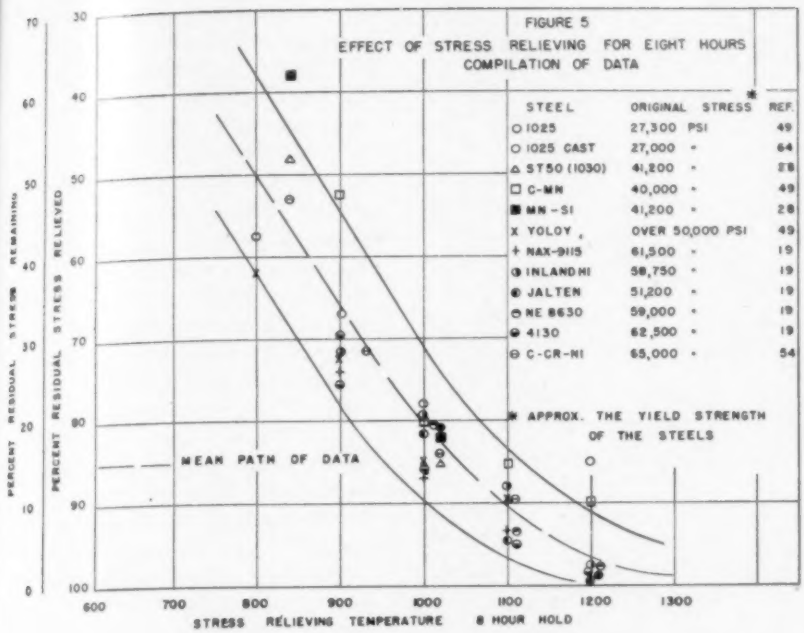


Fig. 5—Effect of stress relieving for 8 hr.

data presented were secured on uniaxially stressed parts or specimens; in service structures the internal stresses are generally multiaxial. The effects of this condition on the relief of stresses have not been determined.

Certain amplification is, however, desirable. In Fig. 6 the effect of time of stress relieving is evident; this varies somewhat with temperature. At 900° F. increasing the soaking time from one to four hours is equivalent to increasing the temperature of annealing to 935° F.; at 1150° F. this same increase in time is equivalent to raising the temperature to 1200° F. Further increasing the time of hold from 4 to 8 hr. is roughly equivalent to raising maximum temperature of cycle 20° F.

The total effect of increasing the time of hold from one to eight hr. or the temperature 55° to 70° F. is to reduce the stresses an additional 9 per cent at 900° F. and 5.5 per cent at 1150° F., a relatively small factor when it is considered that the total reduction in the residual stresses at these temperatures is 65 per cent and 94 per cent, respectively. The relative effect of increasing the time of annealing decreases with increasing temperatures, at least above 800° F.

Discussion heretofore has been confined to the percentage relief of stresses and not to actual stress values; it is apparent that the two are not the same. If two structures built of materials of 30,000 to 60,000 psi. yield strengths with internal stresses approximating these values, are stress relieved to eliminate 50 per cent of the stress (825° F. for 4 hr.), it is obvious that the first will have a maximum remnant residual stress of 15,000 psi. and the latter 30,000 psi., equal to the maximum initial stress of the former.

To illustrate this condition more effectively, Fig. 8 has been prepared from the data of Fig. 4, plotting the maximum residual stress values as a function of stress relieving temperature (4-hr. hold) for three different classes of steel having yield strengths of 30,000, 50,000 and 70,000 psi. and initial residual stresses approximately equal to these values. If, for example, it is desired to reduce the maximum residual stress to 10,000 psi., it would be necessary to treat the 70,000 psi. yield strength steel at 1070° F., the 50,000 psi. yield strength steel at 1020° F., and the 30,000 psi. yield strength steel at only 920° F.

Reduce Residual Stress

The difference of 150° F. in the maximum stress relieving temperature is worth considering. This effect has also been considered as a function of soaking time in Fig. 7—to reduce the residual stress in a 50,000 psi. yield strength steel to 10,000 psi. maximum, eight hours are required at 990° F., four hr. at 1020° F., and only one hr. at 1055° F. Similar calculations can be made for other desired stress conditions and steels from Fig. 6 data.

It must be recalled that the foregoing data and discussion have referred only to residual stresses which were initially approximately equal to the yield strengths of the material. In welded structures, quenched

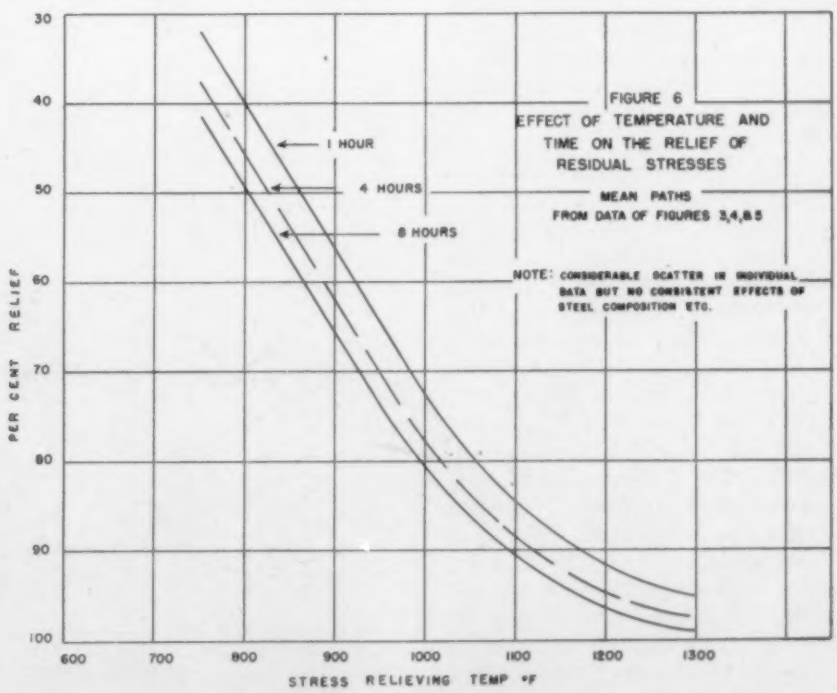


Fig. 6—Effect of temperature and time on the relief of residual stresses.

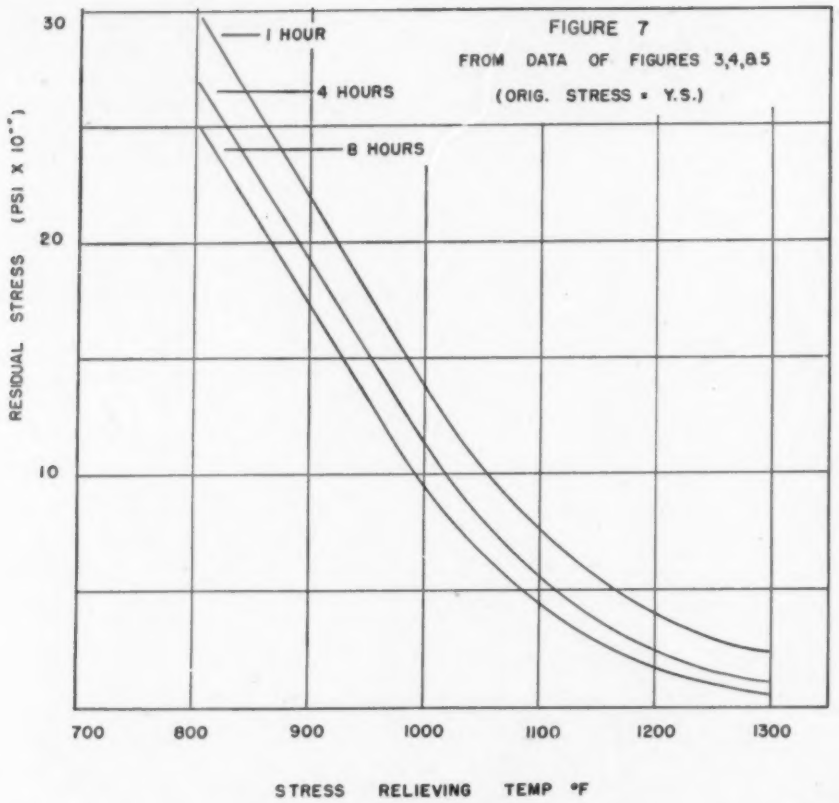


Fig. 7—Effect of stress relieving temperature and time on the residual stress of a 50,000 psi. yield strength steel.

parts, structures quenched or rapidly cooled from the tempering temperature, and some "as cast" structures generally approach this condition. This is readily apparent when it is realized that a temperature differential within a structure of only 200° F., established at a sufficiently high temperature so as to be initially stress-free, will cause a final residual stress at room temperature of about 45,000 psi.

There are, however, many instances when the residual stresses are less than the yield strengths of the steels. Unfortunately, insufficient data are available to present a quantitative analysis of the effect of initial stress level, but its effects are illustrated in Figs. 9 and 10, derived from the data of Mailander[54]. Two stress levels, 30,000 psi. and 50,000 psi., for a Ni-Cr steel are represented. Fig. 9 shows that the percentage relief of stresses is greater in the specimens with the higher initial stress, and Fig. 10 that the actual remnant stress values are also higher. It can be qualitatively expressed that the higher the initial stress the greater the percentage relief, and the higher the residual stress values after any given thermal cycle, limited, of course, by the values which can be calculated from Fig. 6.

In the usual stress relieving range of 1000° F.—1200° F. in the illustrated example (Fig. 10), though the initial stress difference was 20,000 psi. the final stress difference was only 1,500 psi. In general, the higher the stress relieving temperature and the longer the soaking the less will be the effect of initial stress level on the final stresses.

The relief of internal stresses by thermal treatment can be considered due to the transformation of elastic strain (stress) to plastic strain. This is accomplished by the combined action of reducing the yield (flow) strength of the material on heating to an elevated temperature and creep. These two actions, though related, can be most conveniently considered as two distinct factors. Except under conditions of constraint or suitably combined stresses a material will not support stresses, either external or internal, in excess of its yield strength (as measured in the tensile test) without deforming plastically.

On heating, the yield strengths of most materials are lowered appreciably as shown in Fig. 11, and as a result the internal stresses are lowered due to the decrease in the elastic strain. Also, as is generally known, metals at elevated temperatures (some even at room temperature) flow plastically under stresses which are considerably less than their yield strengths—this behavior is known as creep. In stress relieving this occurs in a somewhat different

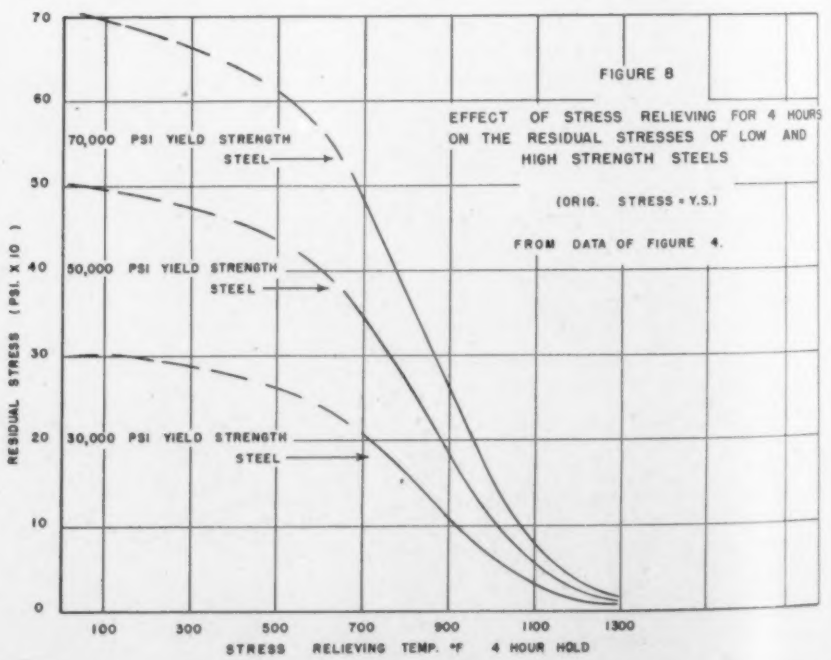


Fig. 8—Effect of stress relieving for 4 hr. on the residual stresses of low and high strength steels.

form than determined in the creep test as the total strain, not the load, is maintained essentially constant which results in a continual reduction in stress due to the plastic flow which occurs. The phenomena of creep under decreasing stress has been termed "relaxation." It occurs on heating up into the creep range (above 400°-500° F. in most steels) while holding at temperature and during cooling from the annealing temperature.

The amount of stress reduction which occurs due to relaxation in a given steel depends upon their rate of heating, maximum temperature, the holding time, and the rate of cooling. The higher the temperature and the longer the time spent in heating and cooling and at temperature the greater is the relative effect of the relaxation. As temperature is one of the chief determining factors of the rate of relaxation, the rate of cooling from the annealing temperature will have only a small effect on the relief of the stresses as the stresses have been reduced to such a point by heating to and soaking at temperature that further relaxation can only take place to a very limited extent.

If the structure is held at the annealing temperature for any reasonable length of time, the rate of heating becomes relatively unimportant as the stress reduction due to relaxation which occurs in any normal heating-up period would also occur in a few minutes while holding at the maximum temperature. The effect of holding at temperature can be most readily discussed with reference to Fig. 2. The relaxation of the stresses is initially quite rapid but the rate decreases rapidly as the stress is reduced, and after a relatively short period of time further reduction is limited. In Fig. 2, the 12,000 psi. initial stress has been reduced to about 5,000 psi. in 2 hours; holding an additional 10 hours at temperature will reduce the stress only another 1,500 psi., a negligible amount considering the time factor involved.

At higher temperatures the relaxation curves are steeper and the leveling off occurs in a shorter time. For example, at 1200° F. one hour accomplishes most of the relief, and at 1380° F. 15 to 20 minutes is ample[64].

The effect of holding time on the relief of stresses can be quantitatively analyzed from relaxation data. Rominski and Taylor[64] suggested that the rate of relief could be expressed as a function time such that

$$-\frac{dS}{dt} = mt^{-1}$$

which when integrated gave

$$S = -S_o \log \frac{t}{t_o}$$

where $\frac{dS}{dt}$ is the rate of relief, S the stress, t the time and m, S_o, t_o appropriate constants. This analysis appears approximately to represent the data but extrapolation of the stress-time formula to zero time (the beginning of the test) shows that the initial stress would be infinitely large and in the other direction that the internal stress is eventually completely effaced; an impossibility as indicated by other investigators[52, 56, 58, 61].

There appears to be no physical basis for considering the rate of relaxation as a function of time as the properties of the material remain substantially the same throughout the test. The fact that the relaxation tests of Rominski and Taylor[64] at the higher temperatures completely relieved the applied stress can be accounted for by the insensitivity of the apparatus.

It appears more feasible that the rate of relief would be a function of

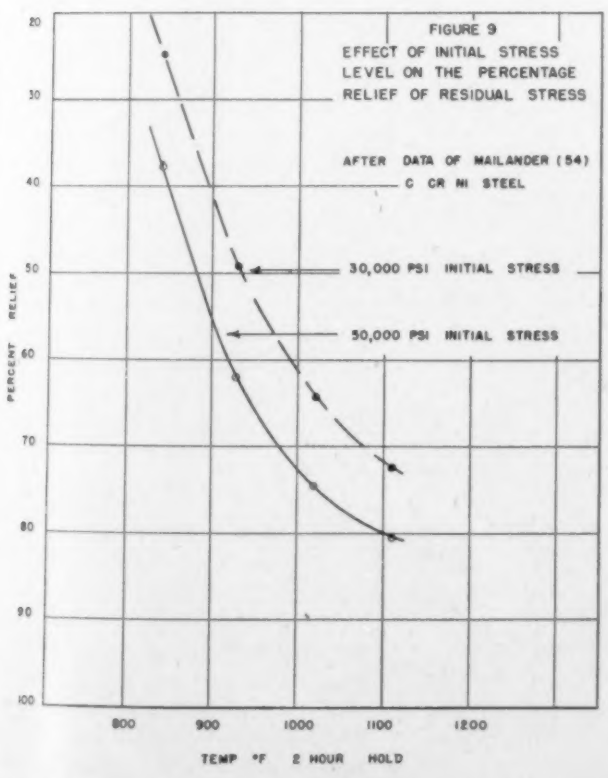


Fig. 9—Effect of initial stress level on the percentage relief of residual stress.

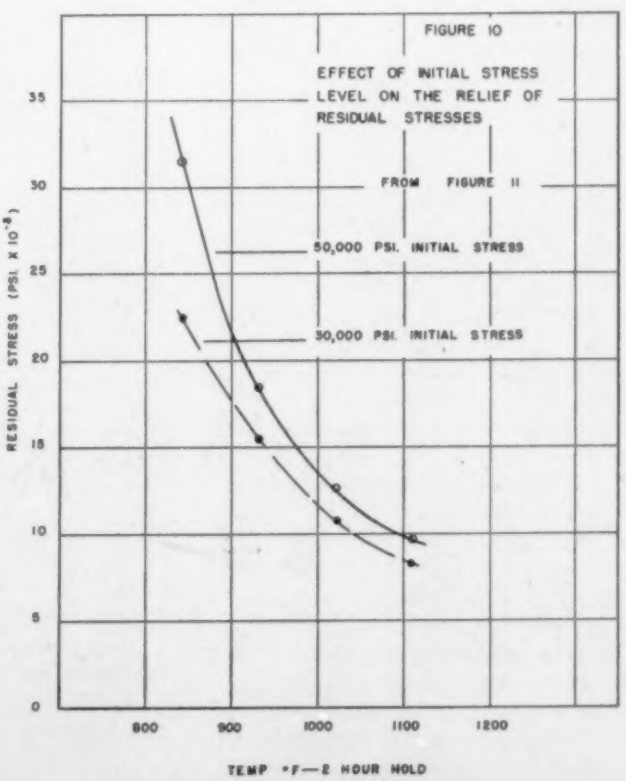


Fig. 10—Effect of initial stress level on the relief of residual stresses.

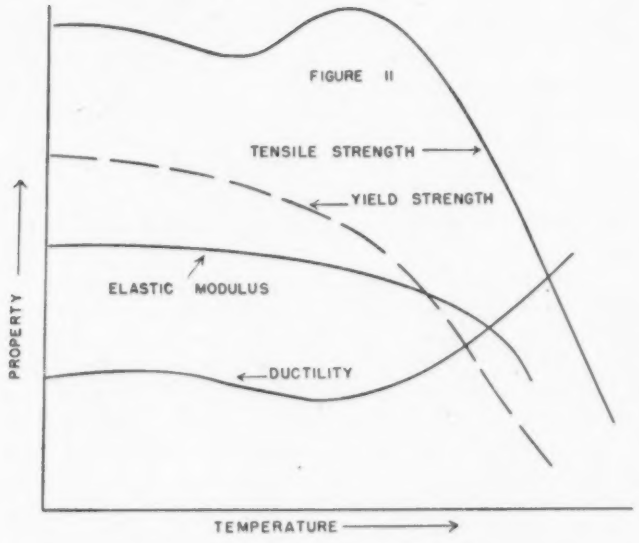


Fig. 11 (above)—Schematic representation of the effect of temperature on the tensile properties of steel.

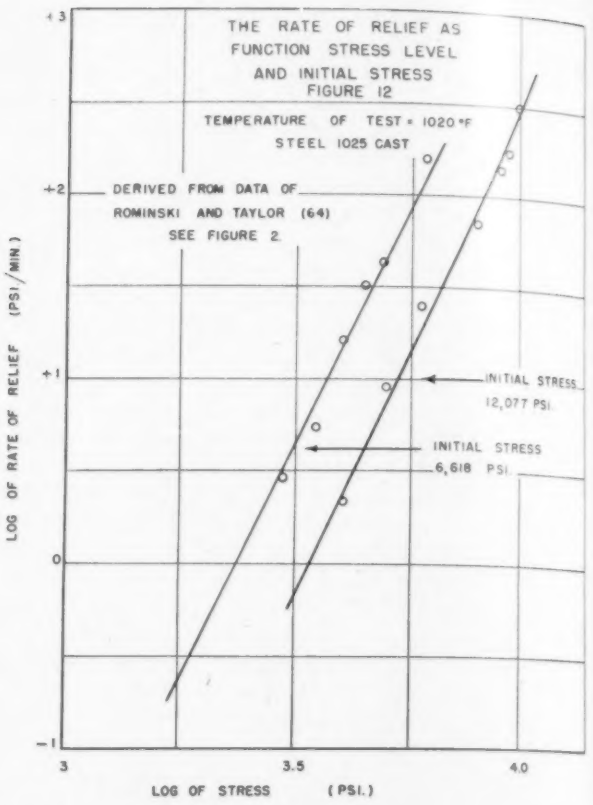


Fig. 12 (right)—Rate of relief as function stress level and initial stress.

the applied stress. The stress-time curves of Fig. 2 are replotted in Fig. 12, where the log of the rate of relief is plotted vs. the log of the stress. It will be noted that the best representation of the data is a straight line. This line can be represented by the equation

$$\frac{dS}{dt} = KS^n \quad (1)$$

where $\frac{dS}{dt}$ is the rate of relief at stress, S the stress, and K and n are constants, n being the slope of the line. The remainder of the data of Rominski and Taylor[64] can be interpreted by this formula as can the data of other investigators[54, 58, 62]. Bailey[56], Robinson[57] and others in the study of the behavior of bolts similarly developed this rate-stress relationship.

Integration of this relationship gives the stress as a function of time

$$S = [(n-1)\,Kt + S_i^{1-n}]^{\frac{1}{1-n}} \quad (2)$$

where S is the stress at time t and S_i is the initial stress, K and n being the same constants as before. Insufficient data of the proper nature are available at the present time to evaluate completely the effects of time and steel composition and treatment on the constants of the equation, but if and when such data does become available the stress relief behavior of various steels can be readily compared by a consideration of these constants. Analyses of the data of Mailander[54] and Rominski and Taylor[64], however, indicate that n, the principal shape factor of the stress-time curves, decreases rapidly with increasing temperature.

Another effect of heating on the elastic properties of metals which effects the relief of stresses is also illustrated in Fig. 11; i.e., the modulus of elasticity decreases with temperature. The effect of this is to decrease the stresses existing at the elevated temperature even if no plastic deformation has occurred. For example, if the elastic modulus is decreased 50 per cent, the internal stresses also would be reduced 50 per cent as the strain remains constant. Unfortunately, this effect is reversible, and if no plastic flow occurs due to other considerations the stresses will return to their original values on cooling the structure or specimen to room temperature.

In interpreting stresses applied or existing at an elevated temperature, as in the constant gage length tests, it is necessary to multiply them by a factor $\frac{E_0}{E_t}$, where E_0 and E_t are the elastic moduli at room and elevated temperatures respectively[54] in order to obtain the corresponding stresses which would exist at room temperature under similar strain conditions. Determinations of elastic moduli at elevated temperatures, above 900° F., are difficult and are not reliable unless obtained dynamically.

The size or thickness of the structure does not affect, as is occasionally inferred, the relief of residual stress. The only consideration as to soaking time that need be given to size is to insure that all portions are heated to the desired elevated temperature. With large sections the soaking time to insure temperature uniformity is naturally greater, but once the structure is *at temperature* its size need not be considered as affecting the relief of stresses.

Application of Data—Principles

The most common thermal treatment to accomplish the relief of internal stresses[69] is to heat the structure uniformly to 1100°-1200° F., to hold at temperature one hr. per in. of thickness of the heaviest section, and subsequently to cool slowly in the furnace. Reference to Fig. 6 indicates that anywhere from 85 to 95 per cent of the internal stresses should be removed by this treatment.

This cycle has been found to be adequate for most applications, and in some cases may have been more than sufficient.

As discussed in the previous section there are four variables in the stress relieving cycle: Heating rate, maximum temperature, time of hold, and cooling rate; and three variables in the structure—material and prior treatment, section size, and residual stress pattern. The heating and cooling rates do not significantly affect the relief of residual stresses but must be controlled because of other considerations.

Heating Rate

With complicated and heavy structures the rate of heating must be slow enough to prevent cracking and undesirable distortion due to nonuniform expansion of the structure. Likewise, the rate of cooling from the annealing temperature should be slow enough to prevent the establishment of new *harmful* residual stresses. The size of the structure does not affect the relief of stresses but must be considered in establishing heating and cooling rates and holding times to insure temperature uniformity throughout the structure at the annealing temperature and to prevent any harmful thermal gradients.

Therefore, for a given structure (material and stress pattern) the relief of the internal stresses depends principally upon the maximum temperature and the time at temperature. The effect of time and temperature on a highly stressed structure may be evaluated approximately from the data summary of Fig. 6. The present available data do not indicate any significant or consistent differences in the percentage relief of stresses with different steel compositions and types (additional accurate data are necessary). However, as discussed previously, the maximum remnant stresses increase with the strength of the steel.

The relief of internal stresses is in many instances not the only effect of annealing at elevated temperatures. Softening[70] and tempering or precipitation hardening may occur depending upon the material, annealing temperature, and time. In weldments annealing will also soften any hardened heat-affected zone which may have developed in welding and may partially release any absorbed gases. These and other effects, though they may slightly alter the stress relieving characteristics of the material, should not be considered as the predominant factor in the determination of the stress relief cycle. Nevertheless, a compromise may often be necessary to secure the optimum physical properties and maximum stress relief.

A commercial copper bearing precipitation hardening steel offers an excellent illustrative example[71]; to obtain the maximum strength a 3-hr. treatment at 900° F. is required and only one hour at 1050° F. Figure 6 indicates that 19 per cent greater stress relief could be achieved by the latter treatment with only slight reduction in the maximum strength. Similarly, the effect of temperature and time on the properties of other materials must be considered in specifying stress relieving treatments.

The elimination of residual stresses is desired for three principal reasons: (1) Machining or service stability; (2) improvement in static strength; and (3) improvement in resistance to shock restraint, progressive, and other "non-static" failures. When stress relief annealing is conducted to secure machining stability, the amount of relief necessary obviously depends upon the degree of stability desired.

Machining Stability

The machining tolerances and the clearance tolerances for operation in service, the magnitude and distribution of the residual stresses in relation to the sequence and amount of machining to be conducted, the geometry of the structure and other miscellaneous factors affect the degree of stability required, and the proper minimum thermal stress relief treatment will therefore depend upon the individual application.

The minimum thermal treatment necessary to secure adequate machining stability for the particular application can be approximated in several ways. One of the more direct ways would be to machine the "as stressed" structure to determine the amount of undesirable "walkout," estimate the required reduction in "walkout" to meet the machining tolerances or other requirements, and finally, select from Fig. 6 an approximate thermal cycle to give the desired relief. If thermal stress relieving is already being performed, the severity of the thermal cycle can in some cases be reduced; the amount of existing "walkout," the required stability, and the relief data of Fig. 6 permit a determination of a more favorable reduced cycle.

Even if machining stability is obtained the influence of residual stresses on service instability must still be considered for certain applications where the service is such that a combination of service loading and residual stresses may produce dimensional changes which interfere with proper functioning of the structure. For such applications a degree of thermal stress relief greater than that needed for machining stability alone is required.

Static Strength

The effect of residual stresses on the static strength of the structure are somewhat conjectural. It is true that the proportional limit of the structure is lowered considerably by their presence, but this has little relation to the usual design figure of yield strength at 0.1 per cent offset which is generally not appreciably affected[18]. Further, the region of high internal stress in welded and other structures in most cases represents only a small portion of the supporting dimension; therefore the improvement to result from stress relief annealing in this respect is at best small.

If improvement is sought by stress relief annealing in the resistance to shock, restraint, progressive, or other types of failure accompanied by a brittle fracture, consideration must be given to the direction, distribution, and magnitude of the residual stress existing in relation to the stresses that are to be imposed during service and the ability of the material to withstand these conditions. Residual stresses can be both favorable (as evidenced by peening of parts subject to fatigue stresses and the cold working of gun tubes) and unfavorable to failures of this type depending upon these conditions.

Without proper consideration stress relief annealing may actually make the structure more prone to failure. Discussion of stress relieving for improvement in these cases in the light of present knowledge of their mechanism would be hypothetical. However, again it should be emphasized that each application

represents a special case in itself and that no rules are generally applicable.

The reason for the desire to minimize stress relief annealing treatments is entirely economical. Aside from the obvious savings in time and fuel, a more important consideration is arising in the development and use of high strength structural steels with yield strengths of 70,000 psi. and above.

With the general practice of stress relieving in the range of 1100° to 1200° F. these high strengths can be secured only by the use of higher carbon and/or alloy additions and consequent greater fabrication difficulties, regardless of whether "hot rolled" or heat treated steels are used. Reconsideration of stress relief annealing practice may indicate that in certain applications temperatures and times of annealing cycles can be substantially reduced resulting in higher strengths with smaller alloy additions provided that the annealing temperatures are not excessive for the particular application.

Summary

Three basic methods have been utilized for various investigators for determination of the effects of stress relief annealing: (1) Determination of the residual stresses or movement upon machining an actual part or structure before and after annealing; (2) determination of the remnant elastic extension or bend after annealing suitable specimens prestressed in a jig, such a bolt fixture, or a strip specimen bent to a predetermined radius; (3) determination of a stress-time curve by loading a tensile specimen at an elevated temperature and reducing the load to maintain the gage length constant at the initial extension.

Certain differences exist among the data obtained by the three methods, but properly interpreted the results are comparable. The choice of the method to use will depend principally on the exact information desired, but the second method appears to be the simplest to control and operate and yields the most readily interpretable data.

The relief of stresses during annealing is due to the combined action of the reduction in yield strength of the material on heating to elevated temperatures and creep under decreasing stress. This latter effect decreases at temperature with decreasing stress level (Fig. 12) as expressed approximately by the relation:

$$\frac{dS}{dt} = KS^n,$$

where $\frac{dS}{dt}$ is the rate of relief, S is the stress level, and K and n are constants depending upon the material, temperature, and other defining conditions.

Analysis of the available stress relief data (Figs. 3-5) indicates that at the present time (i.e., within accuracy of present test methods) the percentage relief of residual stresses, brought about by stress relieving in the temperature range of 750° to 1300° F., appears independent of steel type, composition, or yield strength. However, the higher the original residual stresses the higher will be the remnant stresses after a given thermal stress relief treatment.

In specifying thermal stress relief treatments, it should be recognized that the requirements of various individual applications will differ widely, as to required degree of machining stability and allowable residual stress.

Other effects of thermal stress relieving treatments, such as tempering, precipitation hardening, softening of a weld heat-affected zone, release of absorbed gases, and others, should be considered independent of the relief of residual stresses, though a compromise may sometimes be necessary between optimum physical properties and maximum relief of stresses.

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CONVENTIONAL APPRENTICESHIP

The author describes the several indentured apprenticeships offered at his plant. The paper covers apprentice standards, qualifications, selection and classification of apprentices.

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Asst. Foundry Supt.
Falk Corp.
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CONVENTIONAL APPRENTICESHIP is a long range training program designed to groom the applicant thoroughly in the fundamentals of the trade or craft chosen as his life's work. Apprenticeship as old timers remember it differed from the present training program.

In olden times, so we are told, the head of the family determined the line of work his off-spring was to follow. Elementary education was all the youngster could expect, and that usually terminated at the age of twelve, and would generally carry the young man through the fifth or sixth grade.

If he was an exceptional student he might complete the seventh grade. Anything beyond that was considered higher education and not essential in a learner. So, when the young man reached the age of twelve he began to follow in his father's footsteps in mastering a trade.

Fundamentals Mastered

Fundamentals of the trade were mastered though it was a long and painful process with minimum of pay. We wonder sometimes how anyone was persuaded in the olden days to learn a trade. Nevertheless, a number of competent mechanics were developed and they have probably formed the backbone of the various industries in the shop. Many top-notch supervisors, inspectors, etc., have come up through the ranks the hard way, nevertheless, they have the advantage of knowing what they are talking about. We have all had contact with these men and have learned to rely on their judgment.

A molder, for instance, is a better all-around mechanic if he has had actual experience in core making, pattern making, melting and other necessary operations in the foundry, regulating of course, the number of hours allotted to each according to the type of training taken. In addition to this, a number of communities have elaborate educational facilities to supplement the shop work. These are Vocational Schools, Trade Schools and the like, which are an asset in training of young people.

Training Programs

The training programs at the author's plant are separated into the following divisions: Apprenticeship, Graduate Engineers, Shop Training, Co-operative Training, and Education.

Apprenticeship. The author's plant offers indentures in the following apprenticeships: Welding, Core Making, Steel Molding, Electrical Maintenance, Mechanical Drafting, Pattern Making, and Machinist. Each of these apprenticeships have two classifications depending upon the previous schooling and experience of the apprentice. Also each classification has its individual wage rate. We realize, however, that these wage rates will have to be determined by the prevailing rate in the industry of the various communities.

Graduate Engineers. Two year programs are offered university graduates in Machine Shop, Foundry, and Engineering Departments. Duration of this program may vary with the conditions and background of the trainee.

Shop Training. Numerous types of programs are offered in Shop Training, including Machine Work, Drafting, Inspection, Foundry Work, Welding, Tool Room, Blue Print Reading, and Precision Instrument Reading. These programs are set up to meet a specific need and to train men that show ability, but for definite reasons cannot be apprenticed.

Co-operative Training

Co-operative Training. The author's plant maintains a Co-operative Training Program with the College of Engineering at Marquette University. Training for these men includes time in all the major departments of the Plant, with a view to giving a broad background in methods and policies, rather than any great degree of manual skill. Most of these trainees are absorbed into the organization upon graduation and are given an informal additional training in whatever specialty they may be placed.

Education. Educational facilities in the author's plant in Milwaukee are excellent both in day and evening courses and this additional supplementary training is encouraged by the company.

Apprentice standards. Qualification. Age. The applicant shall have

Presented at an Apprentice Training Session during the Fiftieth Anniversary Convention of the American Foundrymen's Association at Cleveland, May 9, 1946.

passed his 16th birthday and a birth certificate must accompany his application.

Education. A high school graduate is preferred and required on most of the company's apprenticeships. An official transcript of school records showing the applicant's grades and courses satisfactorily completed by him must be secured by the company.

Physical Requirements. The applicant will be required to meet the physical standards of the company, which will be determined through physical examination by a company physician.

Tests. The applicant may be required to pass mechanical aptitude tests administered by the Training Division.

Selection and Classification. Selection. Final selection of the applicant shall be the responsibility of the Training Division.

Classification. The apprentice shall be classified according to trade and given an intermediate or regular classification according to his school qualifications.

Ratio. It is not a company policy to load the number of apprentices in any trade classification. The author will outline the Regular Steel Molding Apprenticeship in his Foundry. Inasmuch as we are primarily a jobbing foundry we have drafted the following program.

Extent of Apprenticeship. The Apprenticeship shall consist of eight periods of 1040 hours each or a total of 8320 hours or approximately four years. The first 520 hours shall constitute the probationary period, but in no case shall this period extend beyond four months.

School Attendance. School attendance shall be eight hours per week until 576 hours of school attendance shall have been completed. School attendance shall be counted as part of the hours of labor.

Schedule of Processes to be Worked:

Assignment	*Hours*
Helping Molder	350
Machine Molding	350
Bench Molding	550
Large Cores	600
Small Cores	400
Green Sand Molding	2000
Sweep Molding	850
Dry Sand Molding	2300
Chipping Room	350
Heat Treating, Furnace, Laboratory, Pattern Work, Special Work	570
Total	8320

The wage scale is graduated for an increase after every period of the program and upon satisfactory completion of the training course a bonus of $100.00 is paid by the company. The graduate then after a consultation selects some division of the foundry he prefers and if there are no hinderances or objections he is ready to start as a journeyman molder.

This is the training program we consider ideal for conditions at our plant. Variations of the above are necessary to meet changing situations as they may arise due to either the pressure of work or the man power problem.

DISCUSSION

Chairman: F. C. CECH, Cleveland Trade School, Cleveland.

Co-Chairman: W. J. HEBARD, Continental Foundry and Machine Co., East Chicago, Ind.

C. J. FREUND[1]: *(Summary).* Apprentice training such as the old-fashioned 3 or 4-year foundry apprenticeship which has survived through the years and continues to flourish in spite of development and inauguration of other special training programs is bound to be good.

It is desirable for all foundries to introduce and operate at least some limited development of the old-fashioned foundry apprenticeship. The conventional apprenticeship is probably the only type of plant training which will produce the type of workman, mechanic, or craftsman who has ability with all-around general broad competence.

The automobile industry is not a jobbing type industry, but a production industry. It is one of a number of important American production industries. Some ten years ago a top automobile industry official remarked that his particular corporation depended almost exclusively on two types of men for top manufacturing or key jobs.

The first type is the old-fashioned trained machinist, or apprenticeship graduate, the kind of man who migrates from New England, Cincinnati, Philadelphia, or some old established machine-building community, where they have long had a regular machinist apprenticeship. The other type is almost invariably the railroad man. They, too, in many cases had formal apprentice training.

So there you have a big production industry with thousands of men doing repetitive work, but nevertheless, the men in charge are men who have come through the old-fashioned procedure.

[1]University of Detroit, Detroit.

MICROPOROSITY IN MAGNESIUM ALLOY CASTINGS

Interrelation of various major factors determining amount of microporosity in magnesium alloy castings . . . part of general program sponsored by Office of Production Research and Development of WPB . . . published with permission of Office of Production Research and Development of Civilian Production Administration.

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MICROPOROSITY IS A DEFECT which is by no means confined only to magnesium-base casting alloys. Because the principal uses of magnesium castings have been for aircraft, the quality standard has been high and the defect known as microporosity has received much public attention. As a result of this publicity, many users of castings have been led to believe that microporosity is a defect peculiar to magnesium alloys.

As a matter of fact, some aluminum-base alloys and many copper-base alloys are subject to this defect. All of these alloys, regardless of the base metal, have certain characteristics in common which make them subject to microporosity. These general characteristics, the factors involved, and the mechanism of the formation of microporosity have been described.[1] However, it will be well to review this phase of the problem before presenting the results of the investigation described herein.

Microporosity Formation Mechanism. The general factors involved are shown by Fig. 1, which represents any given section in a casting, vertical, horizontal, or inclined, fed by a riser, and chilled to promote directional solidification. Although the section shown by Fig. 1 is vertical, the principles are the same regardless of the shape or position of the section in the casting.

When the metal is poured through the sprue and ingate (Fig. 1), the metal gradually rises in the mold cavity. At the instant the mold cavity is filled, the hottest metal and hottest portion of the mold cavity are at the bottom adjacent to the gate, while the coldest metal and coldest part of the mold cavity are at the top of the riser. This temperature gradient is the reverse of that which is desired.

It is necessary to make this particular section solidify from the bottom toward the riser. To accomplish this it is necessary to reverse the temperature gradient which, in this particular instance, is effected by the use of a riser at the top and a chill at the bottom. The temperature gradients are thus reversed, and the casting begins to freeze adjacent to the chill, solidification proceeding upward toward the riser.

At any particular time during the process of solidification of the casting, the section will be solid at all temperatures below the solidus temperature, t_s (Fig. 1). Likewise, the section will be entirely liquid above the liquidus temperature t_L.

Between these two temperatures, the section will be partially solid, and this portion of the metal is referred to as the "mushy zone." The liquid content grades from 100 per cent at t_L to zero per cent at t_s. The greater the length of this mushy zone, the greater the difficulties of compensating for the solidification shrinkage.

Hence, there may be some temperature such as t_c (Fig. 1), below which the mushy zone contains such small liquid channels that it is impossible for the liquid metal to feed through them and entirely compensate for the solidification shrinkage of the alloy.

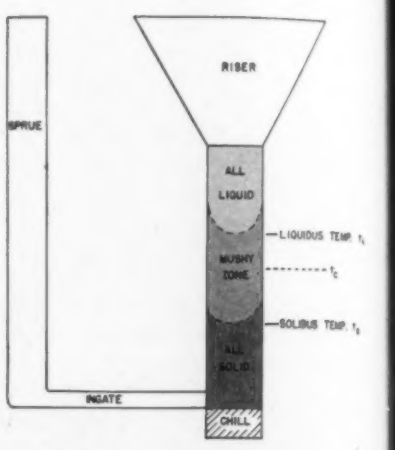


Fig. 1—Sketch showing formation of a mushy zone during solidification.

Crystallization Range

All other factors the same, the greater the length of the zone lying between the temperatures t_s and t_c, the greater the amount of microporosity which will be formed; and the greater the temperature interval between t_s and t_c, the greater the total liquid contraction.

If gas is precipitated from the melt during solidification, it fills the voids and greatly hinders the flow of melt into these small cavities. If the gas content of the melt were nil, the vapor pressure in the small cavities would be equivalent to the vapor pressure of the metal which would, of course, be very low. The small voids would then be essentially a

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high vacuum and atmospheric pressure would help to force the liquid metal into them.

It is logical, therefore, that the greater the gas content of the melt, the greater the tendency to form microporosity. If the gas content of the melt is very high, some of the gas will be precipitated early in the process of solidification and macrovoids will be formed, such as the well-known pinholes in aluminum and copper-base alloys and the larger gas holes produced in magnesium alloy at very high gas content.

It is also logical that the length of the mushy zone, all other factors the same, will be more pronounced the greater the solidification range of the alloy and the lower the amount of eutectic liquid formed. Consequently, solid solution type alloys, which have little or no eutectic liquid solidifying at the end and which generally solidify over a considerable range in temperature, are the most prone to the formation of microporosity.

Thus, the occurrence of microporosity will depend to a considerable extent upon the alloy composition, since the composition will largely determine the amount of eutectic and range in solidification.

It is also evident that the length of the mushy zone will be shorter the steeper the temperature gradient in the solidifying section. Much more metal is provided in the riser than is required to feed the voids resulting from solidification shrinkage.

The principal function of the riser is as a heat reservoir which will establish temperature gradients increasing toward it. This problem of providing adequate temperature gradients toward the risers involves the entire one of proper gating, risering, and chilling the casting to produce sound metal.

Experimental Work

It is thus evident that the principal factors which determine the occurrence of microporosity in a casting are (1) the alloy composition, (2) the gas content of the melt, and (3) the steepness of the temperature gradient toward the feed reservoirs. The interrelation of these three factors to the occurrence of microporosity will now be considered.

The investigation of the effect of the steepness of the temperature gradients has been limited to the determination of the effect of section thickness and of riser size on horizontally cast plates gated and risered on one end. Investigation of the effect of alloy composition has been restricted to a comparison of the four most common American and European casting alloys.[5] These alloys have the nominal composition shown in Table 1.

Table 1

NOMINAL COMPOSITION OF THE ALLOYS INVESTIGATED

Alloy	—Components, per cent— Al	Zn	Mn
A-8	8.0	0.4	0.2
AZ-91	9.4	0.4	0.2
ASTM-AZ63*	6.0	3.0	0.2
ASTM-AZ92**	9.0	2.0	0.2

*Formerly designated ASTM-4.
**Formerly designated ASTM-17.

The gas content of the melts has been investigated over a range which might normally be expected to occur under commercial practice. None of the melts contained sufficient gas to cause the risers to rise, forming a convex surface, or to form microscopic voids in the casting. However, gas content was encountered which was sufficient to accentuate greatly the microporosity, and which did produce unsound metal over the entire vertical length of a wedge casting used to measure the gas content of the melt. Use of the wedge casting, illustrated by Fig. 2, has been described elsewhere.[2] In general, the lower the gas content of the melt the greater the height of sound metal in this wedge.

If the gas content of the melt is relatively high, microporosity will occur from the bottom to the top of this casting. As the gas content is decreased, the height of sound metal measured from the bottom of the wedge will increase until a maximum of about 3⅓-in. of sound metal is obtained with best quality melts.

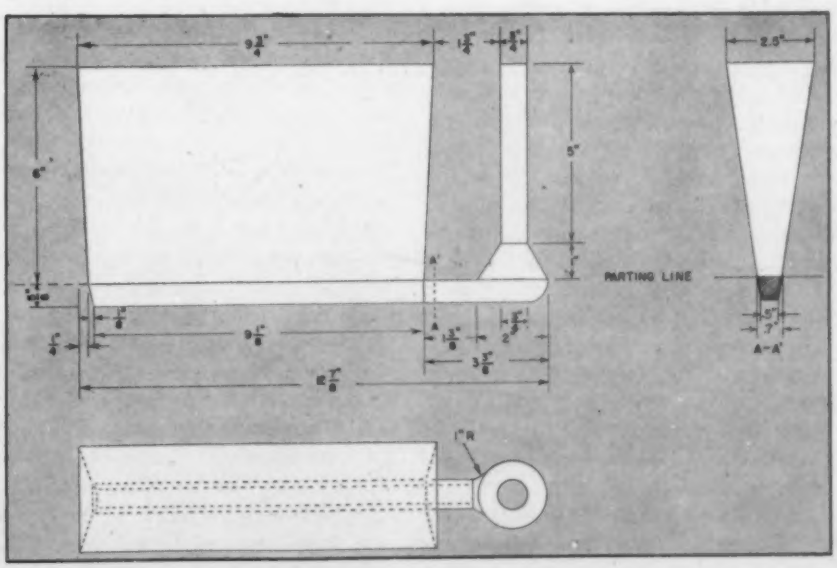
The height of sound metal is determined by cutting a vertical ⅜-in. section from the center of the wedge casting, facing it off, and x-raying the resulting section to show the occurrence of microporosity. Under these conditions, the x-ray is very sensitive and very small amounts of microporosity are revealed.

Relationship of Alloy Composition, Gas Content of Melt, and Section Thickness to Occurrence of Microporosity. One melt of each of the four alloys was prepared with a relatively high gas content by stirring unfused No. 310 flux into the melt. Four heats of ASTM-AZ63 and A-8, three heats of AZ-91, and five heats of ASTM-AZ92 were prepared and degassed by chlorine fluxing.

Grain refinement was obtained by various methods, including superheating and carbon inoculation with natural gas, and by the inclusion of carbon tetrachloride vapor in the chlorine, as described elsewhere.[3, 4] A typical synthetic sand was used with ethylene glycol, boric acid, and sulphur as the inhibitors.

In addition to the wedge casting

Fig. 2—Views of the wedge casting used in the investigation.



(Fig. 2), various plate castings 6 in. wide and 10 in. long of various thicknesses were cast horizontally. Plate thicknesses were 3/16, ⅜, ½, ¾, 1, and 1½ in., and all were gated on one end with the riser placed at the gate end. The 3/16-in. plate is shown in Fig. 3. The other plates were identical to this, except that the plate thickness was greater.

Pouring temperature was found to have little or no effect upon the occurrence of microporosity in these plates. Accordingly, the variations which occurred while pouring a series of molds were of little consequence to the results obtained. However, pouring temperatures were kept between 1500 and 1350° F., the highest temperature being used for the thinnest plates.

One wedge casting and the six 6x10-in. flat plates, varying in section thickness from 3/16 to 1½ in., were poured from each heat. Plates which are of ½-in. or more thickness are not suitable specimens from which to obtain radiographs that reveal accurately all of the microporosity present.

Consequently, such plates were sectioned lengthwise into three vertical, parallel sections. One of these vertical sections was cut from the center of the plate; the other two from midway between the center section and each edge. These sections were machined to ⅜ in. and then radiographed.

The 3/16- and ⅜-in. plates were radiographed in entirety; and such microporosity as was present tended to be widely distributed, i.e., it would occur at either end and generally across the plate. On the other hand, any microporosity in the ½-, ¾-, and 1½-in. plates would occur in the end adjacent to the riser, and would be more severe in the center section than in either of the two side sections.

A fairly satisfactory rating could be obtained by measuring the number of inches of sound metal in the center section. Since the plates were 10 in. long, the rating would be No. 10, if the entire length of the section was sound. If 4 in. of the center section were sound, the radiographic rating would be 4. If the center was unsound over its entire length, the radiographic rating would be 0, etc.

In all instances, the center section would exhibit more unsoundness than either of the two side sections. This is also true of the 3/16- and ⅜-in. plates which were radiographed over the entire surface. The length in inches of sound metal along the center line of these plates was also taken as the radiographic rating, as described previously.

However, since the microporosity tends to be more generally distributed across the 3/16- or ⅜-in. plates, a porosity rating of 3, for example, might contain more widely distributed porosity across the plate in a thin section than in a heavy section. The rating as described is simple and has proved to be quite satisfactory. The average results obtained on the 20 heats are shown in Table 2.

Table 2

EFFECT OF ALLOY COMPOSITION, GAS CONTENT,[1] AND SECTION THICKNESS ON AVERAGE[2] POROSITY RATING[3]

(Riser thickness at the base adjacent to the casting was twice the plate thickness. The top of riser was ½ in. thicker than base of the riser, and all were 6 in. high.)

Section Thickness, in.	Radiographic Rating of Alloys: ASTM-AZ63 Gassed	ASTM-AZ63 Degassed	A-8 Gassed	A-8 Degassed	ASTM-AZ92 Gassed	ASTM-AZ92 Degassed	AZ-91[4] Gassed	AZ-91[4] Degassed
3/16	0	2⅛	3	4¼	5	5½	10	7⅛
⅜	0	3½	1	8¼	0	8¾	10	10
½	0	6¼	0	6½	0	10	10	10
¾	0	6	0	9¼	0	10	10	10
1	0	5¾	0	9½	0	10	3	10
1½	0	½	0	3¼	0	9	0	10

[1] Gas content was indicated by the height of sound metal in the wedge casting. Gassed melts contained 0 in. of sound metal (see Footnote 4 below), and degassed melts generally produced over 2 in. of sound metal in the wedge.

[2] Only one gassed melt of each alloy was poured. Four heats each of ASTM-AZ63 and A-8 degassed melts, three heats of AZ-91, and five heats of ASTM-AZ92 were poured.

[3] Length in inches of sound metal in the center longitudinal vertical section of the plate (see text).

[4] This heat was only moderately gassy, as it produced 1⅛ in. of sound metal in a wedge casting.

Discussion

The conclusions from the data listed may be summarized as follows:

1. When the thickness of the riser is twice the thickness of the plate, the 3/16- and 1½-in. plates are the most difficult to make sound. The ½-, ¾-, and 1-in. plates are the most readily made sound, while the ⅜-in. plate is intermediate between this group and the 3/16-in. plate.

2. Using the degassed melts as a criterion, the order of alloy composition having decreasing susceptibility to microporosity is as follows: ASTM-AZ63, A-8, ASTM-AZ92, and AZ-91 alloys. This rating holds for plates of all thicknesses, and the difference between ASTM-AZ63 and A-8 compositions is larger than that among A-8, ASTM-AZ92, and AZ-91 alloys.

3. Degassing the melt has improved the soundness of the 3/16-in. section, but the improvement effected is far greater in the ⅜-in. and heavier sections.

4. While ASTM-AZ63, A-8, ASTM-AZ92, and AZ-91 alloys show decreasing susceptibility to microporosity in all section thicknesses, it will be noted that greater soundness can be effected in the 3/16-in. section by changing the alloy from ASTM-AZ63 to one that is less susceptible to microporosity than is obtained by decreasing the gas content of the melt in the range investigated.

5. It will be shown later that a ⅜-in. riser on a 3/16-in. plate is inadequate, and with degassed metal, the 3/16-in. section fed with a ¾-in. riser will, in general, be sound when poured in A-8, ASTM-AZ92, and AZ-91 alloys. The soundness of ASTM-AZ63, however, is increased but slightly if the riser on this 3/16-in. section is increased to ¾ in. or heavier. If then the 3/16-in. section had been fed with a 1-in. riser instead of a ⅜-in. riser, the data in Table 2 would have indicated a still further advantage of A-8, ASTM-AZ92, and AZ-91 over the ASTM-AZ63 composition.

6. It is only by a combination of a degassed melt and the selection of an alloy having the least susceptibility to microporosity that maximum soundness can be obtained for a given degree of feeding provided. It will also be shown that, in addition, adequate risering is necessary to produce substantially complete soundness.

Relation of Riser Size and Section

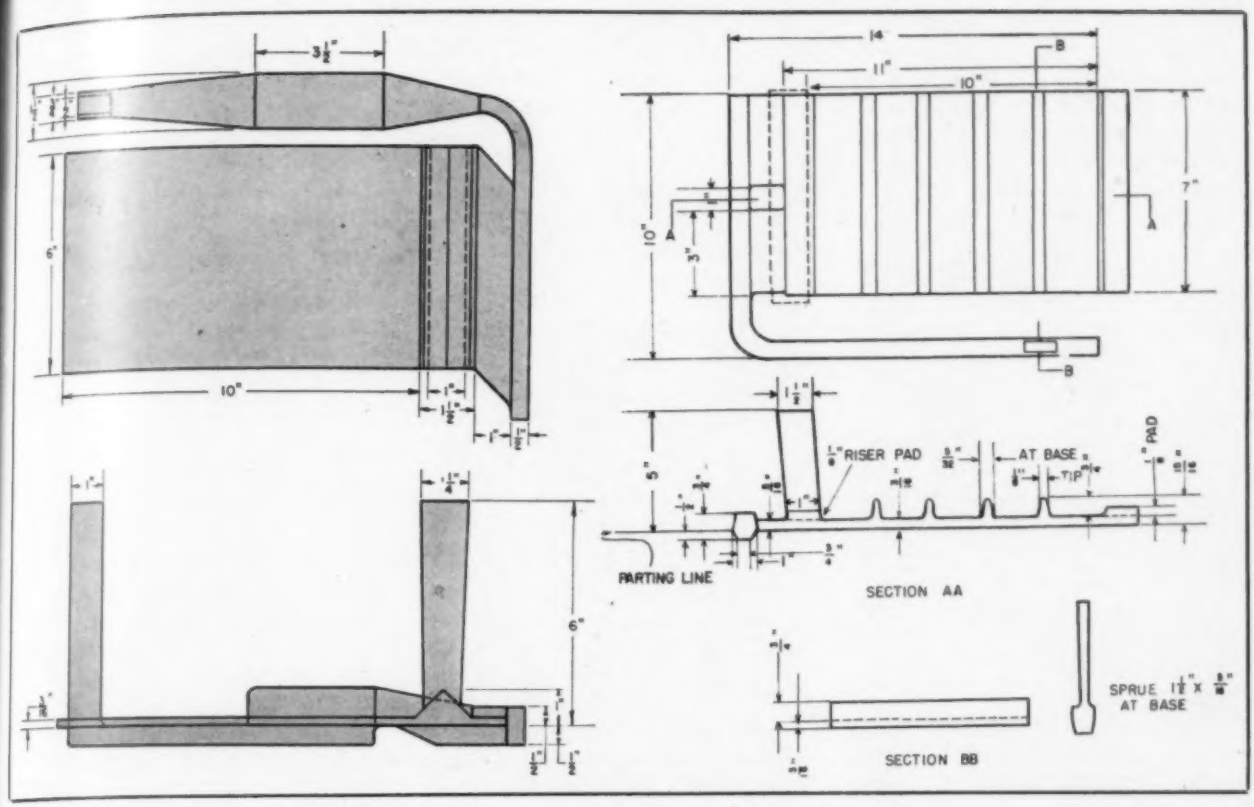


Fig. 3—Views of 6 x 10 x 3/16-in. plain plate test casting.

Fig. 4—Sectional views of 6 x 10 x 3/16-in. ribbed plate test casting.

Thickness to Occurrence of Microporosity in Castings of Four Different Alloys. The relationship of these factors to the amount of microporosity formed in castings of ASTM-AZ63, A-8, ASTM-AZ92, and AZ-91 was investigated. Four to six plates, similar to that shown by Fig. 3, were cast from each melt; but the plate thicknesses were varied from 3/16 to 1½ in. as described previously. Riser size was varied from zero to 3 in. at the base.

All risers extended across the width of the plate, were 6 in. high, and the thickness at the top of the risers was ½ in. greater than the thickness at the base. It should be emphasized that all of the plates were cast horizontally with the riser on one end and with the gate entering this riser.

This method of gating makes the temperature gradients all in the right direction at the instant the mold is filled. If it were necessary for the metal to flow through the mold cavity into the riser, as happens in many commercial castings, the results might vary somewhat from those reported here.

Each melt was thoroughly degassed by using chlorine or a mixture of chlorine and carbon tetrachloride and poured as wedge castings to indicate the melt quality. The considerable amount of sound metal in wedges poured from these heats indicated that all of the melts had a low gas content and, consequently, the results obtained are typical only of good quality, degassed melts.

Procedure

Nine heats of ASTM-AZ63 alloy were poured, producing a height of sound metal in the wedge of 2½ to 3½ in. Nine heats of A-8 alloy were poured in which the height of sound metal in the wedge casting ranged from 2 to 3½ in. Eleven heats of ASTM-AZ92 alloy produced 2¼ to 2⅞ in. of sound metal, and eight heats of AZ-91 alloy produced 2¼ to 3⅜ in. of sound metal in the wedge casting.

The 3/16- and ⅜-in. plates were radiographed in entirety and the ½-in. and heavier plates were sectioned and radiographed. These radiographs were rated from zero to 10, corresponding to complete unsoundness over the entire length to substantially complete soundness and freedom from microporosity. The results are listed in Table 3.

In some instances the porosity was very light or the microporosity was of the surface type only. Such surface microporosity apparently is associated with sand reaction and is not directly related to shrinkage or to the gas content of the melt. Accordingly, the data indicate the ratings in which a very light microporosity occurred as well as those ratings which are less than 10 and where only the surface type of microporosity occurs.

The portion of the data enclosed in parenthesis (Table 3) indicates the combination of plate thickness and riser thickness which will produce sound or substantially sound castings of each of the four alloys. The data listed in Table 3 indicate the following:

1. The 3/16-in. section is more difficult to make sound than the ⅜- or ½-in. section.

2. The 3/16-in. section can be made sound in A-8 alloy when the riser base is increased to ¾ in. and to ½ in. for ASTM-AZ92 and AZ-91 alloys. On the other hand, ASTM-AZ63 alloy cannot be made

sound in the 3/16-in. section regardless of riser size. The data tend to indicate that in the 3/16-in. section of ASTM-AZ63 alloy only a slight benefit is obtained by the riser, and certainly no additional benefit is obtained by increasing the riser base to a thickness greater than one in.

3. Table 4, based upon data taken from Table 3, indicates the minimum riser size required to produce substantially sound 6x10-in. plates of various thickness, cast horizontally and gated and risered on one end.

Investigations Using a 6x10x3/16-in. Ribbed Plate. A ribbed plate (Fig. 4) was also employed to check the relationship of alloy composition, gas content of the melt, degree of feeding provided, and the occurrence of microporosity in the casting. The distribution and amount of microporosity was determined by x-raying the entire plate.

This plate can be risered in various ways. Three-eighths-in. pads are attached to both ends of the plate, and a 1-in. riser may be placed at either end or at both ends. Thus, the relative merits of the following were determined: (1) gating into the riser, (2) passing the melt through the mold cavity to a riser at the end opposite the gate, (3) placing a 1-in. riser at each end, and (4) casting the plate without risers of any kind.

Two heats of ASTM-AZ63 alloy and one each of ASTM-AZ92, A-8, and AZ-91 alloys were prepared by degassing with chlorine, and then obtaining grain refinement by means of the addition of aluminum carbide. Another set of castings, one heat each of the four alloys, were prepared from metal which had been cleaned with No. 310 flux and then superheated.

Two heats of ASTM-4 alloy and one heat each of ASTM-AZ92, A-8, and AZ-91 were also prepared by cleaning the metal with unfused No. 310 flux, thereby producing gassed melts.

Along with the ribbed plates with no risers or risered in various ways, a wedge casting was also poured with each heat to determine the relative melt quality or gas content as measured by the height of sound metal in this wedge casting.

Examination of the ribbed plate by radiography revealed the relative effects of the degree of feeding provided, the alloy composition, and the gas content of the melt. The height of sound metal produced by the degassed melts varied from 1¼ to 3⅜ in. On the other hand, the gassed melts produced ⅛ to ⅞ in. of sound metal in the bottom of the wedge.

Radiographs shown elsewhere[6] are not reproduced here, but the results may be summarized as follows:

1. The degassing treatment, which has produced a considerable increase in the height of sound metal in the wedge casting of all four alloys, decreased only slightly the occurrence of microporosity in the 3/16-in. section of the plate.

2. Castings poured without risers represent the poorest degree of feed-

Table 3

RELATION OF RISER SIZE AND SECTION THICKNESS TO OCCURRENCE OF MICROPOROSITY IN CASTINGS OF FOUR DIFFERENT ALLOYS

	Section Thickness, in.	Thickness of Risers at Base**, in. 0	¼	⅜	½	¾	1	1½	2	3
ASTM-AZ63 Alloy	3/16	3	—	3, 4, 0	1	2,0	8,4,2½	4,3	3½	3½
	⅜	0, 0			0,0,3,4½	6½,4,2½	0,0	4,3.6	7½	(10)
	½						7¼,7½,2	—	8*	8
	¾						5½,5	6½,4½,2,6½,6½	—	8
	1								9*,4,1,4½,5½	0
	1½									0,0,1,1
A-8 Alloy	3/16	2½, 3		6, 3½	3	(10,10)	(10,7*,9*)	(10,10)		
	⅜	4, 5		10, 10	0*	(10,10,10,10)	(8*,2,8)	(9*)		
	½	10			2*,2	(10)	(10,10,8*,9*,10)	(10)		
	¾					2	8*,8†	(10,10,10,10)		
	1					0,3½	6½	8*	(10,10)	
	1½					—	0,1½	7½	1, 2	10,0
ASTM-AZ92 Alloy	3/16	10	10	10,10,3½,5,8*,7½*,4	(10)	(10)	(10,10,9*)	(9*)	—	
	⅜	0		(10)	(8½*,10)	(10,8*,10,10,5,10)	(10)	(10)	—	
	½			(10)	(10,8*)	(10)	(10,10,10,10,10,10)	—	—	
	¾				3	3	10,4¼	(8†,10,10,10,10,10,10)	—	
	1					3	3¾	(10,10)	(8†,10,10,10,10,10,10)	
	1½						1	3	2½	2,5,5,10,8,7½†,3½
AZ-91 Alloy	3/16	5, 5	7½*	8*,6½,7*,5,10	(10)	—	(9*,10)	(10)		
	⅜	10, 2	—	(10)	(10,5)	(10,10,10)	(10,10)	(10)		
	½	10	—	(10)	(10)	(10)	(10,10,10)	—		
	¾	—	—		5*	7*	(9*,10)	(5½†,10,10,10)		
	1	—	—			0	3,3	(10,10)	(10,10,10)	
	1½	—	—			—	0	0	1, 4	(10,10,10)

†Microporosity present is the surface type, probably associated with sand reaction and not with shrinkage or dissolved gas.
*Very slight amount of microporosity.
**Riser 6 in. high and ½ in. thicker at the top than at the base.

NOTE: Figures in parenthesis indicate the combination of plate thickness and riser thickness which will produce sound or substantially sound castings of each of the four alloys.

Table 4

Minimum Riser Size Required to Produce Substantially Sound 6x10-in. Plates, Cast Horizontally and Gated, and Risered on One End

	Section Thickness, in.					
Alloy	*3/16*	*⅜*	*½*	*¾*	*1*	*1½*
ASTM-AZ63	*	3	>3	*	*	*
A-8	¾	¾	¾	1½	2	>3
ASTM-AZ92	½	⅜	⅜	1½	1½	>3
AZ-91	½	⅜	⅜	1	1½	3

*Unsound with all riser thicknesses up to 3 in.

ing provided, and those risered at both ends represented the best fed castings. For a specific alloy composition, this range in the degree of feeding provided, however, has produced only a slight decrease in the amount of microporosity obtained in the 3/16-in. sections of the plate, regardless of alloy composition. Thus, risering both ends of the plate has slightly improved the soundness of the 3/16-in. portion of the plate. The ⅜-in. pads, however, must be fed with either a riser or a gate to avoid pronounced amounts of microporosity in this part of the casting.

3. Order of decreasing susceptibility of the alloys to microporosity in this ribbed casting was as follows: ASTM-AZ63, ASTM-AZ92, A-8, and AZ-91, there being a great difference between ASTM-AZ63 and ASTM-AZ92, and only a slight difference among ASTM-AZ92, A-8, and AZ-91. With one 1-in. riser, ASTM-AZ92, A-8, and AZ-91, the 3/16-in. ribbed plates were substantially sound, while ASTM-AZ63 was unsound over its entire area.

4. In this 3/16-in. ribbed plate, the alloy composition is much more important than the gas content of the melt within the considerable range of gas content employed in this portion of the investigation.

Results obtained on the 3/16-in. ribbed plate confirm those obtained on the plain, flat 3/16-in. plate.

Correlation of Theoretical Principles and Experimental Results. All of the foregoing data indicate that the ⅜- and ½-in. sections are not only easier to feed than the 3/16-in., but that they are also easier to feed than the heavier sections. When the section thickness reaches 1½ in., it is quite difficult to obtain sound metal with riser thicknesses twice the plate thickness.

This can be explained by the slow solidification rate in such heavy sections. It is obvious that, if a considerable time is required for solidification, the relatively high conductivity of the alloy will produce very slight temperature gradients toward the risers, producing a long mushy zone.

As indicated previously, the long mushy zone is not conducive to feeding, and the tendency to form microporosity is quite pronounced. On the other hand, the 3/16-in. sections solidify rapidly and almost simultaneously over their entire areas, making progressive solidification toward the riser and adequate feeding difficult.

In other words, under the casting conditions described, the ½-in. section has the optimum cooling rate whereby the most directional solidification toward the risers can take place, producing the shortest mushy zone.

It is clearly evident that ASTM-AZ63 alloy is markedly more susceptible to the formation of microporosity than the other three alloys. The susceptibilities to the formation of microporosity in A-8, ASTM-AZ92, and AZ-91 alloys are of the same order of magnitude, but the susceptibility decreases slightly in the order listed.

Differences in susceptibility to microporosity among the four alloys can be safely attributed to the dissimilarities in composition. The alloy composition determines the solidification range and the amount of eutectic liquid solidifying at the end.

Solidification range of the four alloys is as follows:[5] ASTM-AZ63, 475° F.; ASTM-AZ92, 345° F.; A-8, 315° F., and AZ-91, 300° F.

The amount of eutectic liquid solidifying at the end is indicated by the relative volumes of undissolved beta constituent in the as-cast condition; ASTM-AZ63 alloy has the least amounts of eutectic liquid, A-8 is next, ASTM-AZ92 third, and AZ-91 has the most eutectic liquid or undissolved beta constituent in the as-cast condition.

It is obvious then that ASTM-AZ63 alloy not only has the longest solidification range, but it also produces the least amount of eutectic liquid solidifying at the end. Consequently, this alloy is by far the most susceptible of the four alloys to microporosity.

AZ-91 alloy is less susceptible to microporosity than A-8 alloy because it has a slightly narrower solidification range, but the more important factor is the considerably greater amount of eutectic liquid which solidifies at the end. ASTM-AZ92 alloy has a solidification range markedly less than ASTM-AZ63, but considerably greater than A-8 and AZ-91 alloys.

Because ASTM-AZ92 alloy has a greater solidification range and a lesser amount of eutectic liquid solidifying at the end than has AZ-91, the former is slightly more susceptible to occurrence of microporosity.

On the other hand, ASTM-AZ92 alloy has a greater range of solidification, but the amount of eutectic liquid solidifying is slightly greater than that of A-8. As a result, A-8 is only slightly more susceptible than ASTM-AZ92 to the formation of microporosity.

Liquidus and Solidus Contours

A third factor exists which is not entirely indicated by the consideration of the solidification range and the amount of eutectic liquid as represented by the relative volumes of the beta constituent in the as-cast condition. This factor pertains to the shape of the solidus and liquidus lines.

The shape of the solidus and liquidus surfaces in the complex alloys determines the relative amount of solid and liquid at any given temperature intermediate between the solidus and liquidus (Fig. 5).

If the contour of the solidus and liquidus surfaces is such as to permit a relatively large proportion of liquid to exist to the lower part of the solidification range, the alloy will be less susceptible to the formation of microporosity than if the contour of solidus and liquid surfaces are such as to produce a relatively greater amount of solids in the higher portion of the solidification range.

Examples of these two different types of solidus and liquidus systems are represented by Fig. 5. This third factor is fairly difficult to evaluate. However, it is quite probable that

the contour of the solidus and liquidus surfaces, which would be favorable to the retention of a relatively large proportion of liquid to the bottom of the solidification range, would also be reflected by a relatively large amount of eutectic liquid solidifying at the end. This is illustrated by Fig. 5.

It has already been noted that the proportion of eutectic liquid can be estimated from the relative volume of undissolved beta constituent in the as-cast condition. This third factor, therefore, is not entirely separate from the factor relating to the relative amount of eutectic liquid solidifying at the end of solidification.

In any event, the known facts about the solidification range and the relative amounts of eutectic liquid solidifying explain quite satisfactorily the relative susceptibility of the four different alloys to the formation of microporosity.

Effect of Pouring Temperature. The 6x10-in. plates, 3/16, ½, and 1½ in. thick, were used to determine the effect of pouring temperature on the occurrence of microporosity. These plates of all four alloys were poured over a range from 1500° F. to the lowest temperature at which they could be poured without serious misruns.

In some instances, when an extremely coarse grain was obtained at pouring temperatures near 1250° F., the microporosity appeared to be accentuated; otherwise the pouring temperature had no noticeable effect on the occurrence of microporosity.

This conclusion is somewhat at variance with foundry experience since, under production conditions, it is frequently found that high pouring temperatures produce less microporosity than lower pouring temperatures. Some observations in the laboratory indicated that, when chills were used to produce solidification, the tendency to form microporosity decreased somewhat with increasing pouring temperatures.

Commercial castings, of course, are not always gated into risers, and chills are used frequently to effect progressive solidification toward the risers. Under the conditions of a higher melt temperature in the mold in conjunction with chills and complex sections, it is quite conceivable that there will be sharper temperature gradients increasing toward the risers, which condition will tend to decrease the amount of microporosity formed.

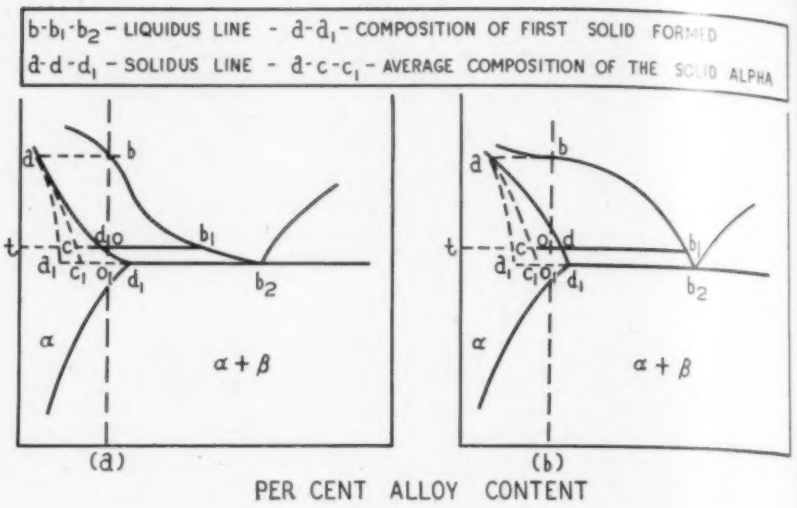


Fig. 5—Intersections of liquidus and solidus surfaces showing two compositions having same solidification range and eutectic temperature, but at temperature (t), system (a) contains 26 per cent liquid while (b) contains only 7 per cent. At the eutectic temperature, (a) contains 13 per cent eutectic liquid, while (b) contains 6 per cent. Alloy (a) will be less susceptible to microporosity because a greater proportion of liquid is retained to lower temperatures. This is also reflected in a greater amount of eutectic liquid and beta constituent, although both have the same maximum solid solubility at the eutectic temperature.

It is to be expected that the experimental plates risered and gated at one end would have a favorable temperature gradient initially, which would not be altered greatly by an increase in melt temperature.

Effect of Mold Material. The wedge casting (Fig. 2) and the 3/16-in. plates (Fig. 3) were prepared in green sand and in a dried sand. The green sand used was a regular synthetic sand containing bentonite for the binder, ethylene glycol, sulphur, and boric acid as inhibitors.

This green sand baked well to produce a dried mold. Accordingly, molds of the same sand were baked at a low temperature to prevent serious loss of inhibitor, but at a sufficiently high temperature to remove the water from the sand.

Approximately six melts of ASTM-AZ63 and ASTM-AZ92 alloys were prepared. One set of molds was poured before fluxing with chlorine, and a second set was poured after fluxing with chlorine. Each set consisted of one green sand 3/16-in. plate, one green sand wedge mold, one dry sand 3/16-in. plate, and one dry sand mold of the wedge.

Careful examination of the plate and wedge castings by radiography showed that the mold materials had no influence upon the amount of microporosity in either the plate or wedge castings. The chlorine flux improved the soundness of both.

These observations indicate that gas is not absorbed by the melt in the mold prior to its solidification. It has been established, for example, that gas will be absorbed by melts of aluminum-base alloys if they are poured into green sand having low permeability.

All sand used in this investigation has high permeability, and it is not to be expected that gas would be absorbed by the melt in the mold. Apparently gas is absorbed by the melt only when the permeability of the sand is sufficiently low to force some of the generated vapor into the liquid melt in the mold cavity.

Summary

Mechanism of the formation of microporosity has been described. The three main factors determining the amount of microporosity are (1) gas content of the melt, (2) alloy composition, and (3) degree of feeding provided. The range of alloy composition investigated was that included by four alloys: ASTM-AZ63, A-8, ASTM-AZ92, and AZ-91.

Range in degree of feeding was

investigated over a range in section thickness including 3/16, ⅜, ½, ¾, 1, and 1½ in., while riser sizes on these sections ranged from zero to 3 in. The relative effects of the three main factors determining microporosity in sections of various thickness may be summarized as shown in Table 5.

Table 5

Relative Effects of Factors Determining Microporosity

Factor	*Effect on Microporosity for Section Thickness, in.*		
	3/16	*⅜ to ½*	*1 to 1½*
Decreasing gas content	Slight Decrease	Marked Decrease	Marked Decrease
Increasing riser size within normal limits	Slight Decrease	Marked Decrease	Marked Decrease
Changing alloy composition from ASTM-AZ63 to A-8, ASTM-AZ92, or AZ-91	Marked Decrease	Marked Decrease	Marked Decrease

If the initial gas content of the melt is very high and the melt is degassed by fluxing it with chlorine, the reduction in microporosity in the light 3/16-in. section will be pronounced. As indicated in Table 5, the decreasing order of susceptibility to microporosity among the four alloys is as follows: ASTM-AZ63, A-8, ASTM-AZ92, and AZ-91. A large difference exists between ASTM-AZ63 and A-8, and only a slight difference exists among the other three alloys.

The ⅜- and ½-in. sections are the easiest to make free from microporosity. Maximum soundness can only be obtained by using (1) degassed melts, (2) alloys least susceptible to the formation of microporosity, and (3) adequate feeding.

Adequate feeding is obtained by the establishment of steep temperature gradients toward the riser or feed head. This is done by the proper use of gates, risers, and chills.

For a given degree of soundness, the ratio of poured weight to casting weight can be reduced by using (1) degassed melts and (2) an alloy in place of ASTM-AZ63 having less susceptibility to the formation of microporosity.

Pouring temperature had little or no effect upon the amount of microporosity formed in these test castings, which were 6x10-in. plates cast horizontally with gate and riser on one end only. If, however, chills were used in the mold and if the gates did not enter the risers, it is quite likely that the microporosity may decrease somewhat with increasing pouring temperature.

It was also found that the amount of microporosity formed was approximately the same in dried molds as in green sand molds, indicating that gas is not absorbed by the melt in the mold cavity.

Acknowledgment

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INSULATING PADS AND RISER SLEEVES FOR BRONZE CASTINGS

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NON-FERROUS FOUNDRY practices have been shown, particularly by wartime production experiences, to be in need of an extensive investigation. Increased emphasis upon inspection and the development of improved and new inspection tools focused attention upon serious shortcomings in castings. More strenuous performance requirements caused an increased number of service failures, and often the replacement casting proved to be little or no better than the original.

Ever-increasing shop production demands and shortage of manpower may have caused a relaxation of otherwise adequate foundry practices in some shops, but even prior to the war the situation pointed to the need for improvement in non-ferrous foundry practices. By far the greater number of service failures have been in castings operated under pressure. The reasons given for failure have been many and varied, and often depend more upon the individual interpretation than upon any factual evidence.

Sometimes the foundryman blames gassy metal caused by improper melting or wet molds, sometimes impurities or improper composition are blamed, and occasionally shrinkage caused by inadequate feeding is determined to be the trouble. Any one, or all, of these conditions may cause bad castings, but it seems probable that the greatest offender is shrinkage in one or another of its many manifestations.

Research upon the gating and risering of steel castings indicates definite limitations upon the ability of risers to feed castings adequately unless great care is taken to position and proportion gates and risers properly. The type of defect most apt to cause leakage of steel castings under pressure probably is shrinkage, because of the interconnected, filamentary voids.

Gas porosity, unless sufficiently extensive to be readily recognized as such, usually in the form of rounded, disconnected voids, would not necessarily cause leakage. Sections of non-ferrous castings which have failed indicate lack of internal soundness which, more often than not, resemble similar sections taken from steel castings which have been made without proper attention to known laws of solidification.

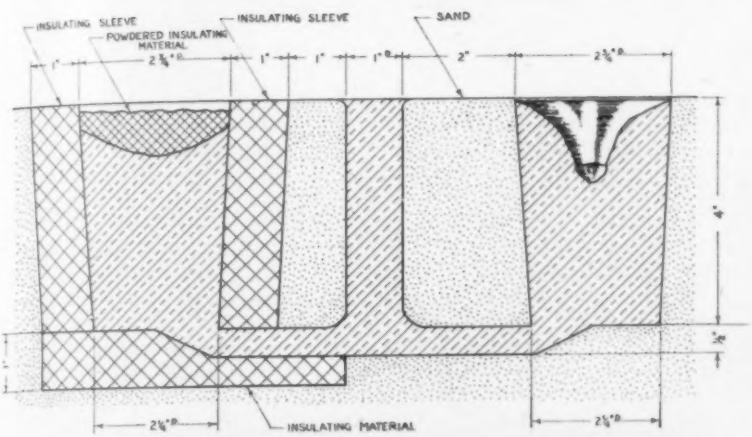
New foundry techniques developed at the Naval Research Laboratory, Office of Research and inventions for production of sound non-ferrous castings, serve to emphasize dependence of good foundry practice upon proper gating and risering. Insulating pads and riser sleeves placed respectively on their sections and around risers keep the metal in these regions hot and molten until the casting has become solid. Applications of the method show the importance of controlled directional solidification. The opinions expressed are those of the authors and do not necessarily reflect the opinions of the Navy Department.

Further evidence in support of the theory that shrinkage in one form or another, if not repaired by welding, is responsible for service failures in steel castings is a survey made by the Westinghouse Electric Co. in which shrinkage was definitely branded as the worst problem. This condition exists in spite of all the research which has been done on this problem in steel castings. It is reasonable to suppose that a similar condition may exist in non-ferrous castings, upon which relatively little fundamental research has been done.

Studies at the Naval Research Laboratory upon the effect of atmospheric pressure as a potent factor in making sound castings and, oddly enough, as an even more potent force in causing bad castings, have led to a realization of the many ways in which gating and risering can affect the end-product of the foundry. For example, serious blows at internal corners of castings, exaggerated areas of low density, and so-called "pulls," "draws," or "dishing" of castings are results of atmospheric pressure acting upon an improperly risered casting.

Certain surface defects, particularly on cope surfaces, discontinuities which appear to be tears caused by hard molds, small cavities beneath vents and in front of ingates, and many other defects normally not associated with shrinkage, more often than not are a direct result of improper gating and risering practices. A natural tendency exists on the part of non-ferrous foundrymen to use as few gates and risers on a casting as possible because their re-





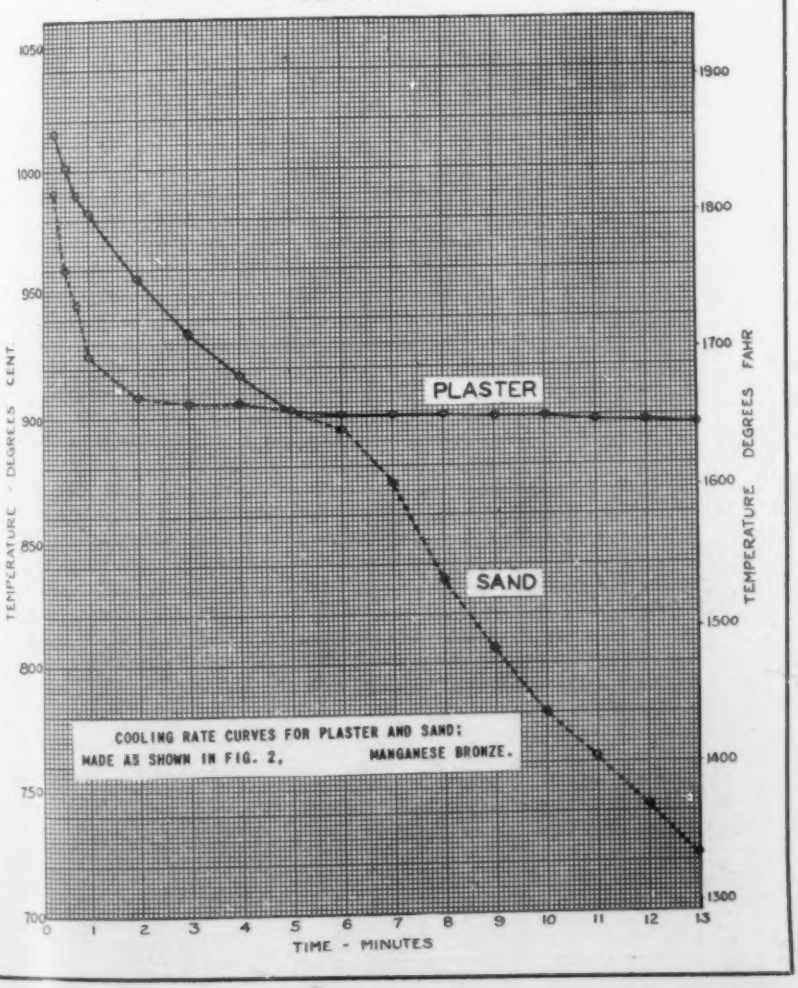


Fig. 1 (left)—Test ingots used to determine relative rates of solidification of manganese bronze in green sand and in gypsum; pouring temperature 1025° C. (1880° F.).

moval from the casting is a serious problem.

It is not possible to use torches for this purpose, as can be done with steel castings, and their removal by saw or cold chisel is laborious and expensive. A balance must be made between the selling price of the product and the amount of work that can be done in the machine shop in order to meet competition.

This factor, rather than good engineering, too often determines the size, number, location, and design of risers, and inadvertently affects adversely the soundness of the casting. The foregoing factors indicate the need for fundamental information on the solidification of nonferrous castings.

General Considerations

In 1928 a project was authorized at the Naval Research Laboratory by the Navy Department to study the fundamental defects in steel castings. Since that time a great deal of practical information has been compiled on the solidification characteristics, gating and risering, and general foundry practices used for steel and the principles are quite generally appreciated and used. Much of the information should apply to other metals but, unfortunately, relatively little effort has been made to determine if this is the case.

The principles of and need for controlled directional solidification in castings have been recognized for many years by steel foundrymen, and numerous articles have been published on the subject [1, 2, 3, 4]. The theory is simple but often difficult to apply in practice.

It means only that in order to obtain maximum benefit from feed heads and the soundest possible casting it is necessary to have the coldest metal within the feeding range of a given feeder at a point farthest

Fig. 2 (center and bottom)—Sketch showing molding technique used in solidification test of casting shown in Fig. 1 and the location of the thermocouples (T). Cooling rate curves for manganese bronze casting made as shown in sketch.

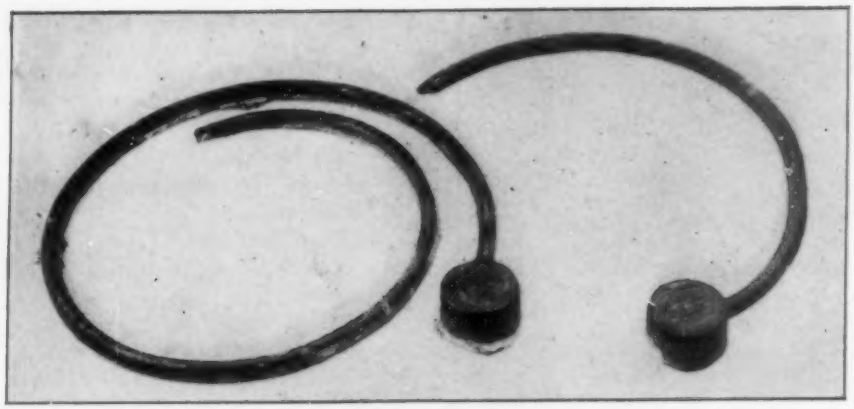


Fig. 3—Fluidity test spirals of ingot copper; pouring temperature 1175° C. (2150° F.). Left—plaster, 34 in. Right—green sand, 15 in.

from the riser, and that the metal be progressively hotter toward the riser at all times during solidification. The normal way to obtain proper directional solidification in a casting is to taper sections slightly toward the riser, being sure that the taper is progressive in all respects and that intermediate thin sections are not present which would solidify prematurely and choke off feeding.

It is not possible to feed heavy sections through light sections, and chills or auxiliary feed heads must be provided whenever the casting is so designed as to limit the feeding range of the main risers. In other words, the "ingot principle" must be used for each and every riser independently, and an adequate number of risers must be provided to feed the solidifying casting.

This principle should be as important in the manufacture of non-ferrous castings as it is for steel, and considerable evidence is available to indicate that such is the case.

Influence of Atmospheric Pressure

Atmospheric pressure as a factor in foundry techniques has been found to have marked influence on the integrity of steel castings, and principles governing its behavior have been carefully determined. It is interesting that, with a proper understanding of the manner in which atmospheric pressure operates upon a body of solidifying metal, sound castings can be produced, while without this knowledge unsound castings often result.

It is surprising to note that little or no work has been published, and apparently little notice given this important principle by non-ferrous foundrymen, although the influence can be even more far-reaching than is the case in steel casting.

An appreciation of the manner in which atmospheric pressure exerts a profound influence in foundry practice can be gained by a brief resume of everyday principles of physics. It is well known that atmospheric pressure acts on all objects to which it has access with a force equal to approximately 14.7 psi.

This is a force capable of causing vessels with structurally weak walls to collapse when the pressure within them has been reduced to near vacuum levels, and in the mercury barometer to force mercury to a height of about 30 in. under similar conditions. What is more important to the problem at hand, it is the force which often determines whether sound or unsound castings will be made.

Any casting, regardless of size, shape, or contour can be used to illustrate the point. Atmospheric pressure acts equally on all parts of the casting, and it has been carefully determined that a sand mold is sufficiently permeable to permit this. As the casting solidifies and the metal contracts in volume a tendency for

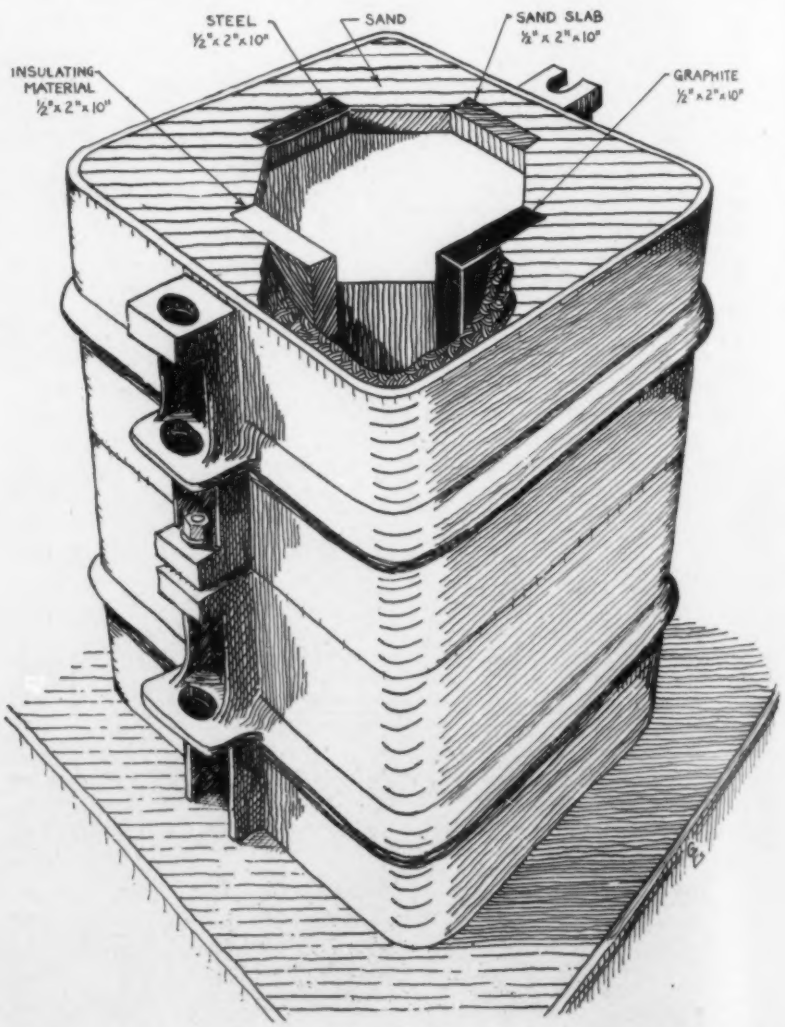


Fig. 4—Sketch of test mold for casting shown in Fig. 5.

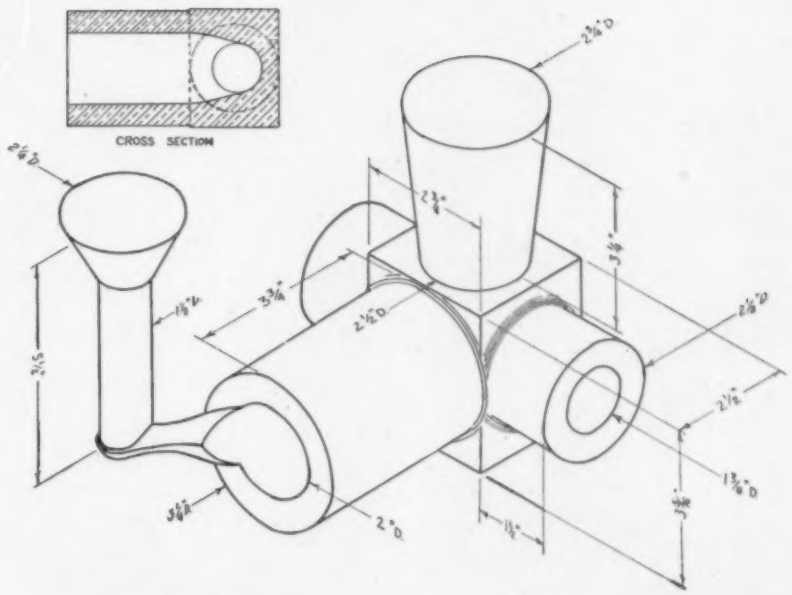


Fig. 7 (right)—Sketch of hollow tee casting showing dimensions and cross section.

a partial vacuum to form within the casting exists.

If liquid contact is made at all times with the riser and progressive solidification maintained, the tendency for a vacuum to form will be offset by the liquid metal which moves in to accommodate the shrinkage. However, if for some reason feeding is not adequate a shrinkage cavity will form, a partial vacuum will be created, and a difference in pressure between the inside and outside of the casting will exist.

The magnitude of this difference will depend upon the amount of unfed shrinkage and, whenever it becomes greater than the strength of the solidified skin of metal around the casting, the walls will collapse or puncture at a weak spot to relieve this condition. Castings with extensive flat sections such as plates or large flanges on valves and fittings, which are difficult to riser adequately, are especially prone to "dishing," while castings with internal corners usually puncture through at the hot spot.

It has become general knowledge that if risers are to function properly they must be kept open to the atmosphere in order for this added pressure to aid in feeding. This is accomplished in ferrous castings by use of pipe eliminator compounds to cover open risers, or by the use of a core of some suitable material cast into the cavity of blind risers.



Fig. 5—Section from center of manganese bronze casting; bled after 5 min.; poured at 990° C. (1810° F.).

Non-ferrous foundrymen generally have not taken precautions to insure that the risers remain open to the atmosphere at all times. Often sand is sprinkled on open risers for insulation, a practice which does not appear desirable because of the high specific heat of sand which causes it to absorb heat rather than to insulate, and which causes the riser to freeze prematurely. If risers are not kept open feeding may be materially reduced and atmospheric pressure may gain entry to the casting at some other point.

This brief treatment of a really extensive subject should suffice to indicate that a thorough knowledge of atmospheric pressure, as it applies in the foundry, is essential to the making of good castings, either ferrous or non-ferrous. A paper by Taylor and Rominski[5] entitled "Atmospheric Pressure and the Steel Casting" is a comprehensive treatment of the subject as it applies to steel, and no attempt will be made here to discuss the development in detail since fundamental principles are therein clearly covered.

Attempts to make sound shop castings from copper-base alloys at the Naval Research Laboratory indicated the need for even greater attention to gating and risering than was necessary for steel castings. Risers often have to be larger and more numerous than for steel and,

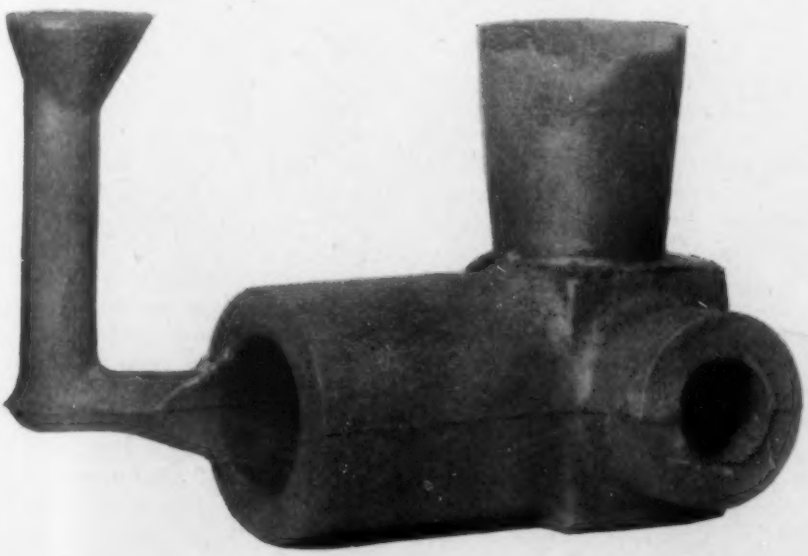


Fig. 6—Hollow tee casting used for testing insulated riser sleeves and insulating pads.

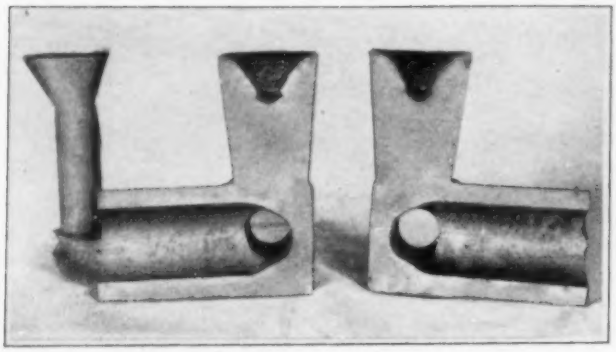


Fig. 8—Hollow tee casting made entirely in a sand mold. Note shrinkage. Manganese bronze; pouring temperature 1000° C. (1830° F.).

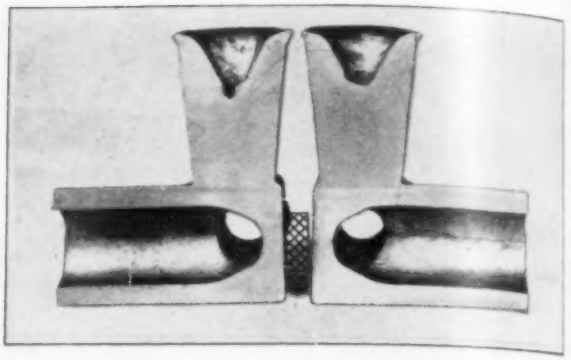


Fig. 9—Hollow tee casting made in sand with an insulating pad placed as shown. No shrinkage present. Manganese bronze; poured at 1000° C. (1830° F.).

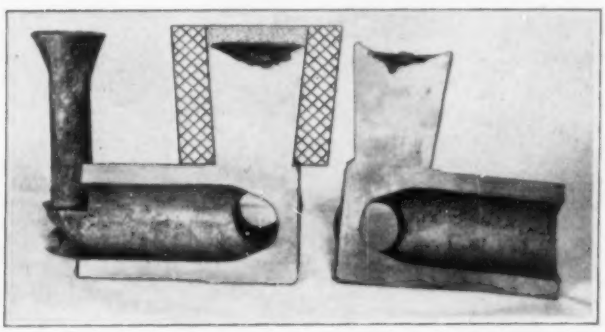


Fig. 10—Hollow tee casting made in sand with an insulated riser as shown; no insulating pad; insulating powder used on top of riser; small amount of shrinkage present; manganese bronze; pouring temperature 1000° C. (1830° F.).

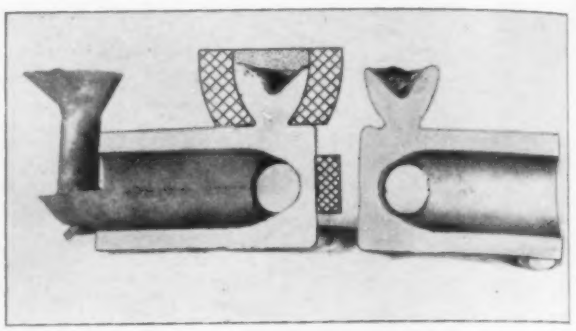


Fig. 11—Hollow tee casting made in sand with pad and riser insulation and insulating powder on top of riser; manganese bronze; pouring temperature 1000° C. (1830° F.).

in general, the temperature gradients within the casting must be more carefully controlled.

Although tin bronzes show markedly less shrinkage, as manifested by riser cavities, than does manganese bronze, it was found that gating and risering for a sound casting in tin bronze was even more difficult than for a casting in manganese bronze.

In order to combat the interdendritic type of shrinkage characteristic of tin bronze, risers have to be very large, and temperature gradients carefully controlled by gating into or through risers as much as possible so that the metal in the riser is kept molten and molten metal contact is maintained with all solidifying portions of the casting until they are completely solid.

In spite of this it is general practice to gate tin-bronze castings into thin sections away from risers and to use few and small risers as compared to those used for similar manganese-bronze castings. This is due to the fact that such a practice distributes shrinkage uniformly and it is not readily noticeable upon casual examination. The use of risers which seem large to foundrymen, and yet are really not large enough for their intended purpose, serves only to localize shrinkage underneath them, shrinkage which can be readily found.

Gating under or through inadequate risers aggravates the condition. However, adequate risers do give sound castings, even in tin bronzes, particularly when temperature gradients are favored by gating through them, and it is not possible to make perfectly sound castings by circumventing this principle. The demand upon risers can be lessened and castings made very sound by an extensive use of chills, a practice which is rapidly gaining favor.

For many years attempts have been made to develop a suitable material from which to make insulating pads and riser sleeves for steel castings as an aid in promoting proper temperature gradients within a solidifying casting. This is common practice for ingot making in which sand is an effective insulator relative to the metal of the ingot mold.

However, no suitable material has yet been found for steel castings, and the closest approach to effective riser insulation is the use of powdered compounds placed on the surface of the metal in the riser to decrease radiation of heat and to insure against premature solidification.

Except as a means of keeping the risers open to the atmosphere, riser compounds are relatively ineffectual as their influence does not extend to the base of the riser where insulation is most needed in order to prevent early solidification at neck.

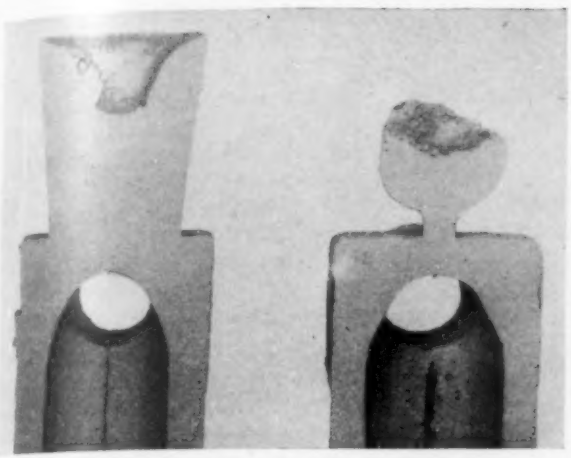


Fig. 12—Hollow tee castings poured vertically. (Left)—in a sand mold and with uninsulated riser; (right)—in a sand mold with an insulated riser; total weight of uninsulated riser casting—18½ lb., insulated riser casting—13½ lb. Manganese bronze.

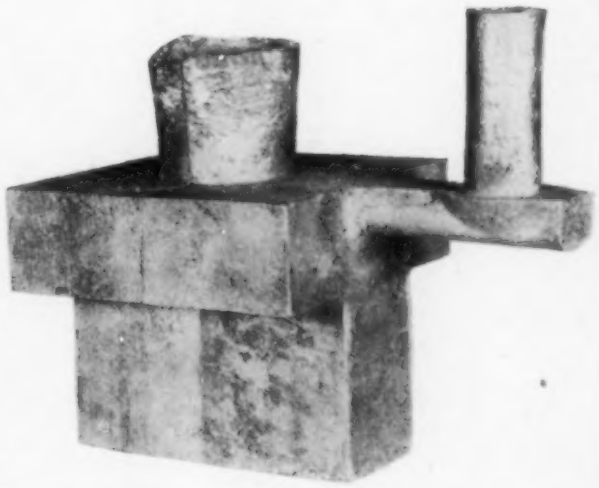


Fig. 13—Photograph of solid tee test casting.

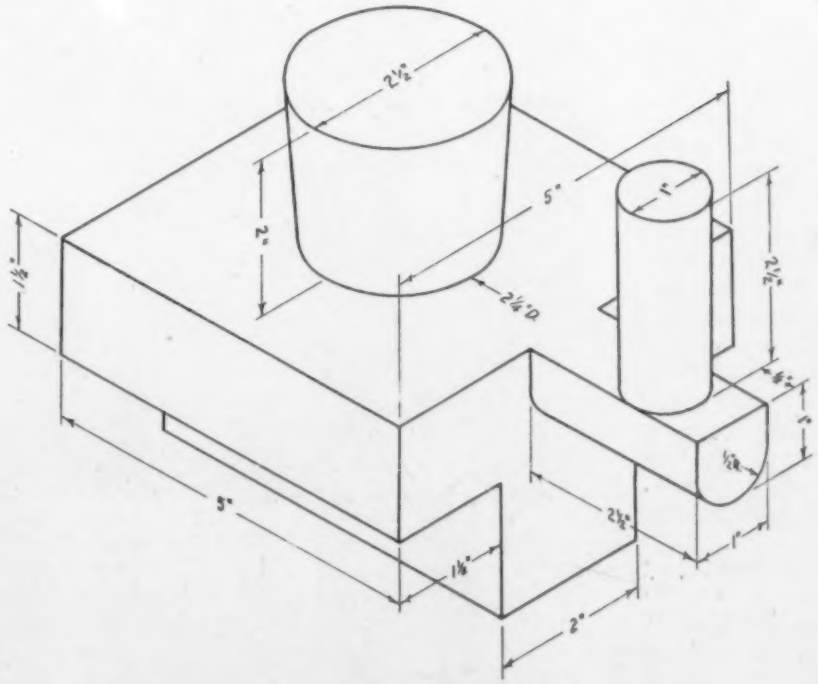


Fig. 14—Sketch of solid tee test casting showing dimensions.

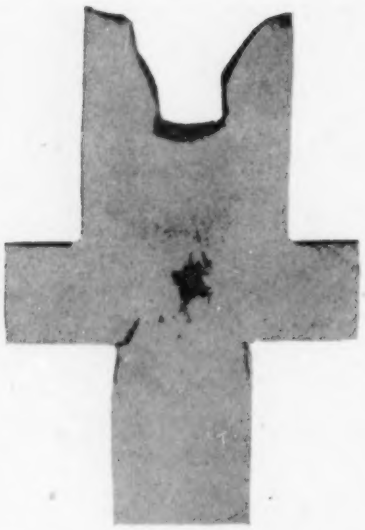


Fig. 15—Section of solid tee casting made entirely in sand, showing effect of atmospheric pressure and inadequate riser; manganese bronze; poured at 1830° F.



Fig. 16—Section of solid tee casting made with an insulated riser sleeve, showing a completely sound casting, manganese bronze; poured at 1000° C. (1830° F.).

The commercial use of gypsum as a mold material for making non-ferrous castings and its well-known insulating effect when used for this purpose suggested it as a material with which to accomplish in the non-ferrous field that which has not been as yet satisfactorily accomplished for casting of steel. Gypsum cannot be used with ferrous metals because of the reaction between the sulphur of the gypsum and the iron.

Figures 1, 2, 3, and 5 illustrate practical examples of the insulating value of gypsum when used in a mold for casting copper-base alloys. Figures 1 and 2 show the first experiment made to determine the relative rates of heat transfer through sand and gypsum molds. One half of the mold was made completely in gypsum and the other half in green sand. The casting cavities were gated identically, were of identical dimensions, and thermocouples were placed so that the temperature would be measured in the center of each mold.

The metal used was manganese bronze poured at 1875° F. (1025° C.). The time for complete solidification as indicated by the thermocouples was more than twice as long in the insulated metal as in the uninsulated metal (Fig. 2 chart). It is interesting to note the difference in temperature of the metal at the time of the first reading, indicating immediately the relative insulating properties of the gypsum, and the much greater absorption of heat by sand as the metal flowed over it.

It will be noticed from the curve that the liquid metal inside the insulated casting approaches the solidus range much slower than it does in the sand casting, and a supply of liquid metal is available for a longer period of time to feed the shrinkage of a solidifying casting.

Fluidity spirals are used at the Naval Research Laboratory as a means for indicating the flow behavior (fluidity) of molten steels. The length of flow is taken as a relative index of the fluidity of the metal. The spiral length is really an indication of how far the metal will flow before enough heat is lost to the mold to cause solidification.

Figure 3 shows a pair of such spirals, one poured in a gypsum mold and the other in sand. The relative length of 34 in. for the gypsum and 15 in. for the sand indicates in a rough way the relative insulating values of the two materials. Pure copper was used for the experiment. Both spirals were poured at 2150° F.

Figures 4 and 5 illustrate still further the good insulating properties of gypsum. Figure 4 is a sketch showing how the casting was made. An octagonal cavity in a green sand mold was filled by pouring molten manganese bronze through the open top. The metal was covered with an insulating powder after filling the mold.

As shown in the sketch, each alternate face of the octagon was formed with a slab of the material to be tested. At the end of 5 min. the cope and drag molds were separated and the unsolidified metal was bled out. Figure 5 shows a section through the center of the casting, and the lack of solidification on the gypsum face is clearly indicative of the insulative properties of this material.

This test demonstrates how effec-



Fig. 17—Section of solid tee casting made entirely in sand; etched surface shows shrinkage at internal angles; "G" bronze; poured at 1150° C. (2100° F.).



Fig. 18—Section of solid tee casting made with an insulated riser; etched surface shows casting sound. "G" bronze; pouring temperature 1150° C. (2100° F.).

tively several of the well-known materials can be used to promote the vitally important directional solidification within a mold cavity. The two contrasting materials were steel and gypsum, the former producing a wall thickness of 1¾-in. while the latter had retarded the heat flow so drastically that no skin had formed after the 5-min. period. From this experiment alone it can readily be seen that by using these materials in conjunction with conventional mold mediums a much greater control of directional solidification is available.

In order to test the practical value of pads and riser sleeves made from gypsum two typical castings were selected, one involving cored holes and thin sections and the other a solid mass containing troublesome internal corners. Figure 6 is a photograph of an actual casting and Fig. 7 illustrates schematically the molding technique used and the dimensions of the casting.

In Figs. 8 to 11, inclusive, four sectioned manganese bronze castings of this type are shown; each was made differently to illustrate the advantages of gypsum insulators. The casting shown in Fig. 8 was made completely in sand by ordinary foundry technique. No chills or insulators were used, and no insulating powder was used on the top of the riser.

Extensive shrinkage is evident in the heavy section at the lower corner beneath the riser. Freed metal could not reach this heavy section to accommodate shrinkage because the thin section solidified first and prevented further feeding. It is important to note the typical secondary shrinkage of the riser in the central portion.

In the casting shown in Fig. 9 no change was made except to place a small pad of gypsum against the thin section at the end, as shown in the sketch of Fig. 9, to retard solidification in this area and make it possible to feed the heavy corner section. The casting is sound and adequate feeding has been accomplished. In Fig. 10 no insulating pad was used at the light section but the casting was made with insulating sleeves around the riser and with insulating powder on top, as the photograph indicates.

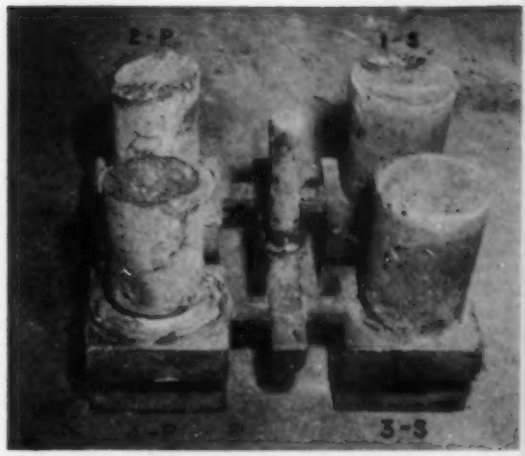
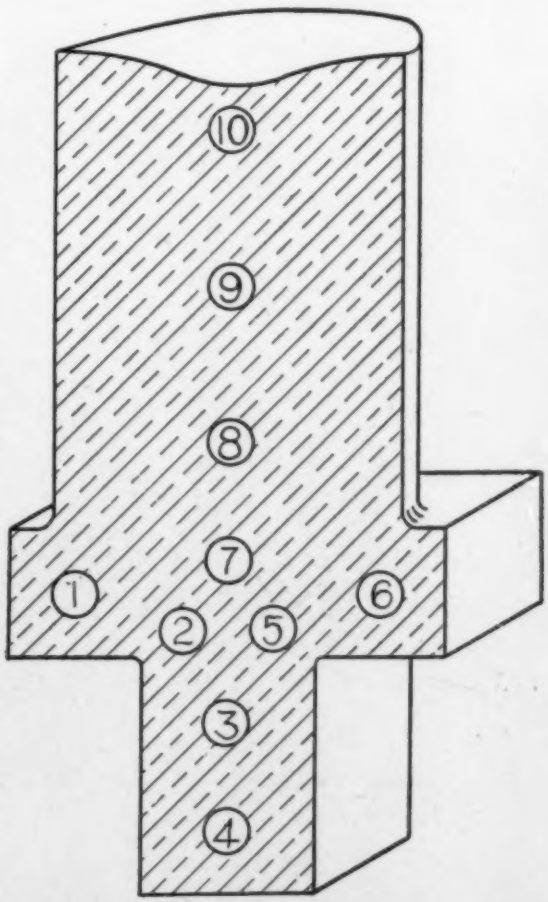


Fig. 19—Four solid tee castings made in the same mold; two made with insulated risers (2-P and 4-P) and two made in sand (1-S and 3-S). "G" bronze; pouring temperature 1150° C. (2100° F.).

PLASTER		
SPECIMEN NO.	TENSILE STRENGTH P.S.I.	PERCENT ELONG. IN 2"
1	41,200	50.7
2	36,400	32.0
3	35,750	33.6
4	37,300	33.6
5	38,100	46.1
6	41,180	48.5
7	41,000	44.5
8	39,100	41.4
9	31,800	25.0
10	* ——	—

*Riser too short

TABLE 1



SAND		
SPECIMEN NO.	TENSILE STRENGTH P.S.I.	PERCENT ELONG. IN 2"
1	33,080	23.4
2	* ——	—
3	28,500	19.5
4	35,220	25.8
5	18,250	5.5
6	32,100	23.4
7	14,400	7.0
8	* ——	—
9	35,400	32.2
10	39,400	51.5

* Broken during machining

TABLE 2

Fig. 20—Sketch of solid tee casting showing location of test specimens and test values; Table 1, insulated riser; Table 2, uninsulated riser.

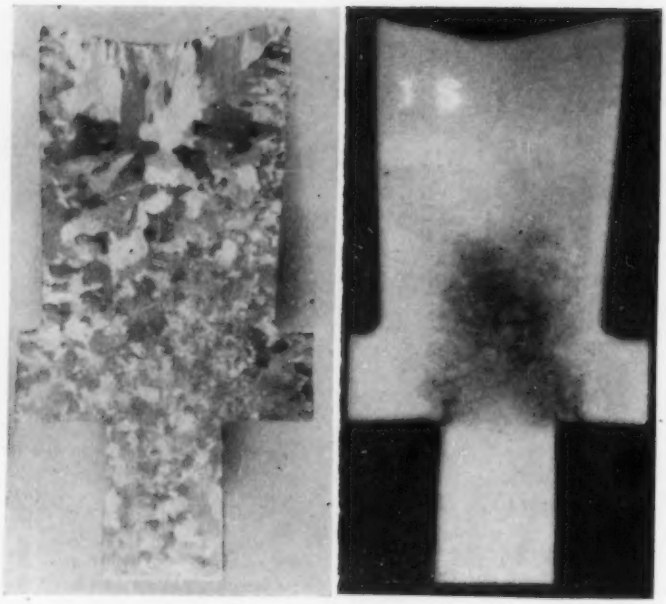


Fig. 21 (left)—Etched section of solid tee casting made in sand with uninsulated riser; broken grain pattern is indicative of shrinkage. "G" bronze; pouring temperature 1150° C. (2100° F.). Fig. 22 (right)—Radiograph of section shown in Fig. 21. Casting unsound.

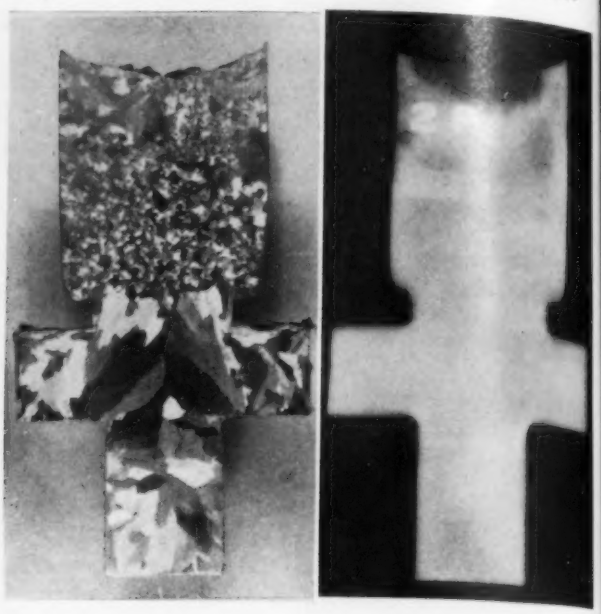


Fig. 23 (left)—Etched section of solid tee casting, made with insulated riser. "G" bronze; pouring temperature 1150° C. (2100° F.). Fig. 24 (right)—Radiograph of section shown in Fig. 23. Casting sound.

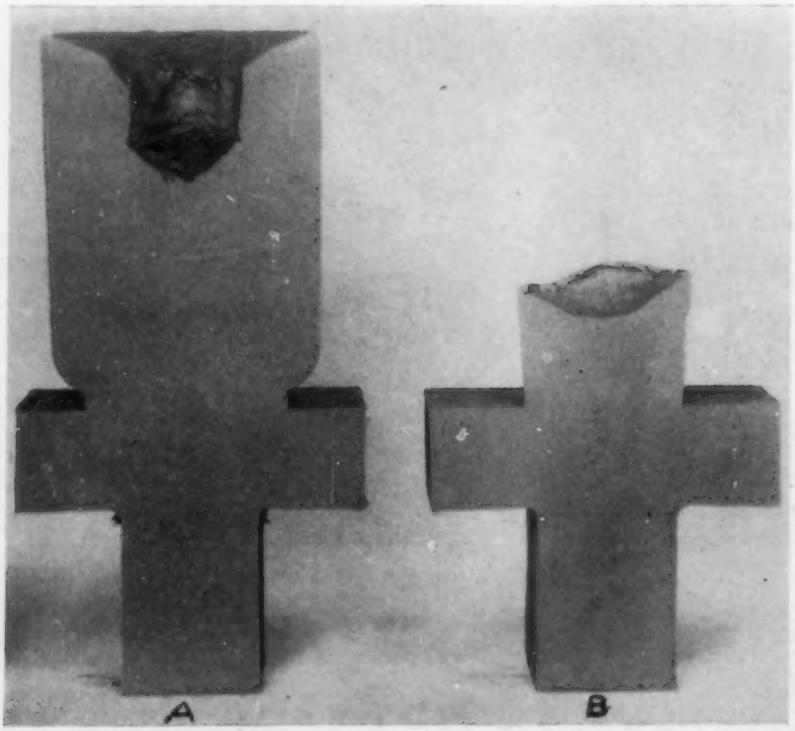


Fig. 25—Solid tee castings made in sand showing relative sizes of riser needed to produce a sound casting with and without insulation. Manganese bronze; pouring temperature 1000° C. (1830° F.). Total weight of A—35 lb., B—21¾ lb.

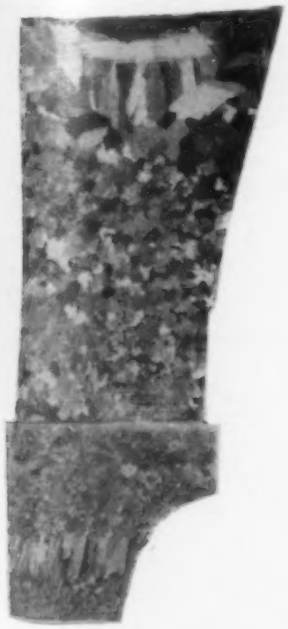


Fig. 26—Etched section of gear blank; the broken grain pattern in the midsection of the casting and riser is indicative of shrinkage; sand cast; uninsulated risers. "G" bronze; poured at 1150° C. (2100° F.)

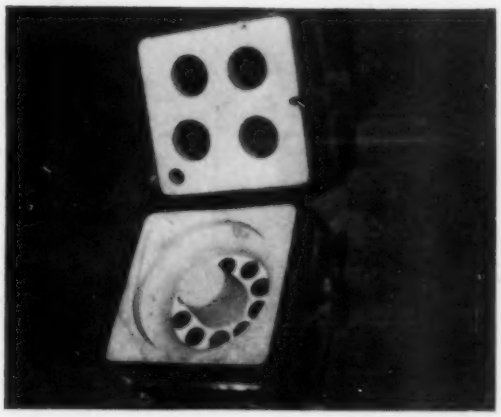


Fig. 27 (left)—Mold for bronze gear blank of Fig. 26 showing chills in drag and risers in cope.



Fig. 28 (right)—Gear blank casting of Fig. 26; total weight 122½ lb., casting weight 50½ lb.

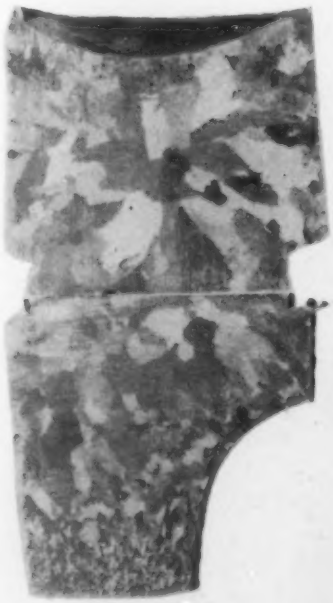


Fig. 29—Etched section of gear blank; the clearly defined grain pattern is indicative of proper directional solidification; sand cast; insulated risers. "G" bronze; pouring temperature 1150° C. (2100° F.).

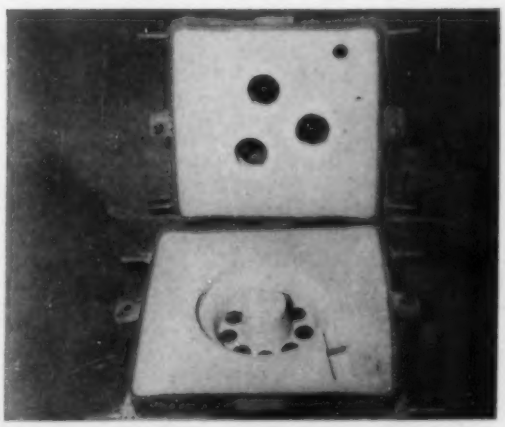


Fig. 30 (right)—Mold for casting of Fig. 29 showing location of chills in drag and insulated riser sleeves in cope.



Fig. 31 (left)—Gear blank casting of Fig. 29; weight of casting 55½ lb., total weight 87 lb.

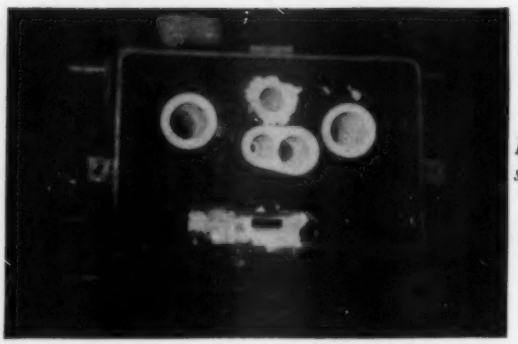


Fig. 33—Cope of mold showing insulated riser sleeves in position.

Fig. 32—Mold for valve casting showing core and location of risers.

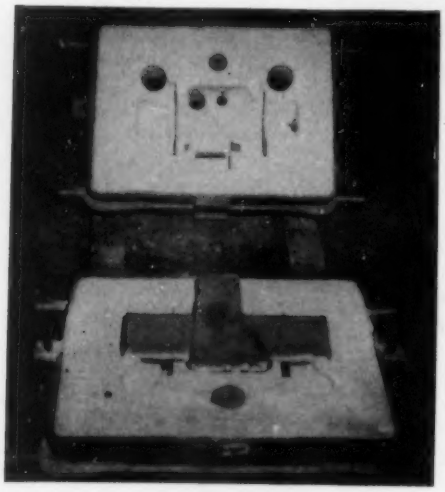


Fig. 34—Valve casting made with insulated risers. "G" bronze; pouring temperature 1150° C. (2100° F.).



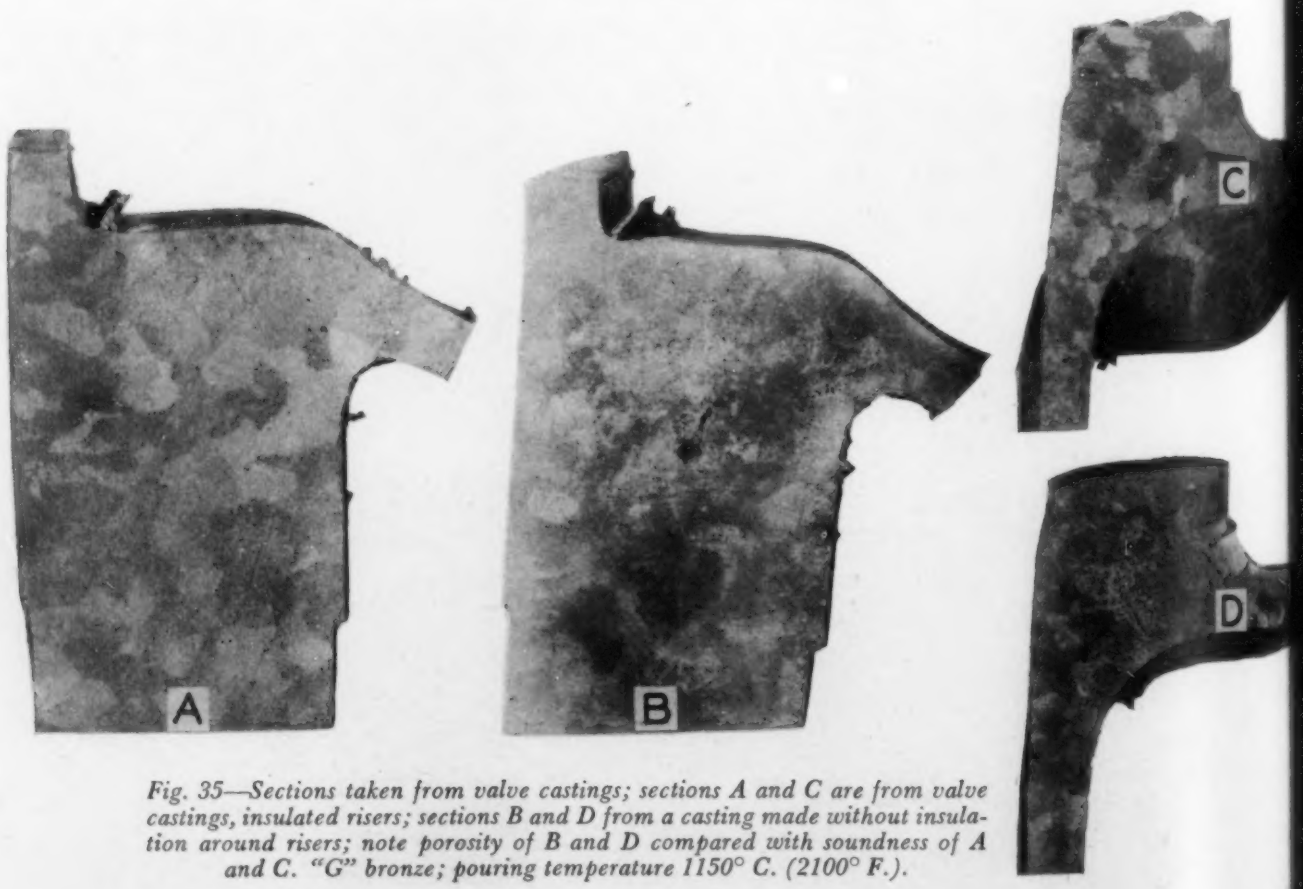


Fig. 35—Sections taken from valve castings; sections A and C are from valve castings, insulated risers; sections B and D from a casting made without insulation around risers; note porosity of B and D compared with soundness of A and C. "G" bronze; pouring temperature 1150° C. (2100° F.).

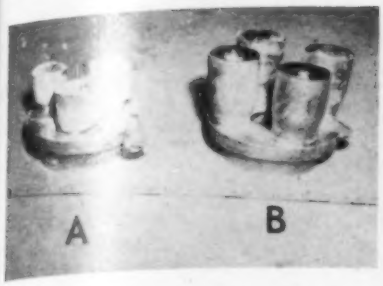


Fig. 36—Manganese bronze gear blanks made with A—insulated risers, and B—uninsulated risers; casting weight—55 lb., total weight of A—75 lb., total weight of B—119 lb.



Fig. 37—Manganese bronze gear blank made with two large uninsulated risers.

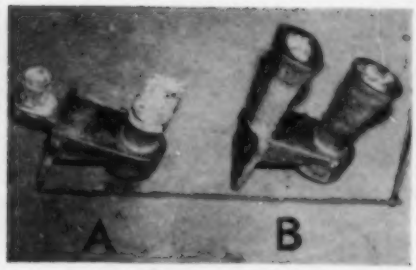


Fig. 38—Comparison of two similar manganese bronze castings made with A—insulated risers, and B—uninsulated risers.



Fig. 39—Manganese bronze ring casting showing small insulated blind risers.



Fig. 40—Globe valve casting; "G" bronze; made with uninsulated risers. Total weight—72 lb.

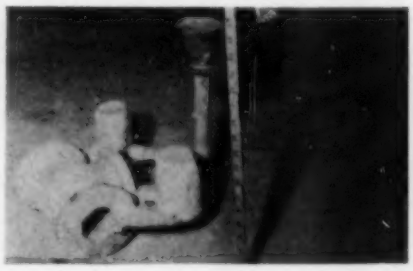


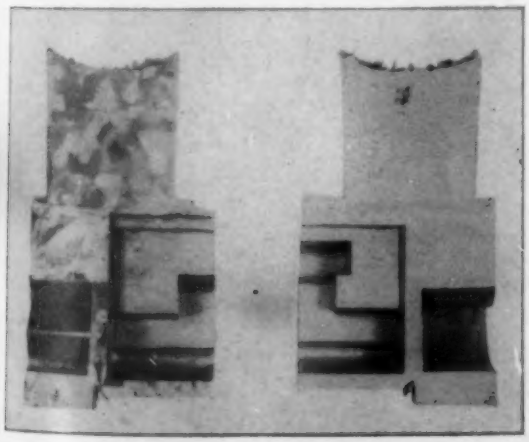
Fig. 41—Globe valve casting; "G" bronze; made with insulated blind risers. Total weight—55 lb.

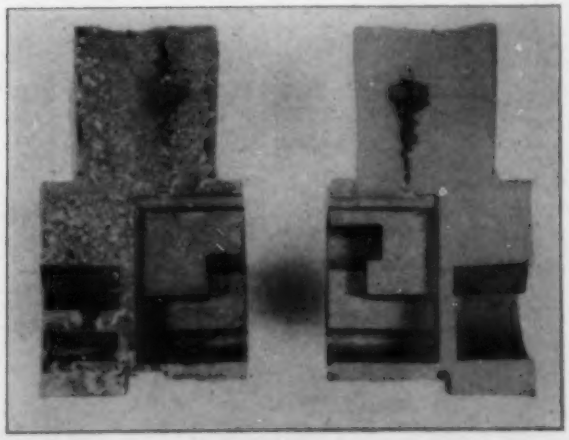
Fig. 42 (right)—Shop casting showing molding method; "G" bronze.



Fig. 43—Sectioned shop casting made with plaster cores and insulated riser in a sand mold and showing complete soundness. "G" bronze; poured at 1150° C. (2100° F.).

Fig. 44—Sectioned shop casting made with plaster cores and uninsulated riser in a sand mold and showing the effect of atmospheric pressure. Note shrinkage cavity and broken grain pattern. "G" bronze; poured at 1150° C. (2100° F.).





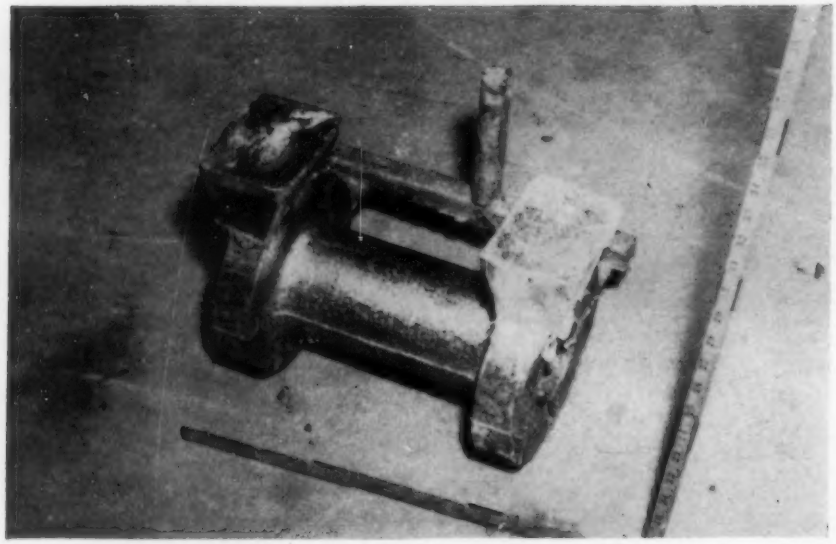


Fig. 45 (left)—Flanged pipe casting made with uninsulated riser on left flange and insulated riser on right flange. Manganese bronze; pouring temperature 1000° C. (1830° F.).

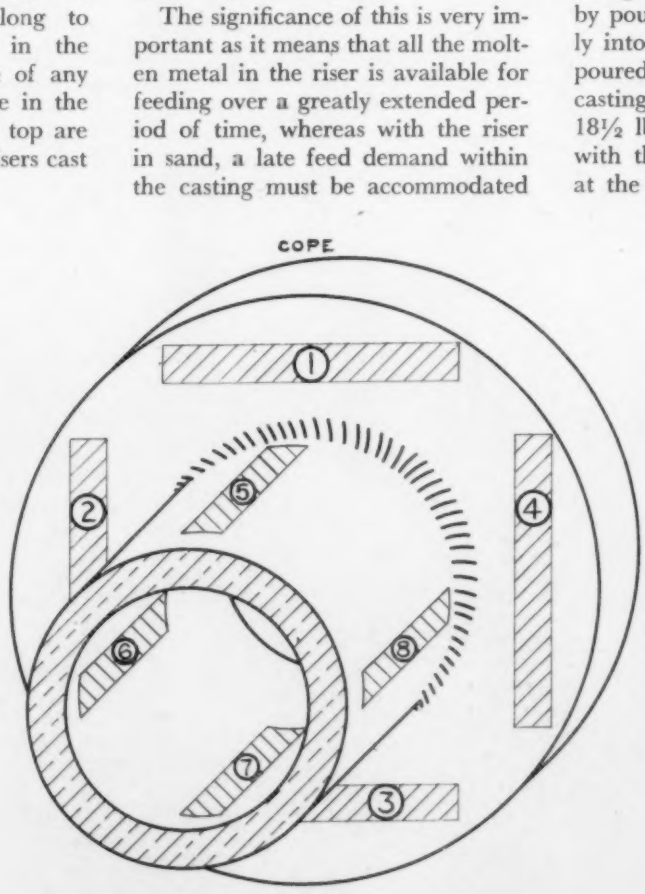
By careful examination a very small pinpoint cavity of shrinkage is evident in the heavy corner section, but the casting is very nearly sound and feeding is almost adequate even though no pad was used on the thin end section. The insulating effect of the sleeves extends far enough to cause the thin section to remain open sufficiently long to nearly feed the shrinkage in the heavy section. The absence of any trace of secondary shrinkage in the riser and the flatness of the top are very important features of risers cast in gypsum insulators.

The risers of Figs. 8, 9, and 10 were filled to the same height originally, but the final riser height of the insulated riser of Fig. 10 is ⅜-in. shorter than the others. This indicates that the metal in insulated risers settles as a molten body and does not freeze along the walls as in the case of sand risers.

The significance of this is very important as it means that all the molten metal in the riser is available for feeding over a greatly extended period of time, whereas with the riser in sand, a late feed demand within the casting must be accommodated by metal draining from the central part of the riser. The results of the thermocouple experiment previously discussed indicate also that the feed metal in sand risers is colder and less effective than in insulated risers.

Figure 11 shows a casting in which the riser size was reduced almost to a minimum by the use of a gypsum sleeve and necked down to facilitate removal from the casting. An insulating pad was used at the thin section. Made in this way the casting is sound, the riser is easily removed, and the amount of riser metal is small. No attempt was made to develop optimum gating conditions. The technique shown is not recommended as best, and any number of gating systems could have been used just as readily.

The castings of Fig. 12 are of manganese bronze and were made by pouring the molten metal directly into the risers. The castings were poured vertically as shown. The casting at the left in Fig. 12 (weight 18½ lb.) was made in a sand mold with the riser in sand and the one at the right (weight 13½ lb.) in a



SAND

SPECIMEN NO.	TENSILE STRENGTH P.S.I.	PERCENT ELONG. IN 2"
1	73,250	32.8
2	72,750	26.8
3	74,150	35.9
4	71,750	24.2
5	67,250	15.7
6	65,500	14.8
7	66,000	14.8
8	73,250	28.1

TABLE 3

PLASTER

SPECIMEN NO.	TENSILE STRENGTH P.S.I.	PERCENT ELONG. IN 2"
1	72,250	27.4
2	74,000	32.8
3	74,200	35.2
4	73,750	34.4
5	63,750	12.5
6	72,200	21.1
7	70,250	20.3
8	74,200	28.1

TABLE 4

Fig. 46—Sketch of flanged pipe casting showing location of test specimens. Table 3—uninsulated riser, Table 4—insulated riser.

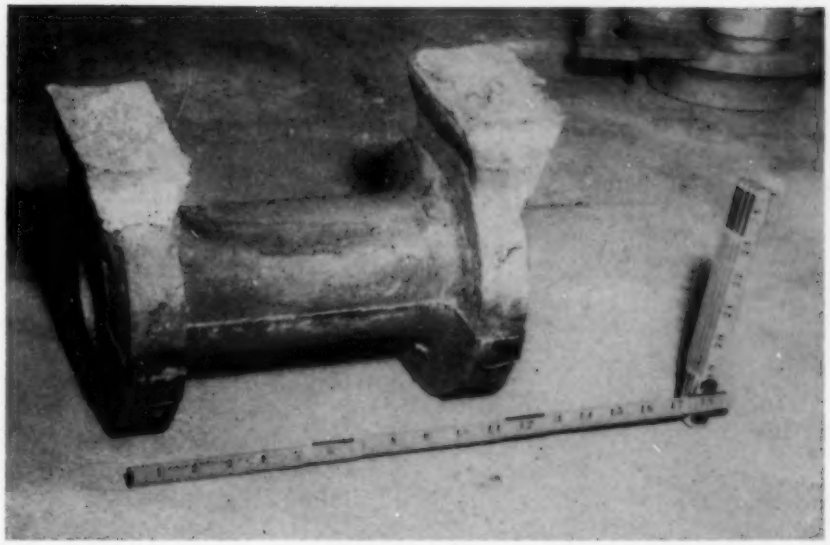


Fig. 47 (right)—Flanged pipe casting made with insulated riser on left flange and uninsulated riser on right flange. "G" bronze; pouring temperature 1150° C. (2100° F.).

sand mold with the riser insulated. Due to the secondary shrinkage evident in the uninsulated riser it could not have been made materially smaller without producing a degree of unsoundness in the casting.

The second test casting chosen was the type shown in Figs. 13 and 14. Because of the compact mass and sharp internal corners such a casting usually is difficult to make in the foundry unless extremely large risers are used. The sketch, Fig. 14, gives the details of the casting, gate and insulated riser.

Figures 15 to 18, inclusive, are photographs of sections of solid tee castings made as indicated. Several very interesting and important metallurgical and foundry principles are evident from these illustrations. For example, the effect of atmospheric pressure is clearly shown in Fig. 15. The riser, being too small and uninsulated, froze prematurely at the neck and could not supply molten metal to feed the shrinkage taking place in the casting.

As a result a partial vacuum was created within the casting, setting up a difference in pressure between the inside and outside of the casting which exceeded the strength of the solidified skin of metal, and the walls were forced inward at the internal corners. The extensive tear is actually caused by this condition, and if the casting were thoroughly sectioned and examined it would be likely that filaments of the tear actually reached through the solid wall and made contact between the inside cavity and the atmosphere.

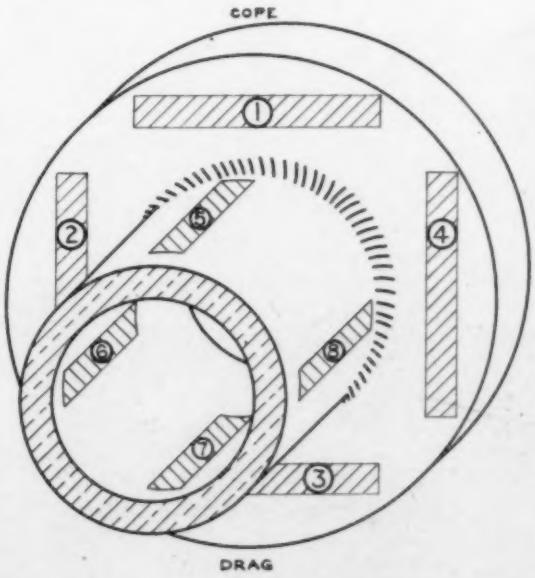
The casting shown in Fig. 16 was made identically with that of Fig. 15 with the exception of a gypsum sleeve around the riser and powdered gypsum on top of the riser. Both castings were made of manganese bronze which was poured at 1000° C. (1830° F.).

Figures 17 and 18 show castings made in the same way as those of Figs. 15 and 16 except that composition "G" bronze was used. The sections were etched in a solution of 1:1 nitric acid and water, in order to shown the typical shrinkage at the internal corners of the casting (Fig. 17) made without insulating the riser. The interdendritic type of shrinkage in tin bronzes is not readily apparent to the eye on polished surfaces unless they are etched, whereas the piping type of shrinkage in manganese bronze is clearly evident.

SAND		
SPECIMEN NO.	TENSILE STRENGTH P.S.I.	PERCENT ELONG. IN 2"
1	* ———	——
2	26,800	15.7
3	18,300	7.8
4	23,100	9.4
5	42,500	40.6
6	41,950	42.2
7	41,200	35.2
8	39,550	31.2

*Broken during machining

TABLE 5



PLASTER		
SPECIMEN NO.	TENSILE STRENGTH P.S.I.	PERCENT ELONG. IN 2"
1	37,700	53.1
2	33,800	23.4
3	36,600	25.8
4	33,100	23.4
5	43,000	48.5
6	42,600	44.5
7	46,100	49.1
8	44,600	52.3

TABLE 6

Fig. 48—Sketch of flanged pipe casting showing location of test specimens. Table 5—uninsulated riser, Table 6—insulated riser.

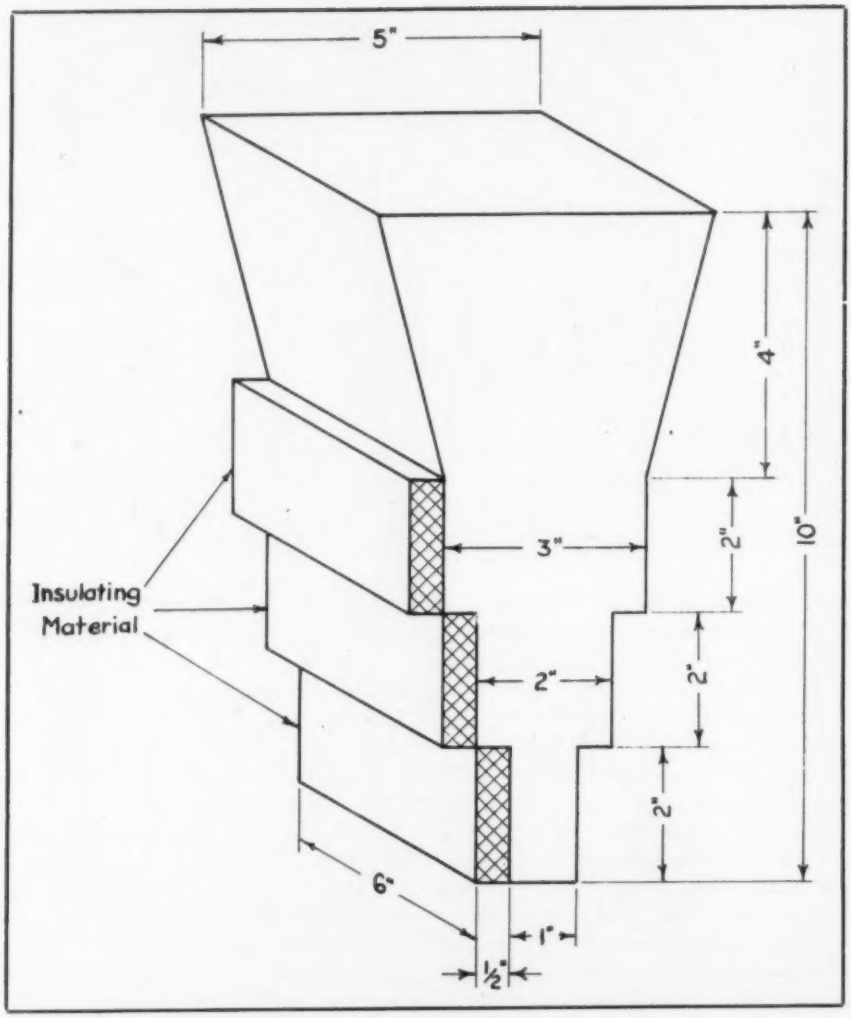


Fig. 49—Sketch of step casting showing dimensions and method of making tests (Figs. 49, 50 and 52).

The broken, undefined grain seen at the internal corners (Fig. 17) depicts shrinkage, and the slight caving in of the corners again shows the effect of atmospheric pressure acting on a low density area. The risers of the two manganese bronze castings (Figs. 15 and 16) originally were poured to the same height.

As in the instance previously discussed, the metal in the insulated riser has settled down as a body and no secondary shrinkage is present. This same phenomenon is shown in a somewhat lesser degree in a comparison of Figs. 17 and 18.

Figure 19 is a photograph of four identical castings made in the same mold to insure as nearly as possible identical pouring conditions. Two of the castings were made with insulated risers and two were not. The metal used to pour these castings was composition "G" bronze (88 Cu, 9 Sn, 3 Zn).

Poured at 1150° C. (2100° F.) one pair of castings was used to study physical properties by cutting blanks for tensile specimens from the positions shown in the sketch in Fig. 20. The results of this comparison are shown in the Tables 1 and 2 on either side of Fig. 20, and the casting made with the insulated riser proved superior in soundness, as shown by the tensile properties.

The other two castings were sectioned and a ½-in. thick slab was removed transversely from the center of each. This thin slab was radiographed and later etched to show relative soundness. The results are shown in Figs. 21 to 24, inclusive, which are self-explanatory. To the casual observer the macrostructure shown in Fig. 21 might not be particularly disturbing because the shrinkage is not evident as extensive voids.

However, the radiograph of Fig. 22 shows the defect in its true light. Adequate feeding is not present. No insulation of any type was used on this riser. The casting of Figs. 23 and 24 is sound and shows no porous areas when etched or radiographed. Again much significance can be given the metallurgical and foundry principles displayed.

The broken, poorly defined grain of the casting shown in Fig. 21 is proved by the radiograph of Fig. 22 to be indicative of low density, whereas the larger, clearly defined, clean grain of Fig. 23 is indicative of sound metal. It is interesting that due to rapid dissipation of heat by radiation, the grain structure in the upper part of the riser in Fig. 21 indicates the metal in this area to be sound with a porous region beneath.

The importance of studying the soundness of castings by x-raying thin slabs removed from critical areas cannot be emphasized too strongly. It is safe to say that an x-ray taken through the entire casting of Fig. 21 would not have shown too clearly the extent of the voids. X-ray pictures taken normal to the main plane of a defect do not show



Fig. 50—Section of step casting showing gas porosity inside casting skin on uninsulated side; indicating slower skin formation rate on insulated side allowing gas to escape. "G" bronze; pouring temperature 1150° C. (2100° F.).



Fig. 51—Section of step casting; right side uninsulated; left side insulated. Brinell hardness survey indicated slight difference in hardness from right to left; etched to show grain pattern. "G" bronze; pouring temperature 1150° C. (2100° F.).

the defect in its more serious aspects.

Measured under such conditions the actual magnitude of the defect may equal only about 3 per cent of the total thickness of the casting, an amount which approaches the lower limit of sensitivity of film. However, a thin section cut transversely across an unsound area presents a maximum percentage of defective metal for the x-ray to reveal and maximum sensitivity is obtained.

It is true that in this sense radiography becomes a destructive rather than a non-destructive test, and the technique could not be used for routine inspection. However, it can be used to great advantage as a research tool and as a means for critically inspecting pilot castings.

The casting of Fig. 25*(A)* weighed 35 lb. as cast and shows the solid tee-block casting of manganese bronze made sound by using the smallest possible riser without the aid of insulation. A smaller riser caused the secondary shrinkage to reach into the casting. Figure 25*(B)* shows the same casting made sound with an insulated riser of greatly reduced size. The casting weighed 21¾ lb.

In the various experiments no attempt was made to determine the effect of riser design upon the relative feeding efficiencies of uninsulated risers. Improvements in the design of uninsulated risers would also improve the behavior of insulated risers, and this should be a subject for independent investigation. The straight-sided riser is certainly not the optimum design.

Several shop castings of a miscellaneous nature made in the foundry during the progress of the experimental work indicated the practical significance of the development. These castings are of sufficient interest for detailed discussion.

The tin bronze gear blank shown in Fig. 28 is of 10-in. diameter and the finished casting weighed approximately 50½ lb. after the risers had been removed. The mold for this casting made with uninsulated risers is shown in Fig. 27. Round chills were placed as shown in the drag. Several attempts to make the casting with fewer and smaller risers failed, and even with the arrangement shown, a slight amount of shrinkage was found when thin



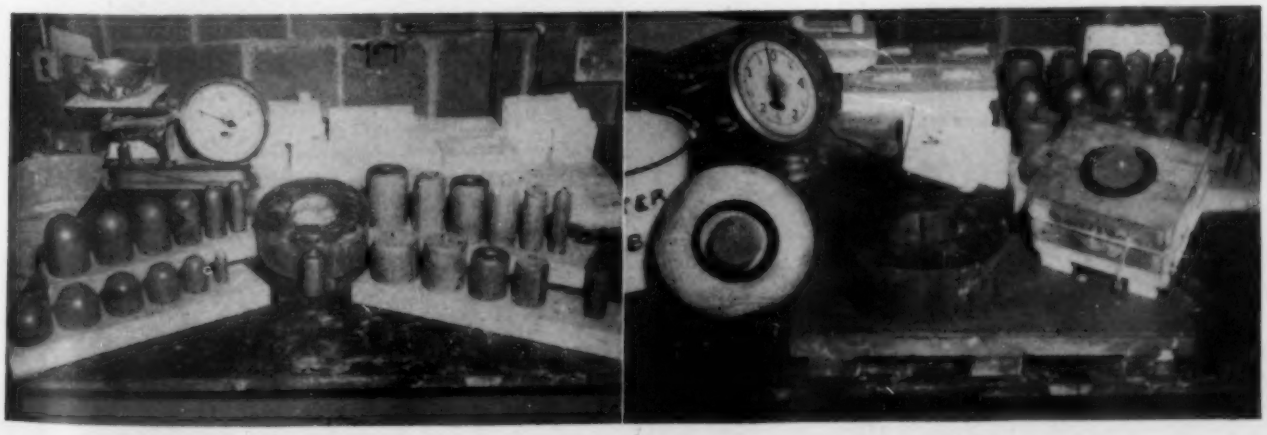
Fig. 52—Section of step casttng made and tested same as in Fig. 49 and showing same relative properties. Manganese bronze; pouring temperature 1000° C. (1830° F.).

transverse slabs were cut from several positions and radiographed.

The macrostructure shown in Fig. 26 is again typical of castings made with uninsulated risers. The fine-grained structure immediately above the chills is caused by a different phenomenon than that which causes the apparently fine grain in imperfectly fed sections, and is sound where the latter is not.

The method used for making this casting with insulated risers is shown

Fig. 53 (left)—Assortment of riser sleeve mold forms used to make insulated riser sleeves. Fig. 54 (right)—Assembled mold forms for making insulated riser sleeves.



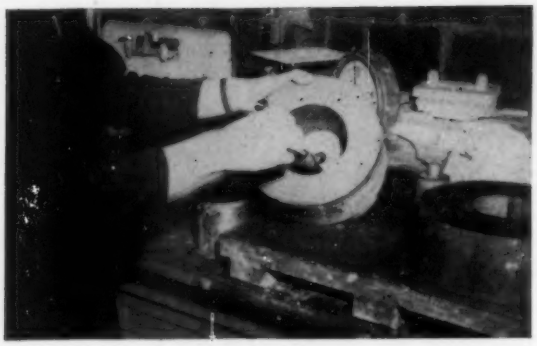


Fig. 55—Brushing separating agent on mold form.



Fig. 56—Weighing water for plaster slurry.



Fig. 57—Weighing gypsum for plaster slurry.



Fig. 58—Sprinkling weighed gypsum onto surface of weighed water.



Fig. 59—Hand mixing gypsum and water to form a smooth, lump-free slurry.



Fig. 60—Pouring slurry into mold forms.

in Figs. 27 and 28. Figure 29 shows the clean, well defined structure with the grains progressively larger toward the riser, indicative of sound tin bronze, and x-ray examination revealed a completely sound casting.

Both castings, however, appeared to be sound when tested non-destructively by x-ray examination of the whole casting. Figure 30 shows the insulated riser sleeves in place and Fig. 31 shows the casting. Three risers were used in this case as compared with four in the former.

A particularly interesting development is shown in Figs. 32 to 35, inclusive. The casting shown in Fig. 34 is a "G" bronze valve which has been extremely troublesome in production at the Norfolk Navy Yard. A cooperative project was arranged to determine if the use of insulated risers would improve the soundness of this critical casting and reduce the number of castings rejected after pressure testing.

The casting was approximately 15 in. long, 8 in. in diameter, and the average wall thickness was 5/16 to 3/8-in. in the body and 3/4-in. on the flanges. The experiment consisted of making two castings simultaneously without changing the original gating or risering system. One casting was made with insulation placed around the riser, as shown in Fig. 33, and powdered insulator placed on the metal in the open riser.



Fig. 61—Striking off excess plaster before the riser sleeve is removed from the form.



Fig. 62—Removing "set" riser sleeve from form.

Fig. 63—An assortment of insulating riser sleeves.



Fig. 64—Placing the green plaster sleeves in oven for drying before use.



The other casting was made with uninsulated risers and no insulating compound placed on the metal in the risers. Photographs of typical sections cut from critical parts of the casting are shown in Fig. 35. The sections designated *A* and *C* represent the casting made with the insulated risers.

This casting was sound when inspected by x-ray and did not leak under routine pressure tests at the Norfolk Navy Yard. Of interest is the special double riser sleeve made to accommodate the two risers of different size which were placed so close together that two individual riser sleeves could not be inserted.

Several typical castings made to indicate the practical advantages of riser insulation are shown in Figs. 36 to 41, inclusive. Figure 36 shows again the previously discussed gear blank casting for purposes of comparison with the technique used in making the casting shown in Fig. 37. These castings shown in Fig. 36 (*A* and *B*) are of manganese bronze.

The casting *A* in Fig. 36 was made with 3 insulated risers of 4-in. diameter and 4 in. high, "necked in" to 3-in. diameter at the casting junction. These risers produced a perfectly sound casting. The casting *B* in Fig. 36 was cast in the same manner except that four uninsulated risers were used, each being of 4-in. diameter and 6 in. high.

These risers were inadequate and, consequently, at the base of each riser atmospheric pressure caused a "dish" or pocket in the casting. The casting shown in Fig. 37 was of the same order of soundness as the one made as shown at *A* of Fig. 36.

Figure 38 shows a shaft hanger casting made (*A*) with and (*B*) without insulating risers, and Figs. 40 and 41 show a similar arrangement with a valve casting. Figure 39 shows the use of insulated blind risers placed above the ring casting. This technique supplies adequate feed metal without the necessity of bringing the riser through to the top of the mold.

Another interesting tin bronze casting is shown in Fig. 42. It was necessary to make the surfaces of the cored parts of the casting sufficiently smooth to provide a sliding fit for another part without benefit of machining or grinding. The surfaces of non-ferrous castings made previously in gypsum were extremely smooth and indicated the possibility of using it as a core material.

The casting shown in section as Fig. 44 was made with gypsum cores. The riser and outside of the casting were molded in ordinary sand. The unetched section of Fig. 44 is particularly interesting as it shows how the riser froze over at the top due to radiation and how atmospheric pressure broke through the skin of the casting beneath the riser to relieve the partial vacuum developed due to shrinkage inside the riser.

The gypsum cores conducted heat less readily than the sand and all surfaces next to the sand froze first. The hot spot was located beneath the riser instead of at the top of the riser where it should have been. Figure 43 shows the casting made with

gypsum cores and insulated riser, the outside of the casting proper being in sand. The riser in this case is excessive. In spite of the difference in grain structure between the castings of Figs. 43 and 44, the composition and pouring temperature of the metal are the same for each.

Flange Fittings

Figures 45 and 47 present photographs of flanged fittings made of manganese bronze and composition "G," and physical properties obtained from specimens cut from the castings in the locations shown in the sketch of Figs. 46 and 48.

The results are self-explanatory and no attempt is made to expand upon their significance as too few such tests were made to be conclusive. However, it is interesting to note that the large difference in the physical properties of the "G" bronze casting between the light pipe section and the heavy flange section is reduced when insulation is used on the risers.

The sketch of Fig. 49 shows how the step castings of Fig. 50, 51, and 52 were made. As shown, one side was insulated and the other was not. The metal used to pour the composition "G" bronze casting of Fig. 50 was not properly deoxidized and was gassy.

It is interesting to note that in spite of this, the section of casting adjacent to the insulation was sound. Evidently solidification was retarded sufficiently to allow the gas to escape. The unsoundness of the castings next to the sand is often found in "G" bronze castings not properly degassified. The "G" and manganese bronze castings of Figs. 51 and 52, respectively, were sectioned, etched, and tested for hardness as indicated.

The hardness was uniform throughout, although one side solidified against a sand surface and the other was insulated. Large risers were used above the largest section in each case and feeding was adequate. The hardnesses are approximately the same as for sand castings.

Conclusions

The information presented in this report permits the following conclusions to be drawn:

a. Controlled directional solidification is essential to the manufacture of sound non-ferrous castings. This can best be obtained by careful attention to gating and risering practices.

b. The interdendritic shrinkage of "G" bronze and the gross shrinkage of manganese bronze have a similar origin—lack of sufficient feeding—which can be corrected by proper control over the temperature gradients within the solidifying casting.

c. Atmospheric pressure operates to the advantage or disadvantage of casting soundness in non-ferrous metals to fully as great an extent as it does in steel, and the principles of operation are identical. Whether good or poor castings result depends upon how well this important foundry tool is understood and applied. Theories and techniques have already been worked out by steel foundrymen and should be carefully studied by non-ferrous foundrymen interested in improving their practices. Although fundamental principles apply similarly to all metals, techniques vary somewhat.

d. The x-ray is a powerful tool in revealing unsoundness in non-ferrous castings which is not readily apparent by other testing methods. X-ray studies of thin, transverse sections cut from pilot or test castings will do much to improve gating and risering practices if careful attention is paid to the radiographs.

e. Low density areas in non-ferrous castings are revealed on polished and etched cross sections by a broken, poorly defined grain pattern, usually surrounded by an area of larger clear-cut grains. The shrinkage then can be verified by x-rays of thin, transverse sections cut from the area.

f. A study of transverse x-rays can be used to indicate the reason for leakage in non-ferrous pressure castings. Improperly fed castings, particularly in "G" bronze, usually will show an extensive low density area surrounded by a relatively small amount of sound metal. The outside sound portions often are removed by machining and leakage takes place through the filamentary shrinkage voids.

g. It is possible that the major portion of rejects of non-ferrous castings are a direct result of inadequate feeding rather than the many other variables often blamed.

h. The feeding range and efficiency of risers when designed and used according to usual foundry practices are limited and considerably less effective than might be expected. These limitations exist because the molding sand around the risers is a relatively good conductor of heat and because of loss of heat by radiation to the air if open risers are not covered with some form of insulation.

i. It has been found possible to greatly increase the over-all efficiency of risers by insulating them and by using insulating pads on thin sections when it is necessary to feed through thin sections to reach heavier sections beyond. Gypsum, when used in the manner described in the Appendix, has been found to be an efficient insulator, economical and easy to fabricate into any desired shape. Gypsum powder is also used as an insulator by sprinkling it over the top of open risers.

Advantages of importance to foundrymen which result from the use of insulating pads and riser sleeves are as follows:

A. Sounder castings. (1) This is the result of the metal in the riser remaining molten until long after the casting has solidified. (2) By use of the pads on thin sections the feeding range of risers is increased.

B. Increased yield. (1) The riser volume can be reduced by as much as 75 per cent in certain cases. The contact between riser and casting can be materially reduced. (2) This results in a markedly greater number of castings per ladle of metal.

C. Reduced cleaning cost. (1) Removal of risers and materials handling are simplified because smaller and often fewer risers are required. Also, the contact between riser and casting can be reduced to the point where the risers may be removed with a hammer blow.

Disadvantages are:

A. Cost of preparing and using the insulators. (1) This is not great as the materials are cheap and easy to handle. (2) Nothing is more expensive than an unsound casting.

B. Possibility of contamination of sand. (1) This can be prevented to a large extent by removing the powder and sleeves prior to the shake-out, or by making the prefabricated pieces strong enough to resist breaking up during the shake-out, so that

they may be removed later. (2) The use of a new sand facing would eliminate the problem. (3) It is not expected that this would prove serious, but it is a problem which would have to be solved individually.

Acknowledgments

The authors wish to express appreciation to the various members of the Foundry and Research Staff for their interest and assistance, particularly to George Schmidt for the sketches which serve to clarify certain of the illustrations.

Appendix

The material used in making the insulating pads and riser sleeves discussed in this report was gypsum. A great variety of other materials suitable for use is available and several were tried, but none was superior to gypsum in insulating quality, general ease of fabrication, and cost. Since our interest is mainly in developing the theory and practical technology of a given foundry process or procedure, no attempt was made to test all possible materials or to develop new materials, particularly when gypsum proved so satisfactory.

Several different grades of gypsum are available for special purposes, and the type marketed for use in metal casting was used for the work described in this report. This particular grade has been developed specifically for use as molding material in making non-ferrous castings of superior surface appearance and to close tolerances. It is essentially plaster of paris (calcined gypsum) to which has been added 20 per cent of fibrous material.

The procedure for making riser sleeves and insulating pads with gypsum products is illustrated in Figs. 53 to 64, inclusive. Here are described various steps in the order of their occurrence, although variations may be necessary or beneficial to the practice of individual shops.

Method of Mixing

Figure 53 shows a number of wooden plugs and a wooden mold in which any one of the plugs may be placed to make either blind or open risers. The plugs on the left are used to form blind risers and those shown on the right make open risers, some of which are used as "necked-down" riser sleeves.

Assembled mold forms are shown in Fig. 54. The mold forms must be coated with a separating agent before the plaster slurry is poured into them, and this is illustrated in Fig. 55. The separating agent is a water-repellant coating, and usually is brushed on the mold but may in some cases be sprayed.

Figures 56 and 57 show water and gypsum being weighed to the proper proportions to give what is termed a 160 consistency. It will be found after little experience that the best and most practical way to mix the slurry is to sprinkle the gypsum on top of the water, as shown in Fig. 58, and allow it to settle or soak for approximately 2 min. before hand mixing.

Figure 59 shows a batch of slurry being hand mixed until it is of a lump-free, creamy consistency. After the slurry starts to cream or thicken it is poured into the riser sleeve form, or into a block form for use as insulating pads (Fig. 60). The plaster slurry becomes a solid mass approximately 15 min. after it is mixed, and any excess, as shown in Fig. 61, can be struck off even with the top of the form with a piece of flat metal or wood.

If the mold is properly coated with a separating agent, the "set" plaster sleeve will be easily removed from the form by a few gentle raps to break a vacuum which tends to form between plaster and mold surfaces. This is shown in Fig. 62. Various sizes and shapes of open riser sleeves are shown in Fig. 63, and Fig. 64 shows a number of green slabs of plaster as well as several riser sleeves ready to be dried in an oven before being used in the production of castings.

Plaster Consistency

Some explanation may be in order at this point to clarify the meaning of the term consistency. The word consistency as it is used in this report means the plaster to water ratio used in making insulating sleeves and pads. A 160 consistency which is used for all the sleeves and pads in this report means 160 lb. of water to 100 lb. of metal casting plaster. When the plaster to water ratio is maintained at or close to the 160 consistency, little if any cracking of the sleeves will occur during the operation of oven drying the sleeves.

If the dried sleeves are too soft and fragile for practical use, it may be desirable to make an addition of portland cement to the dry gypsum before it is added to the water. It has been found that 20 per cent of portland cement by weight added to the dry gypsum will substantially increase the mechanical strength of the dried sleeve with little or no effect on its insulating quality.

Drying Methods

Sleeves and insulating pads made of gypsum products must be dried before they are used in molding operations. The temperature that has been found to be satisfactory is from 400 to 1200° F., and the time necessary to dry a given sleeve or pad will depend upon the drying temperature.

If the higher temperature range is utilized for faster drying, a certain degree of cracking can be expected if the sleeves are removed from the furnace and air cooled to room temperature; however, slight cracks have no noticeable effect on the insulating qualities of the sleeves but do weaken them somewhat mechanically.

Several satisfactory parting or separating agents have been successfully used, such as stearic acid, petroleum jelly, light lubricating oil, lard oil, etc. An emulsion of granulated soap, water, and light lubricating oil, 40 per cent, 30 per cent, and 30 per cent, respectively, by volume also works well as a mold form coating.

Several variations of plaster of paris have been tested and found to work very well as insulating materials for riser sleeves and pads if a few precautions are observed. In using 100 per cent plaster of paris it is necessary to keep the slurry in suspension until creaming or thickening has started, to prevent a settling action from occurring which will produce a dense, hard structure on the bottom of a riser sleeve and a porous, weak structure at the top.

An accelerator can be used to speed up the "set" of the plaster, thereby decreasing the time of "set." Approximately 1½ per cent of terra alba is a good proportion of accelerator to dry material.

Various materials, such as short-fibered asbestos, vermiculite (expanded mica), moldene or its equiv-

alent, diatomaceous earth, and other such materials which will absorb water can be used satisfactorily with plaster of paris to insure uniform texture in the finished sleeves.

The metal compositions used for the tests illustrated in this report were as follows:

Manganese Bronze, per cent		*Composition "G" Bronze, per cent*	
58.0	Copper	88	Copper
38.5	Zinc	9	Tin
1.0	Iron	3	Zinc
1.0	Aluminum		
0.75	Manganese		
0.75	Tin		

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DISCUSSION

Chairman: G. K. DREHER, Rogers Pattern & Foundry Co., Los Angeles.

DR. B. M. LORING[1] *(written discussion):* The insulation of risers probably will be much more widely used for many alloys other than bronzes when improvements are made in the technique of insulation. Materials having greater resistance to high temperature and greater ease of fabrication into sleeves will have to be developed.

It might be mentioned that in the 1943-44 Proceedings of the Institute of British Foundrymen, Vol. 37, page B26, the use of asbestos paper for riser sleeves of magnesium castings is advocated. The insulation of risers by this means according to the I.B.F. Proceedings, was originated in Germany several years before. The asbestos paper is easily cut to size and may be stapled together in cylindrical shapes.

For manganese bronze, three or four thicknesses of the asbestos paper 1/32 in. thick are probably necessary for small or medium sized risers. The cost of the paper per sq. yd. is not excessive, and several risers of average size may be insulated with this amount of material. Asbestos paper has been in use for some time as an insulating material in some magnesium foundries in the United States.

There appear to be few reasons why it cannot be used for copper base castings of pouring temperatures like manganese bronze, provided a simple protective treatment of the paper is employed. This is only one example of another insulating material which might have wider application.

G. E. DALBEY[2]: These gypsum liners are fragile and the molder does not like to handle chinaware. They also have a tendency to pick up moisture when placed in the mold and allowed to stand for some time prior to pouring the molten metal into the mold, do they not? In spite of these difficulties it is economical to use the insulated riser sleeve in many cases.

MR. WICK: They do tend to pick up moisture on standing.

MR. DALBEY: It would be possible to coat them with a water repellent. Would stearic acid help?

MR. WICK: Whether the stearic acid would do more harm than the moisture pickup I do not know. We found the moisture risers pick up from the mold does not affect the casting. We did notice the boiling action which seems to help feeding rather than to hinder it. The boiling action seems to help the metal to feed downward and form a more solid casting, especially in G-bronze.

As far as the breakage of these sleeves is concerned, they can be strengthened so they can be rammed up into the mold with an air rammer. That is, you can not hit the risers with an air rammer, but you can strengthen them with 20 per cent Portland cement and 40 per cent short fiber asbestos.

H. F. TAYLOR[3]: The general principle presented here is not new. We do not mean to imply that we have anything radically novel but to emphasize the way in which an old principle can be used to make sound castings. We are particularly interested not in selling gypsum, but in getting across a theory of solidification.

There are a great many insulators that can be used, such as shredded asbestos or the insulating material that is used to cover hot water pipes. There are probably 20 or 30 such different materials. At the Dodge Chicago plant in casting magnesium they have been using asbestos paper for as long in this country as it has been used in England. So there are a great number of ways in which you could insulate a riser.

Dr. Loring made the statement that simpler methods of insulation would have to be developed. Actually, manufacturing these sleeves is not troublesome. It is a simple thing to do. This technique was used at the Navy Yard at Norfolk and also at Cramp Brass. The troubles of preparing these sleeves are not going to be a hindrance to the adoption of the principle. What you use to do the job with is unimportant. The theory behind the practice is important.

MEMBER: We have used insulated risers for a number of years. I was interested in Dr. Loring's statement that the simple protective treatment of the asbestos paper would solve many problems. I am interested in knowing what that simple protective measure is?

DR. LORING: Perhaps if you will wait for the A.F.A. paper next year, we will be ready to tell you at that time.

CHAIRMAN DREHER: I believe the implications of the theory are great enough to interest many foundrymen. The saving of feeding metal and the more efficient use of the feeding metal that you have left is a problem that can not be passed off lightly. It costs money to melt metal. Cost varies from 0.008 cents per lb. up to 3 cents per lb. If you can save 10 lb. in a riser, you have from 10 to 30 cents to pay for the extra work that this plaster involves. That is only a mild case. There are more extreme cases than that in some of your research work.

How large a casting have you poured in your research work? Perhaps Mr. Brinson can supplement your work with some statement as to the size of casting he has made.

S. W. BRINSON[4]: We have made castings up to about 8 in. in size.

CHAIRMAN DREHER: You have made risers up to 8 in. in diameter in plaster sleeves?

E. D. BOYLE[5]: We have tried this process the authors worked out, but we encountered considerable difficulty in contamination and I do not think that anyone making castings commercially can handle this procedure correctly. At Puget Sound we have taken a blind riser and fed molten steel into the blind riser. We use a type of plaster Paris for riser insulation and go down about a half inch into the top of the riser and have had satisfactory results. We use it in pouring aluminum and bronze castings and have even cut down the shrinkage in the castings.

MR. TAYLOR: I would like to remind Mr. Boyle that it is being done economically and efficiently.

MR. WICK: I would like to ask Mr. Boyle if he could give us a definite case of what trouble the contamination has caused in his castings? Something that he could definitely trace back to the contamination.

MR. BOYLE: I am preparing a technical paper on that point at this time and will shortly submit it for publication.

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ATMOSPHERIC AND WASHBURN RISERS

Proper selection and use of risers results in sound castings and effects operating economies, particularly in the casting cleaning room. Presented at a Steel Session, Fiftieth Annual Meeting of American Foundrymen's Association, May 10, 1946, at Cleveland.

W. E. Jones
General Foundry Superintendent
Ordnance Steel Foundry Co.
Bettendorf, Iowa

RISERS, atmospheric and Washburn, have been the subject of considerable interest in the steel foundry with which the author is associated. It is felt, however, that these risers, particularly the atmospheric riser, may still be largely in the experimental stage. Most of the work carried on in this plant with the atmospheric riser has also utilized a graphite rod.

Whether this method represents a truly atmospheric riser or not may be open to some question. In addition, some considerable experimentation has been conducted utilizing the Williams core, and the results in general have been excellent.

In the case of the atmospheric riser, when the metal is poured into the mold and fills the blind riser, it begins to freeze from the outside surfaces in contact with the mold toward the center of the riser. Should the freezing rate be too rapid, little feeding to the casting will be realized unless the riser is very large or some other agent is utilized to prevent the riser from freezing. With a Williams core placed in the proper location in the riser, the metal within the riser is subject to atmospheric pressure, thereby eliminating the establishment of a partial vacuum within the riser as the metal begins to solidify, the core acting as a passageway for the ingress of air.

The blind riser with Williams core inserted is diagrammatically shown in a section through a mold in Fig. 1.

Thus, with atmospheric pressure within the riser rather than a partial vacuum, the molten metal is free to flow and feed the casting.

Several advantages are obtainable in using a Williams core. This type of core can be made in a "gang" core box, which may either be blown or rammed, and therefore results in a low cost operation. With some care in design, a suitable core print may be developed which will be applicable to a number of similar molding jobs. This again results in further economies. Because of its simplicity, the Williams core may be inserted at the time that the mold is being rammed, thereby saving the time of setting and nailing the cores when the mold is being finished.

Use of Graphite Rods

Some considerable experience has also shown that the utilization of graphite rods in blind risers is advantageous. When a graphite rod of the correct size is properly located across the diameter of the riser, the pipe or shrink cavity within the riser will be shallow and well rounded rather than long and taper-

Fig. 1—Diagram of mold section illustrating use of blind riser with Williams core.

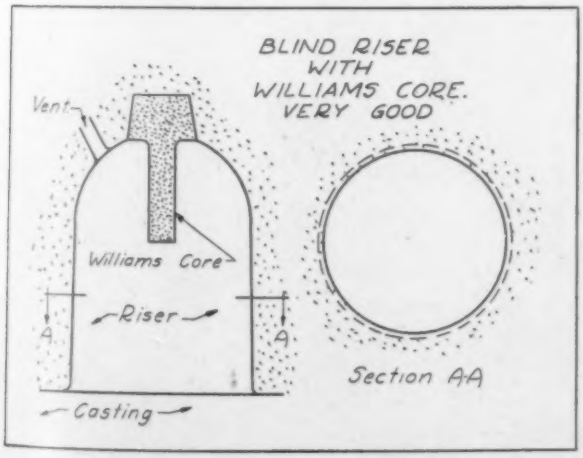
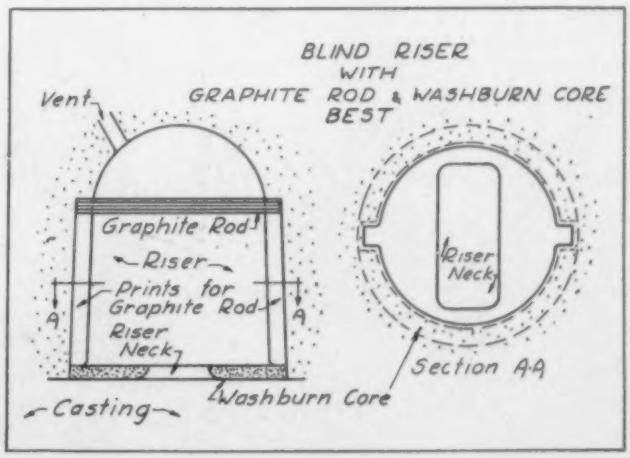


Fig. 2—Sketch showing use of blind riser with graphite rod and Washburn core.



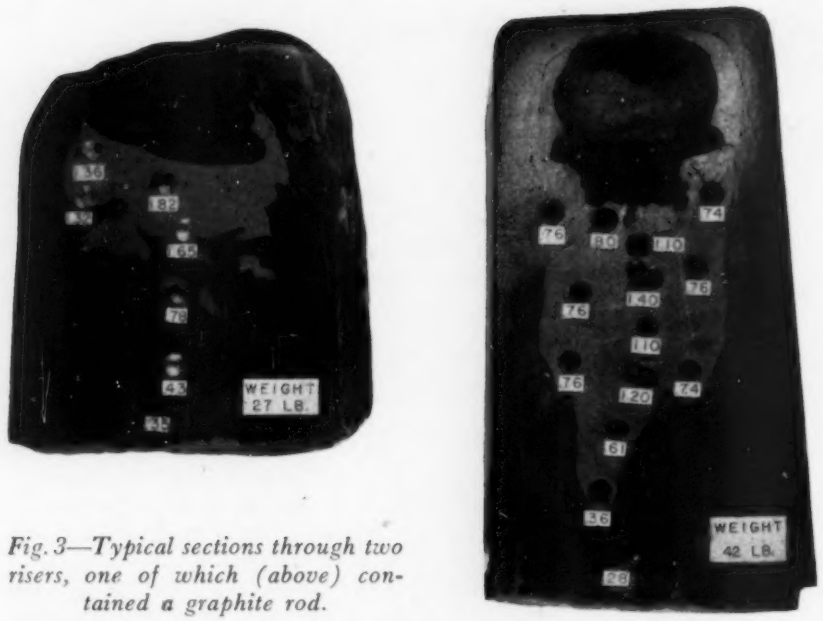


Fig. 3—Typical sections through two risers, one of which (above) contained a graphite rod.

ing. (The location of the graphite rod is diagrammatically shown in Fig. 2.)

A long, tapering, shrinkage cavity within the riser may extend to the casting proper or, to a lesser degree, may cause porosity in the casting adjacent to the riser. In other words, with the use of the graphite rod, the shrinkage cavity will be shaped like a carrot, which obviously is not desirable for reasons already enumerated. By changing the shape and location of the shrinkage cavity within the riser, it is possible to reduce the size of the riser and yet obtain a sound casting.

Figure 3 shows a typical section through two risers, one of which contained a graphite rod. Attention is particularly directed to the amount of metal saved in the reduced size of the riser and the carbon pickup by the steel from the graphite rod, as indicated by the numerical values shown on the photographs.

This change in the character of the shrinkage cavity is quite likely due to the metal having remained in the molten state for a longer period of time, permitting more efficient feeding of the casting. The molten steel in the riser has dissolved carbon from the graphite rod and thereby lowered its freezing temperature, maintaining fluidity for a greater period of time for the feeding of the casting. It is also possible that the hot graphite rod may burn to some extent and liberate a small amount of heat to further delay solidification.

Since graphite has a relatively high thermal conductivity and heat capacity, the diameter or size of the rod and its location in the riser is of considerable importance. A rod which is too large in diameter will cause excessive carbon pickup by the steel and result in a high carbon area within the casting.

Secondly, for reasons of economy, the graphite rod should not be larger than necessary. If the graphite rod is placed too close to the casting, the steel in its immediate proximity, always higher in carbon, may cause an objectionably high carbon concentration in the casting itself. A graphite rod which is too small, on the contrary, will not achieve the desired results, and instead of yielding a shallow shrinkage cavity it will tend toward an elongated or carrot-shaped type.

The diameter of the graphite rod is largely one of selection based upon actual experience for each particular molding job, and quite naturally will vary with the volume of metal within the riser. The manufacturers of graphite rods recommend the following sizes:

Recommended Graphite Rod Sizes

Riser Diameter, in.	*Graphite Rod Diameter, in.*
to 3	3/16
3-5	1/4
5-8	5/16
8-12	3/8
12-15	1/2
15-18	5/8
18-20	3/4

The location of a graphite rod relative to the casting is of considerable importance. If too close, it will cause a high carbon area within the casting, and if not close enough it will be ineffective in reducing the length or height of the riser. The graphite producers recommend that the rod be located a minimum distance from the top of the riser equal to approximately one-half the riser diameter. Experience in the author's plant has shown that the best results are obtained when it is located one-third of the height of the riser as measured from the top of riser.

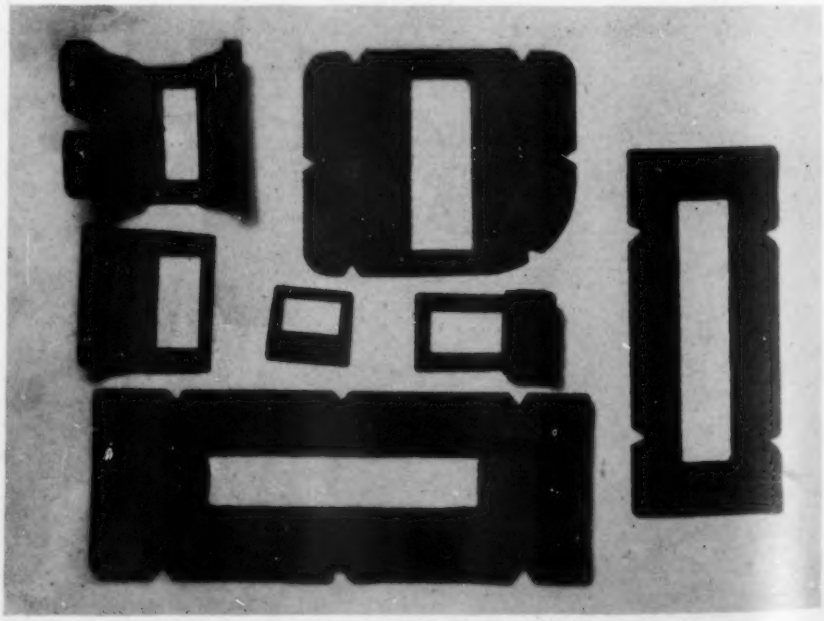


Fig. 4—Several sizes and types of Washburn cores.

The use of the Washburn riser has been so successful in this plant that it is now standard practice to use it wherever possible. It has paid dividends in saleable castings and decreased production costs.

The Purpose of the Washburn Riser

The basic principle of a Washburn riser involves the utilization of a relatively thin core which constricts the riser at the point of attachment to the casting, thereby minimizing the cost of burning off a large riser. The advantages of a heavy riser are retained, since the thin core heats quickly to a relatively high temperature so as to insure feeding of the casting.

The Washburn riser core also has additional advantages in the molding room. It is easily set into the core print, nailed securely, and the job is accurately and uniformly performed. By comparison, the standard riser must be rammed hard, secured or nailed, rounded off with a molder's slick or trowel, and finally scab-nailed. Rarely will two molders (or, in fact, even the same molder) make the same riser identical on two successive molds.

Making a Washburn Riser

Several steps are involved in making a Washburn riser. An attempt will be made to describe briefly the methods utilized in the author's foundry.

First, the location of the riser on the pattern is determined. A frame of wood or modeling clay is then placed around the spot chosen for the riser, this frame being at least 1 in. larger than the size of the riser. The frame is then filled with plaster and given the exact contour of the pattern. When set, the plaster cast is removed and trimmed to proper shape so as to provide for a draft, size and thickness, which should be held to a minimum at the point where the riser joins the casting. The minimum thickness or neck of the riser is dependent on the size and shape of the riser and should be approximately ¼ in. to ½ in.

The plaster print is now placed on the pattern in its proper location, and a plaster frame about 1¼ in. wide poured around the outside contour of the print and struck-off level with the top of the print. Both pieces are then removed, turned face down and a plaster core box



Fig. 5—Example of conventional type of riser. Note large heads which have been burned off.



Fig. 6—Same casting shown in Fig. 5 except made with Washburn risers which have been knocked off.

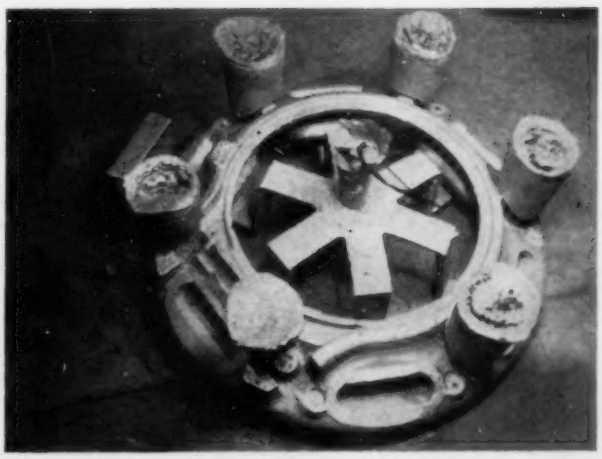


Fig. 7—Casting made with Washburn risers which can be knocked off.



Fig. 8—Another illustration of the advantages of the Washburn riser.

bottom poured over both, utilizing some sort of frame to serve as the core box. The outside frames having been trimmed to size, the riser is placed in position on the bottom core box pattern.

The three pieces described are then sent to the aluminum foundry so that castings may be made. After the castings are filed smooth, the core print is fastened to the pattern with screws. The outside portion is dowelled to the bottom part of the box with four pins which extend through the frame to the top of the box, thereby serving to carry the core print when it is turned over.

The method outlined for making these core boxes is only one of a number of methods available where the position of the core will not afford back draft in the core box. Frequently, simple gang dump boxes can be used. (Fig. 4 shows a number of typical Washburn riser cores.)

Washburn Core Mixture

It is necessary to have good hard cores, otherwise the steel will penetrate or burn into the core, making the risers difficult to knock off and, if large, much more difficult to burn off. The particular core mixture utilized in the author's foundry is as follows:

Crude Ottawa Silica Sand, lb.....	2,000
Silica Flour, lb.....	200
Cereal Binder, gal.....	6
Core Oil, gal.....	4
Moisture, per cent.....	4.4–4.8
Permeability	75
Green Compressive Strength, psi...	3.3

If there is a tendency for steel to penetrate or burn into the core, an addition of 100 lb. of silica flour is added to the above core mixture, making a total of 300 lb.

Burning, Chipping, and Cleaning

On most castings where Washburn risers have been used, the risers may be knocked off easily with about three blows of a heavy sledge-hammer. This eliminates the necessity for cleaning or chipping away the sand that has burned onto or has adhered to the casting before cutting off the riser with the oxy-acetylene torch, since the latter will not burn through adhering sand. Burning off risers with an oxy-acetylene torch is in itself costly and particularly difficult where the riser is placed on a curved surface.

After burning off the risers, particularly large risers, the casting generally is covered with metal and slag, necessitating considerable labor in removal. Relative advantages of the Washburn riser versus the conventional type riser are clearly portrayed in Figs. 5, 6, 7 and 8.

A casting on which a Washburn riser is used arrives in the cleaning room requiring only snag grinding and is finished insofar as risers are concerned. In any case, a minimum amount of burning will be required prior to grinding. Thus, when compared with usual practice where the ordinary riser is used, the burner either burns too low and gouges the casting proper (requiring costly welding repairs) or else, to avoid this, leaves the riser too high, requiring costly grinding operations.

Based upon experience in the author's foundry, it would appear that the Washburn riser will reduce labor in the cleaning room, a constant bottleneck in any steel foundry, and, in addition, will result in decreased production costs.

DISCUSSION

Chairman: C. H. LORIG, Battelle Memorial Institute, Columbus, Ohio.

Co-Chairman: J. A. RASSENFOSS, American Steel Foundries, East Chicago, Ind.

P. G. DE HUFF, JR.[1] *(written discussion):* Mr. Jones, in a very interesting paper, has pointed out the advantages to be obtained by the use of a Washburn core and by judicious application of the blind riser. However, there are certain questions which occurred to me and which I believe merit further discussion.

The first one is, how large a riser can be used with the Washburn core and at what point, that is, at what diameter of riser, will it be necessary to remove the riser by burning instead of by knocking it off with a sledge.

Secondly, it would be very interesting to know the carbon content of the casting proper from which the risers in Fig. 3 have been taken. It certainly appears from Fig. 3 that the one containing the graphite rod would most likely have raised the carbon content of the casting at the point of contact.

A third point in question is in the use of a graphite core extending through the blind riser. I believe that H. D. Phillips has experimented with the use of an exothermic compound such as aluminum and iron oxide. This compound was placed on the graphite rod intended to localize the heat at that point.

Has the author had any experience in the use of such compounds?

It is not quite clear to me just how the thickness of the Washburn core is related to the size of the riser. I should like the author to elaborate a bit further on this matter.

It is also hard for me to realize how it would be possible to burn the graphite rod, for it would be more probable that there would be very little oxygen available to support the combustion and more likely the carbon would be dissolved and form an iron carbide which would tend to raise the carbon content of your riser.

H. H. BLOSJO[2]: In Fig. 3 you showed one riser of 42 lb. in which you did not use a graphite rod and another riser of 27 lb. in which you did use a graphite rod. I can not see how it would be possible to obtain 1.40 per cent carbon if no graphite rod was used.

MR. JONES: They both had the graphite rod.

MR. BLOSJO: It is incorrect in the paper then.

JOHN WILLIAMS[3]: Did you wash one side of the core?

MR. JONES: We washed one side of the core only, next to the casting. I think it would have the same effect as if we had increased the thickness of the core.

S. W. BRINSON[4]: I am a little confused and I would like to be straightened out. You showed a number of illustrations in which the risers were changed. Do you mean to infer by using the Washburn core you can cut down the number of risers on the casting?

MR. JONES: No. By using the Washburn core and the same size riser, for some unknown reason we did seem to get less shrinkage.

MR. BRINSON: I noticed in some cases you gated in the risers and in some you did not. That is one of the best means of obtaining direct solidification, by gating into the riser.

MR. JONES: That is right. I pointed out wherever possible to increase the number of gates so they would be close to the riser when the metal was coming into the mold.

MEMBER: How thick was the core of that 3-in. thick casting?

MR. JONES: Do you refer to the Washburn riser core? At the point where the riser is formed it was about 5/16 in. and out toward the end it was over ½ in., probably close to ⅝ in.

[1]Westinghouse Electric Corp., Philadelphia, Pa.
[2]Minneapolis Electric Steel Castings Co., Minneapolis.
[3]Dodge Steel Co., Philadelphia, Pa.
[4]Norfolk Navy Yard, Portsmouth, Va.

HEAT TREATING FURNACES

Although specifically directed toward the metallurgist and engineer, the present paper establishes several leads into the more practical field of furnace operation. It is hoped to follow up with a paper dealing with points of operation which will be of particular interest to furnace operators.

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Research Associate
Columbia University
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SELECTION OF FURNACE TYPES for the heat treatment of castings—stress relief, annealing, malleableizing, heating for quenching, tempering—is the responsibility of the metallurgist and engineer, to whom this paper is, in the main, addressed.

Furnace Classification. Several ways for convenient classification of heat treating furnaces are available. One refers to the mechanical design of the furnace, another to the source of heat or energy, and a third to the method of heat transfer. These three ways of classification are briefly reviewed in the following paragraphs.

Grouped according to mechanical design, the most commonly used furnace types in foundry work are the following: box type, car bottom type, elevator type, bell type, and continuous furnaces. Among the latter, conveyor furnaces, pusher type furnaces, walking beam furnaces are examples. Figure 1 shows diagrammatic sketches of some of these furnace types.

Grouping of furnaces for heat treating purposes in foundries may be based on the fuel or energy form applied. However, the fuel or energy form has much less influence on quality of the product than is frequently believed, and the question of best results has to take into consideration many other factors besides the form of energy, factors which are frequently more important.

In the third method of classification, referring to the mode of heat transfer from the heat surface to the charge, three basic types, namely, radiation, convection, and conduction furnaces, can be distinguished. Combinations of the three types are sometimes used. In radiation furnaces, heat transfer from the source to the charge is by radiation only (Fig. 2A). Examples are resistor furnaces and radiant tube gas furnaces. In convection furnaces, heat transfer is carried out by movement of gases or air. The atmosphere is heated by an energy source which in pure convection furnaces is located outside the furnace chamber, as shown in the sketch (Fig. 2B).

Circulation is enforced by blowers or fans. Electric resistors or gas or oil burners can be used as energy source; if gas or oil burners are used it is necessary to bleed part of the circulating atmosphere, but a large part can be recirculated. Conduction furnaces are lead or salt baths, in which the heat is transferred to the charge by conduction alone. Among the salt bath furnaces the electrode salt bath (Fig. 2C) deserves particular mention. In this furnace the salt itself serves as resistor and as heat transfer medium to the charge.

Combinations of radiation and convection furnaces or of radiation and conduction furnaces are used in most cases. Particularly, only rarely are pure radiation or pure convection furnaces applied. In radiation furnaces, the atmosphere of the furnace chamber circulates by natural convection, even if no blowers or fans are used. In convection furnaces, the walls radiate to the surface of the load.

Selection of the correct furnace

Presented at a Malleable Foundry Practice Session of the Fiftieth Annual Meeting, American Foundrymen's Association, at Cleveland, May 7, 1946.



Loading white castings into a continuous malleableizing furnace.

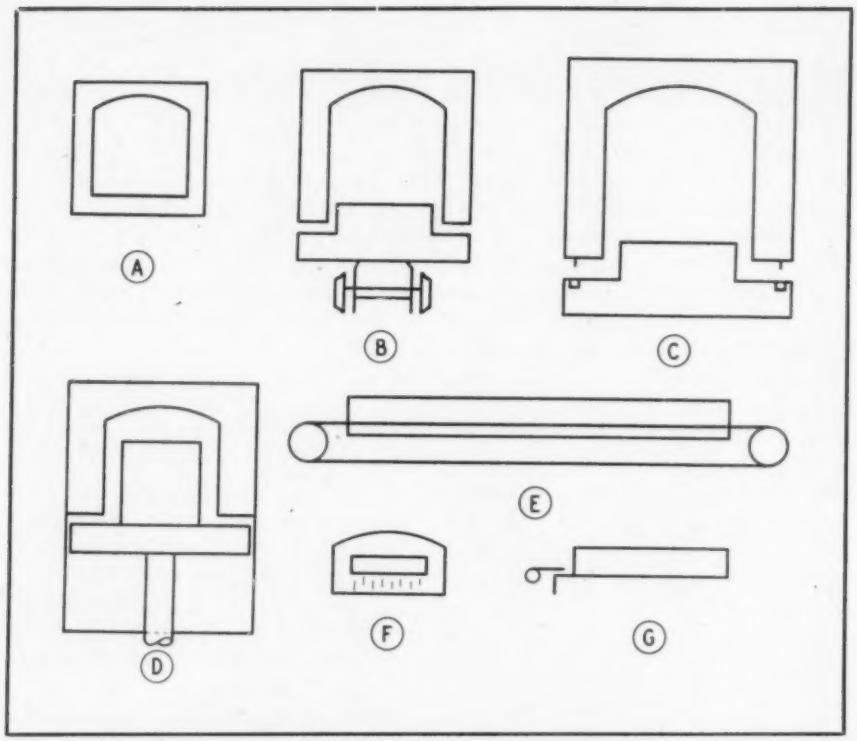


Fig. 1—Diagrammatic sketches of heat-treating furnaces. A—Box type; B—car bottom; C—bell type; D—elevator type; E—conveyor; F—walking beam; G—pusher type furnace.

type among these various possibilities should take into consideration a number of different problems, some of which are often overlooked. Therefore, the basis for selection will be discussed first. Anticipating the later findings, it may be stated here that, for obtaining highest quality, continuous furnaces would be desirable, but their use offers some difficulty in satisfying the other requirements. Later sections will therefore deal with the problem of how to reconcile the different requirements.

Basis of Furnace Selection. Furnace type selection should take into consideration the following items:

1. Uniformity requirements.
2. Heating and cooling requirements.
3. Individual size, shape, cross-section and weight of the greatest casting to be heat treated.
4. Desired temperature range of work.
5. Desired output of the furnace.
6. Space requirements.
7. First cost, including cost of gas, oil or electric lines from the point of delivery to the plant to the furnace.
8. Operating costs, including cost of energy or fuel, labor, rejections, cost of maintaining fuel reserves.

Uniformity; Heating and Cooling Requirement. In this listing the uniformity and heating and cooling requirements were placed first because they are so frequently overlooked. In order to understand them, it is important to define the terms used.

In furnace work, three types of uniformity must be considered; namely, uniformity in time, uniformity in space, and thermal uniformity in the workpiece. *Uniformity in time* is measured by the temperature recording instrument and demonstrated by its record. However, a high degree of uniformity in time proves only that the temperature within the thermocouple protection tube is constant and does not give any information concerning the uniformity of the workpiece, which is the only uniformity governing the quality of the product.

Temperature uniformity in time is a necessary prerequisite for, but in no way a proof of, thermal or other uniformity of the product. Figure 3 may serve as an example. The almost smooth line (a) is a typical temperature record, showing the temperature inside the thermocouple protection tube. The wave line (b) shows the temperature outside the tube. The thicker the tube, the greater is the relative difference between the two curves.

Furnace Temperatures

Uniformity in space refers to the temperature at various places within the furnace chamber, mostly with reference to the empty chamber. Although this uniformity is important for obtaining thermal uniformity of the workpiece, it is not sufficient for safeguarding it and depends on the way of loading the furnace, not on furnace design alone.

Thermal uniformity of the workpiece includes three requirements which should be fulfilled at all points of the charge; namely, equal temperatures of all points in the piece at the end of the heating or cooling process, equal rates of heating or cooling (degrees per unit time), and equal time at any temperature beyond the room temperature. These requirements can never be fulfilled entirely in practice because heating and cooling can be carried out only under the influence of temperature differences.

At times, in foundry practice, the differences in all three respects are so far from the requirements for uniformity that uniform products are

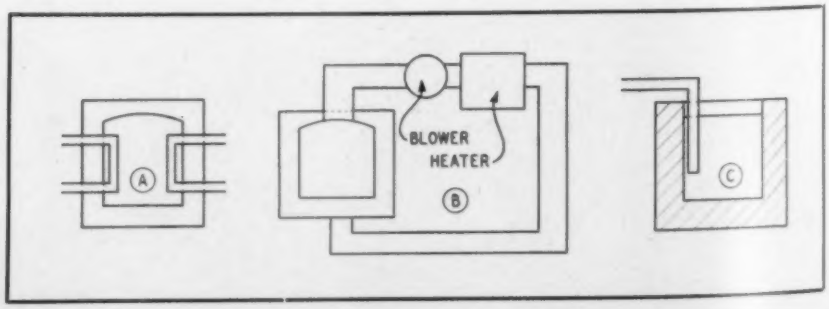


Fig. 2—Furnace types according to heat transfer. A—Radiation type; B—convection type; C—conduction type furnace.

not obtained. The degree to which these requirements are fulfilled depends on furnace design and operation.

It can be proved that the best way of heating and cooling to approach uniformity is to expose each workpiece individually. In heating without individual exposure—for example, when heating piles of castings in radiation type furnaces—the castings located on the outside will be at a high temperature long before the pieces in the center. Moreover, they will heat at a much higher rate than will the castings in the center.

Finally, each casting, if it is of any appreciable size, will heat non-uniformly, the part closest to the heat source heating at a faster rate and to a higher temperature than the part of each casting away from the heat source.

Figure 4 shows schematically the temperature rise in a charge composed of small castings, heated from the side walls of a furnace 4 ft. wide. The furnace walls are assumed to be at a constant temperature of 1600° F. In Fig. 4 time is plotted as abscissa, and temperature as well as rate as ordinates, in logarithmic scale. This kind of scale is, of course, not proportional. The distance from "1" to "2" is larger than the distance from "9" to "10." On the other hand, the distance from "1" to "2" is the same as the distance between "10" and "20."

Heating and cooling requirements comprise the desired temperature of the casting and minimum and maximum rates. It is basically unsound to specify only one temperature for heat treatment because, as explained in the foregoing, complete temperature uniformity can never be achieved in practice. Two final temperatures, one indicating the maximum, the other the minimum permissible temperatures, should be shown in the specifications (for example: "heat to a temperature of between 1550-1600° F.") Based on these temperatures the necessary heating time for a given charge can be calculated with reasonable accuracy.

Assuming that the furnace temperature remains constant during the heating process, only one value of heating time and one furnace temperature will produce a charge in which, at the end of the heating time, the surface has reached the maximum temperature, and the center the minimum temperature. Figure 5 shows the heating curves for surface and center of a large casting of uniform cross section. The difference between the two lines decreases slowly, and the heating time is therefore practically determined by the desired difference at the end of the heating time.

Maximum and minimum temperatures and rates generally are prescribed either on the basis of experience alone or on the basis of laboratory experiments with samples of small size. The values determined for small samples must not be used for heavy castings without change. It is necessary to transform them according to the size and shape of the casting, and with piles of castings, according to the mutual arrangement. Moreover, the mode of heat transfer enters the picture.

It is no longer sufficient to esti-

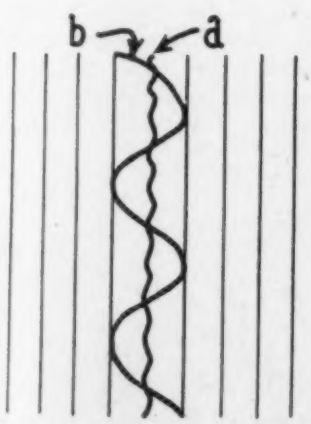


Fig. 3—Typical temperature curve; (a) record within thermocouple well; (b) curve outside the well.

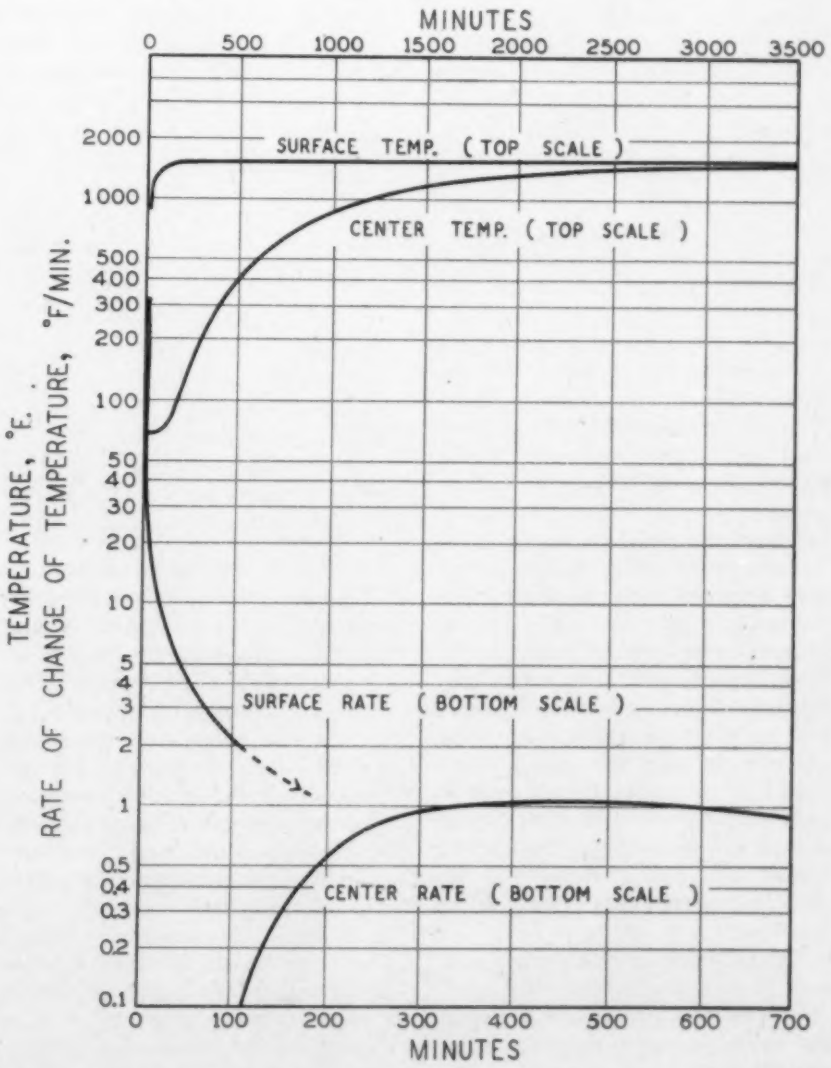


Fig. 4—Temperature rise in charge composed of small castings. In addition to temperature curves, the chart also contains curves for the rate of temperature rise (° F./min.).

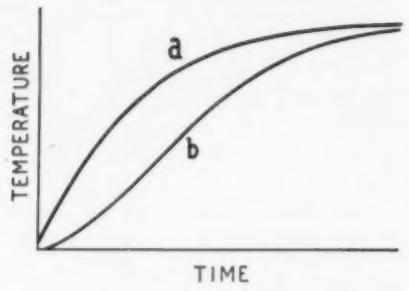


Fig. 5—Typical heating curves for surface (a) and center (b) in heat treatment of large castings.

mate the influence of these factors on necessary heating or cooling time to obtain the desired final temperatures and rates, and it is equally misleading and wrong to assume that the temperatures and rates measured at any one spot within a charge apply to the entire charge equally. Today it is possible rationally to transform the values obtained on samples of small size into values for various arrangements of the charge and various furnace designs.

Heating Rate

Assume that the temperatures and rates have been found acceptable when exposing a cylindrical sample of 0.5-diameter to a furnace temperature of 1800° F. for 3 min. From available charts it can be found that at the end of this period the surface temperature of the sample is 1679° F., and the center temperature is 1675° F. The rate of heating at the surface after 1½ min. is 0.121° F. per sec., and, at the same time, at the center is 0.126° F. per sec.

Also, from existing charts one can now determine that for a rectangular block 4 in. thick, the furnace temperature would have to be 1720° F. in order to obtain the same temperature difference between surface and center at the end of the heating period, and that the heating time would be 138 min. For this condition, the rate of heating after one-half of the heating time (69 min.) would be 0.631° F. per sec. at the surface, and 0.693° F. per sec. at the center.

Maximum Individual Size, Shape and Weight; Output and Space Requirements. In commercial foundries the output in tons per day is subject to daily fluctuations, whereas in production foundries the output can be considered fairly constant over a certain period of time. For the production foundries, no special remarks are necessary. In commercial foundries operations can be made much more satisfactory if the practice of heat treating every casting immediately after cleaning is abandoned. Rather, certain planning of the heat treating is desirable, because operations are more satisfactory if only pieces of the same or almost the same size are heated simultaneously.

Accumulation of castings over several days makes possible the loading of furnaces more nearly as desired. The operation will be more satisfactory the longer the period which can be averaged. For example, if the material is stored up to a maximum of one week, operation will be more satisfactory than if the maximum time of storage is three days. Such grouping of the output, of course, makes it necessary to store part of the output, and this involves problems of rapid delivery and of space requirements.

Castings of one size are often made of steels of different composition, a fact which makes the foregoing recommended practice less desirable. Inasmuch as the composition cannot be recognized from the outside, castings of different composition may become mixed in heat treating. Moreover, with the same theoretical composition, the actual compositions may change slightly from day to day. Consequently, from this point of view, it is desirable that pieces cast on various days be not heat treated in the same furnace operation.

Furnace and Storage Space

Space requirements include the furnace proper and control equipment, as well as possible storage space for production, as mentioned in the preceding paragraph. Floor space generally is larger for continuous furnaces than for batch-type furnaces. A comparison between batch-type and continuous furnaces must, of course, be based on the same output and should be based on the same degree of uniformity and heat.

As previously explained, a degree of uniformity similar although not equal to that in continuous furnaces with individual exposure of each casting can be obtained in batch-type furnaces when heating piles of castings. In the latter case, much longer heating and cooling periods are required than in the continuous furnace. Longer heating times in turn call for larger volume, and this in turn means more floor space. The actual advantage in floor space of the batch-type furnace as compared with the continuous furnace must be attributed to the greater height of the batch-type furnace for equal volume.

Operating Cost. The furnace operating cost should include not only fuel, labor and maintenance, but should take into consideration in some way the quality of the product. If by one method of operation the quality of the product can be increased as compared with other methods, then it is reasonable to introduce into the cost balance calculation an item covering this improvement. This can be done either by lowering the amount for the expense caused by rejections or by introducing an increased amount for the income from a higher quality product.

Cost Comparison

No general statement concerning operating costs of different furnaces is possible. If only the output is taken into consideration without comparing the achieved uniformity, then continuous furnaces are always more expensive than batch-type furnaces. However, if several batch-type furnaces were installed to yield the same uniformity as the continuous furnace, then the first cost of the batch-type furnaces may be larger than that of a continuous furnace of equal output.

Furnace Type Selection. Problem: It has been stated in the foregoing that the requirements of uniformity are best satisfied in furnaces designed for individual exposure of each casting. This requirement must be reconciled with the requirements for space, first cost, and diversity of production. How the three main types

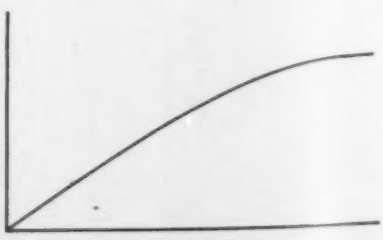


Fig. 6—Example of desired temperature rise of the heating gases in a furnace with progress control.

previously mentioned satisfy these different needs will now be considered.

For piles of small castings, the conduction furnace, although rarely used at present, would be extremely efficient. The salt penetrates rapidly into the spaces between the castings and surrounds each casting quite uniformly. The heating will be rapid and, provided that the castings have the same shape, they will heat uniformly. Although the main advantage of a conduction furnace would be obtained with small castings, it is possible to heat large castings in such furnaces. The large castings are of a reasonably uniform cross section and if rapid heating is permissible, then satisfactory results may be expected.

Rate of Heat Loss

Because of the rapidity of heating in the salt-bath furnace, high output can be obtained. Generally speaking, salt-bath furnaces have a relatively high rate of heat loss because of radiation from the bath surface. No existing cover design entirely overcomes this difficulty. Therefore, the operating costs are quite high if the castings must be held at elevated temperatures for any length of time.

Convection-type furnaces. In such furnaces piles of castings may be heated with reasonable uniformity provided that the flames or the hot air are forced through instead of merely washing the surface of the pile. In such a furnace the castings close to the entrance of the flames or air are heated more rapidly than those near the exit of the heating gases. To overcome this disadvantage and obtain a satisfactory degree of uniformity, the furnace must be equipped with so-called program control. The temperature of the heating gases initially should be low and increase only gradually as the charge picks up heat (Fig. 6). This method of operation results in an acceptable compromise between the desire for high output and good uniformity.

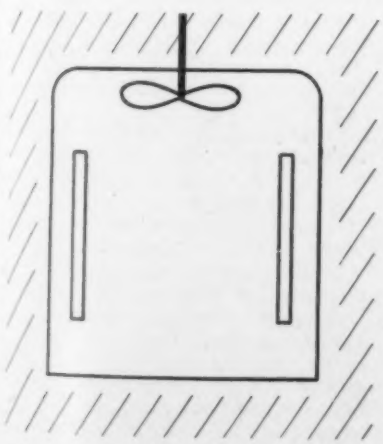


Fig. 7—Sketch of mixed convection radiation furnace.

Temperature limits of such furnaces are in the order of magnitude of 1700° F. It should be understood that reference is made to pure convection furnaces, and that the advantages claimed do not apply to radiation furnaces in which fans are used to whirl the air (Fig. 7). This rules out furnaces with resistors or radiant combustion tubes in the furnace chambers. The requirement of program control eliminates the possibility of direct firing. The furnaces, then, must have a separate heating chamber in which resistors or burners are located. Hot air or combustion gases are circulated through the charge by means of a powerful fan.

Furnace Types Compared

For individual exposure of the casting, convection furnaces generally do not offer any large advantage as compared with radiation furnaces.

Radiation Furnaces. If the temperatures are too high for convection furnaces, and if conduction furnaces (salt bath) must be ruled out, then the only type of furnace left is the radiation furnace. These may be resistor furnaces, radiant-tube furnaces, muffle furnaces, or direct-heat-fuel furnaces. The next problem to be considered is the type selection of such furnace (over stationary furnaces).

If it is possible to store castings so that for a period of at least several hours the continuous furnace can be loaded with castings of equal size and shape, then the continuous furnace with individual exposure of the castings is the desirable solution. If the storage of castings proves impossible, then the continuous furnace must often be ruled out because it may not work satisfactorily if pieces of different size pass through it. Moreover, in many plants only batch-type furnaces are available at present and funds may not be available for immediate replacement.

More satisfactory results may be obtained from batch-type furnaces if special attention is paid to the manner of loading and if special care is taken in the arrangement of the heating means.

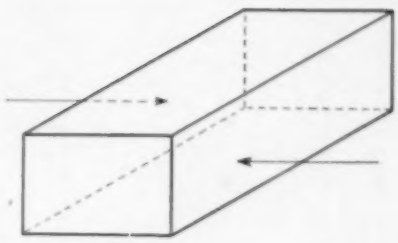


Fig. 8—Correct heating procedure for rectangular pile of castings.

A difficulty frequently encountered is that the corners of the rectangular charge may heat more rapidly and to a higher temperature than the remainder of the charge surface. This higher temperature may exceed the permissible maximum. In order to avoid this, it is desirable to heat only two, not four or six, sides of the rectangular pile of castings (Fig. 8). In the drawing, heating from two sides is shown. The same conditions would prevail if only top and bottom or only front and rear were heated. The other four sides of the furnace not used for heating the castings should be heated only to the extent of providing for heat lost through walls.

Heating Uniformity

The thickness across which heating occurs is of decisive influence on the uniformity obtained. Generally speaking, the heating times necessary to obtain a given degree of uniformity change with the square of the heated thickness of an individual casting or a pile. In the case of heating one casting only, no further comment is necessary. However, if a pile of castings is heated, the decrease in the thickness of the pile will lead eventually to the point. Where only one casting remains, then the heating time to obtain the same uniformity within the casting is still further and greatly reduced.

One method of improving the operation of batch-type furnaces is in loading smaller piles, which will heat much faster to the same uniformity as previously, instead of utilizing the full height of the furnace by loading the thickest pile possible. Uniformity can be improved considerably by slightly reducing the

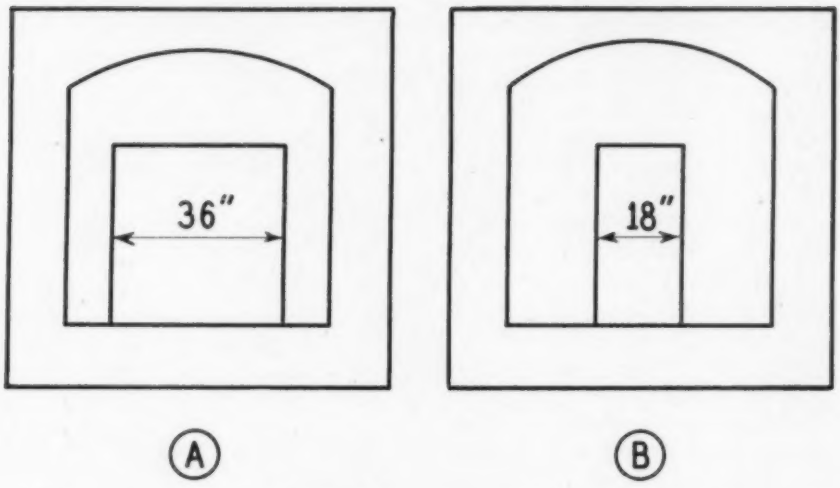


Fig. 9—Size of charge in box type furnace. A—3 ft., heating time, 8 hr. B—1½ ft., heating time, 2 hr.

furnace temperature and increasing the time of the charge.

As an example, let us say that a furnace with heat on two sides only heats at present a 3-ft. pile of castings in a heating time of 8 hr. (Fig. 9A). If the width of the pile is reduced to 1½ ft., approximately the same uniformity would be obtained in 2 hr. of heating time (Fig. 9B).

Increasing the heating time to 3 hr. and simultaneously lowering the furnace temperature will result in two advantages; the same output will be heated in a total of 6 hr. (two batches of 3 hr. each) as compared with the previous 8-hr. period and, in addition, much more uniform castings are obtained. The disadvantage in the procedure is the increased labor of load-charging and discharging the furnace twice. This is a price that should be paid in order to improve the uniformity.

If the total heating time is several days so that the reduction of the size of the pile leads to heating times which are inconvenient because they interfere with the change of shifts, it may be necessary to add batch-type furnaces. If several furnaces are used, it is desirable to have furnaces of different size, which yields great flexibility.

Heating of castings in piles is undesirable from the point of view of uniform production.

Disadvantages usually inherent in this method can be partly overcome by using salt-bath furnaces if permissible, or furnaces with forced convection and program control.

If neither method is possible, then continuous furnaces should be applied. If they too cannot be used, special attention should be given to distribution of heating means and the thickness of the heated pile in radiation and combined radiation and convection furnaces.

Acknowledgments

The first draft of the manuscript was submitted to a number of foundrymen and valuable comments and criticisms were offered by the following: W. D. Bailey, Jr., Westlectric Castings, Inc., East Los Angeles, Calif.; G. K. Dreher, Ampco Metal, Inc., Milwaukee; W. Finster, Reading Steel Casting Div., American Chain & Cable Co., Reading, Pa.; J. W. Juppenlatz, Lebanon Steel Foundry, Lebanon, Pa.; G. A. Lillieqvist, American Steel Foundries, East Chicago, Ind.; J. T. MacKenzie, American Cast Iron Pipe Co., Birmingham, Ala.; G. P. Phillips, International Harvester Co., Chicago; H. A. Schwartz, National Malleable & Steel Castings Co., Cleveland; J. S. Vanick, International Nickel Co., New York; J. D. Wolzney, American Steel Foundries, East Chicago, Ind.; and L. Brown, Magnesium Fabricators Div., Bohn Aluminum & Brass Corp., Adrian, Mich. The author wishes to express his sincere gratitude for their cooperation.

DISCUSSION

Chairman: A. M. FULTON, Northern Malleable Iron Co., St. Paul, Minn.

Co-Chairman: C. F. JOSEPH, Saginaw Malleable Iron Div., General Motors Corp., Saginaw, Mich.

DR. H. A. SCHWARTZ[1]: I should like to endorse and emphasize some of the things Dr. Paschkis said. It is not sufficient just to build a malleable annealing furnace. If you want uniformity of temperature, you cannot have very quick heating because, even if everything else were right, you would still have to maintain temperature gradient in the casting. If all other gradients were eliminated, the casting gradient would remain.

It is suggested that we make an hourly charge, this in one dimension. Dr. Paschkis made reference to the fact that with the continuous furnace of the conveyor type, we have a charge of only one or two castings thick.

The points he makes are eminently true, and they are of enormous importance. Each one of those factors means that you have a small cross-section; a small furnace; that you are confronted with what finally amounts to a very considerable investment in your heat-treating equipment.

You are confronted with that, and you finally end up with the question: Where is the point at which it is cheaper to have something less ornate in furnace equipment and a longer annealing cycle to do the work, as compared with shortening your annealing cycle and trying to economize by the greater annual output of a given unit?

I think all users of malleable castings, most sellers of furnaces for malleable castings, and some malleable foundrymen do not adequately understand that there is this balance of cost of the product and time in which you can deliver a product, and that, generally speaking, if you are going to stay in business you have to approach the economic minimum.

The customer is very seldom willing to pay any more money to get the casting sooner. Naturally we reach the point where increased output costs money but the product cannot be sold at the price necessary to meet the costs of production.

The conditions that Dr. Paschkis outlined are such that they will help evaluate the relative merits of one type of furnace against the other, and aid in selecting that particular equipment which, for a particular industrial condition, produces the best and most salable product.

In the days when I had quite a little contact with Charles Kettering, his statement was that that material is the best material which actually fulfills the minimum requirements for the least price. His statement is also true of heat-treating.

DR. PASCHKIS: I want to thank Dr. Schwartz for his remarks upholding the tendency of this paper. I particularly agree with his statement on the necessity of finding an economic optimum between quality and cost of production.

MEMBER: Some people claim that it is impossible to circulate hot gases with fans, while others claim that you can. Would you care to comment on that?

DR. PASCHKIS: I think there are enough furnaces in operation to show that you can circulate gases up to a temperature of 1700° F. Beyond that, fans do not seem to stand up well.

[1]National Malleable & Steel Castings Co., Cleveland.

R. V. OSBORNE[2]: The conditions in the heat treating of malleable iron are somewhat different than for heat treating some other metals. I suppose you are implying that the design of the furnace should also include the efficiencies and temperature differentials you get in your casting mass, on cooling as well as on heating. Would not that have some bearing, in cooling slowly through quite a range, on how you would design your furnace?

DR. PASCHKIS: Yes. If you cool in the furnace, very definitely it would, and for malleable iron you have to do so. What was said here for heating holds true for cooling. It is still more difficult in cooling than heating to obtain uniformity in the charge. In heating, it is by and large a question of time alone. In cooling you are bound to get great differences if you take the casting out of the furnace. If you cool in the furnace, it would be the same proposition. It is conceivable to build furnaces which would allow the temperature to drop very slowly until, at a known critical value, it would permit the temperature to drop rapidly.

L. C. KIMPAL[3]: Is under-firing alone better than over- and under-firing for more uniform heating of the mass?

With two sets of burners, one set firing under the arch and one set firing under the hearth, using twin nozzle burners below, shutting off the gas on one nozzle and introducing air only through this nozzle do you not get pretty even distribution in the furnace through a wide range of temperature, say from 600 to 2000° F.? This method of firing is common practice in the heat treating field.

DR. PASCHKIS: Under-firing is frequently better than over-firing because, by natural convection, there will be some heating also from the top. Even if you have only one piece, if you heat from two sides it is better than from one side alone, but it is not better to heat from three, four, or six sides. Heating from two sides is better than from one side.

The second question is one of having three sets of burners. It is possible to work it out as described. However, I do not think you could set up a general rule to that effect. Apparently you limit your remarks to one specific furnace design.

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[2]Lakeside Malleable Castings Co., Racine, Wis.
[3]Rochester Gas & Electric Corp., Rochester, N. Y.

MICROSTRUCTURE RELATED TO PROPERTIES OF CAST IRON

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CAST "iron," the term, actually is a misnomer since this material contains more carbon, as well as other alloying elements, than ordinary steel. Some of the carbon, as in steel, is in combined form. In white cast iron substantially all of the carbon is combined with iron as cementite, and thus "white iron" is similar in microstructure and properties to a very high carbon tool steel.

In gray cast iron, however, usually at least three-fourths of the carbon is in the elemental form; i.e., graphite. Mottled irons consist of mixtures of the structures of white and gray cast iron so that many small areas of each are observable in a fracture surface through such an iron or in the microstructure.

Most of the following discussion is concerned with the microstructure of cast irons which are "gray" when sand-cast in moderately thick or heavy sections.

The microstructure of gray cast iron usually is made up of a matrix similar to that of steel, consisting of all or part of the constituents pearlite, ferrite, and massive cementite and, in addition, containing flakes of graphite. The structure of a typical pearlitic steel is shown in Fig. 1. The similarity of this structure to that of the typical gray cast iron of Fig. 2 is readily apparent.

Nature of Microconstituents

Graphite. Graphite is substantially pure carbon which exists in all gray cast iron in the form of small flakes of irregular shape. When magnified at 100 diameters these flakes appear in the microstructure as dark gray curved lines varying in length from a fraction of an inch to several inches. Thus, their actual length is of the order of a few hundredths of an inch or less. The graphite flakes can be detected in the microstructure in both the unetched and etched conditions.

Graphite is soft and weak and, as will be shown later, has a pronounced effect upon the various properties exhibited by the "steel" matrix. Some of these effects are advantageous, some disadvantageous—depending upon the application.

The dispersion of several per cent of graphitic carbon as flakes in cast iron sometimes results in the erroneous conclusion that cast iron normally is porous. Macro-porosity may occur in gray iron castings just as in any cast metal, but the microstructure of cast iron is inherently sound.

Such soundness is shown in all of the photomicrographs accompanying this discussion as well as in those throughout the literature. The inherent soundness of cast iron is well demonstrated in a recent paper by MacKenzie.[1]

Cast Iron Inherently Sound

Reports of "porous" iron encountered during the polishing of castings such as that shown in Fig. 3 have been investigated repeatedly. If porosity were inherent in the material, such very heavy cast sections would accentuate the effect.

In each case investigated, it was shown that the "porosity" was the result of the tearing out of graphite flakes and the metal surrounding them during machining, or to pitting out during polishing. In each case by adjustment of machining, grinding, and polishing methods the "porosity" was eliminated.

Graphite flakes occur in cast iron in several types of distribution and in a wide range of size. The A.F.A. and A.S.T.M. jointly have prepared a standard chart[2] for the purpose of reporting graphite flake size by num-

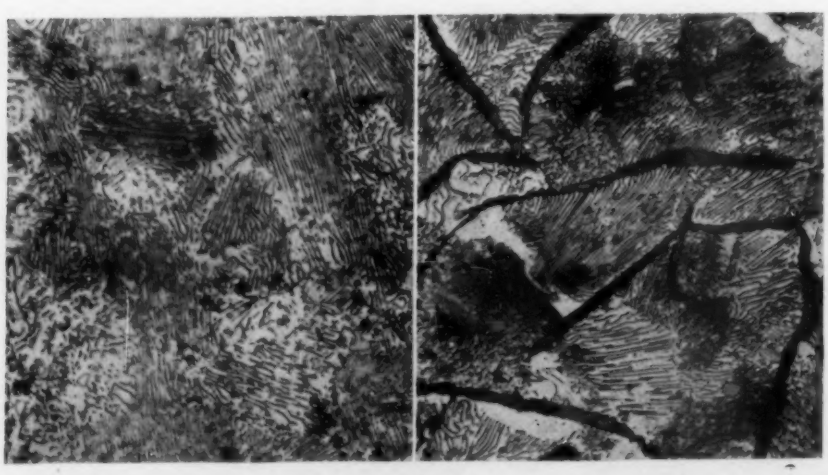


Fig. 1 (Left)—Steel containing partially spheroidized lamellar pearlite. Nital etched, 500X.
Fig. 2 (Right)—Cast iron containing lamellar pearlite, graphite flakes, and a small amount of massive ferrite. Nital etched, 500X.

This paper was presented and discussed at a Gray Iron Session of the Fiftieth Annual Meeting American Foundrymen's Association, at Cleveland, May 6-10, 1946.



Fig. 3—Large cast iron thrust bearing runner polished by lapping to a surface finish of about 6 microinches, root mean square.

ber and type of distribution of flakes by letter.

Flake size varies from Size 1 (longest flakes 4 in. or more in length at 100 diameters magnification) to Size 8 (longest flakes 1/16 in. or less in length at 100 diameters magnification). Graphite flake distribution varies from Type A (uniform distribution (random orientation) to Type E (interdendritic segregation, preferred orientation).

Cementite. Cementite is iron carbide (Fe_3C), the chemically combined form in which carbon exists in ferrous metals. Part of the total carbon content of cast iron occurs as graphite, the remainder as cementite. The cementite in turn occurs in two forms, (1) as lamellæ with ferrite in pearlite, and (2) as primary or massive cementite. This massive cementite appears in the etched microstructure as irregularly shaped white masses such as those shown in Fig. 9b.

Massive cementite is extremely hard (550 Bhn.) and wear resistant. It is detrimental to machinability even when present in relatively small amounts. The presence of massive cementite is desirable only when wear resistance is of importance.

Ferrite. Ferrite is practically pure iron. It occurs as (1) lamellæ associated with lamellar cementite in pearlite and (2) as free or massive ferrite. Although radically different in properties from cementite, massive ferrite appears somewhat similar to cementite in the etched microstructure, but the two can be differentiated by a slight difference in color and lustre.

Massive ferrite is a microconstituent of low hardness (75 Bhn.), low strength, and high ductility. Its presence is beneficial to machinability (when associated with graphite as in cast iron), but is detrimental to strength and wear resistance.

Pearlite. Pearlite is a laminated microconstituent composed of alternate plates or lamellæ of cementite ferrite. Accordingly, its properties are roughly an average of those two materials. Pearlite is stronger and harder than ferrite but appreciably softer and less brittle than cementite. The appearance of pearlite in the microstructure is shown by etching.

Other Microconstituents. Steadite is an iron-carbon-phosphorus eutectic peculiar to cast irons. It is hard and brittle, and its effects are similar to those of cementite. In the normal low-phosphorus irons now being produced for general engineering purposes the small amount of steadite which occurs in small, well-distributed masses is considered to have a rather small effect upon the properties of the iron.

Various other microconstituents such as austenite, bainite, and martensite may be found in cast iron as in steel, but do not occur in ordinary grades of cast iron except when given special heat treatments or specially alloyed.

A thorough treatment of the microconstituents of cast iron and the methods for their examination is presented in an A.F.A. publication by Allen.[3] Reference also may be made to the material and bibliographies pertinent to this subject which are included in the 1944 edition of the Cast Metals Handbook, published by A.F.A.

Effect of Cooling Rate and Composition Upon Microstructure

Figures 4, 5, and 6 illustrate some of the important effects of the cooling rate from the freezing point, and of the composition, upon the microstructure. The changes in microstructure in turn affect the properties of the iron. The effect of cooling rate upon the microstructure and properties of a Class 20 cast iron is shown in Fig. 4.

The variation of cooling rate was obtained by sand casting solid cylinders of increasing diameter. The effect of the same cooling rate conditions upon a Class 40 iron is shown in Fig. 5. Therefore, the difference in microstructure and properties between the Class 20 and Class 40 irons of the same cylinder size is due to the difference in chemical composition.

The cooling rate of the iron shown in Fig. 6 has been controlled by the use of a heavy metal chill. Sections of the test piece close to the chill represent very high cooling rates, whereas sections of the test piece remote from the chill represent much lower rates.

The cooling rate of cast iron in a sand mold is determined primarily by the section size of the casting and the presence or absence of chills. The effects upon the microstructure of increasing cooling rate are:

1. Increase the tendency to form massive cementite and thus decrease the amount of carbon present as graphite.

2. Increase the fineness of pearlite (i.e., decrease the spacing of the lamellæ).

3. Decrease the size and change the distribution of graphite flakes.

4. Decrease the tendency to form massive ferrite. An exception to this is at certain high cooling rates where interdendritic graphite occurs. Ferrite tends to be associated with this type of graphite, as shown in Fig. 7.

Similar changes in microstructure

• Cast iron microstructures and attendant properties may be changed by quenching and tempering treatments to improve strength and hardness. Alloying elements may be used to develop microstructures of increased strength, hardness, greater wear resistance.

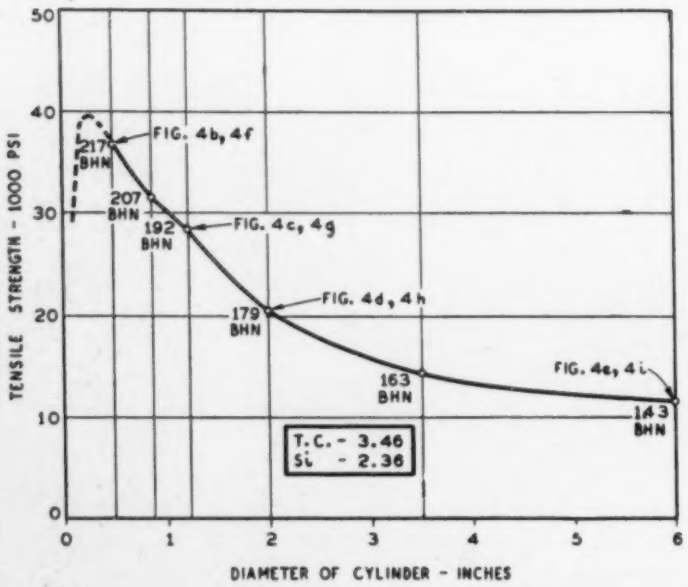


Fig. 4a—The effect of microstructure upon the tensile strength and hardness of a Class 20 cast iron. Test specimens were taken from sand cast cylinders of the indicated diameters.

Fig. 4b—Graphite distribution of Class 20 cast iron from ½-in. cylinder. Fine interdendritic graphite of rating 6E, 7D. Unetched, 100X.

Fig. 4c—Graphite distribution of Class 20 cast iron from 1.2-in. cylinder. Small random graphite flakes of rating 5A. Unetched, 100X.

Fig. 4d—Graphite distribution of Class 20 cast iron from 2-in. cylinder. Moderately sized random graphite flakes of rating 4A. Unetched, 100X.

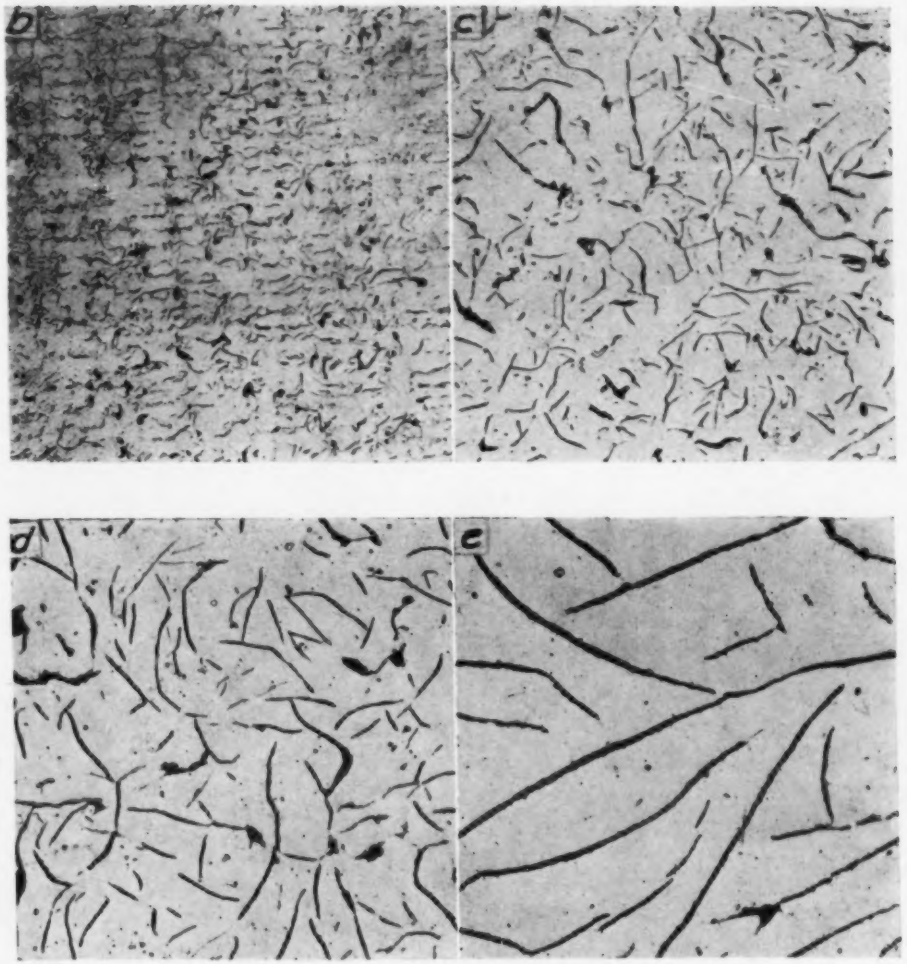
Fig. 4e—Graphite distribution of Class 20 cast iron from 6-in. cylinder. Very coarse random graphite flakes of rating 1A. Longest flakes are more than 4 in. in length at 100X. Unetched, 100X.

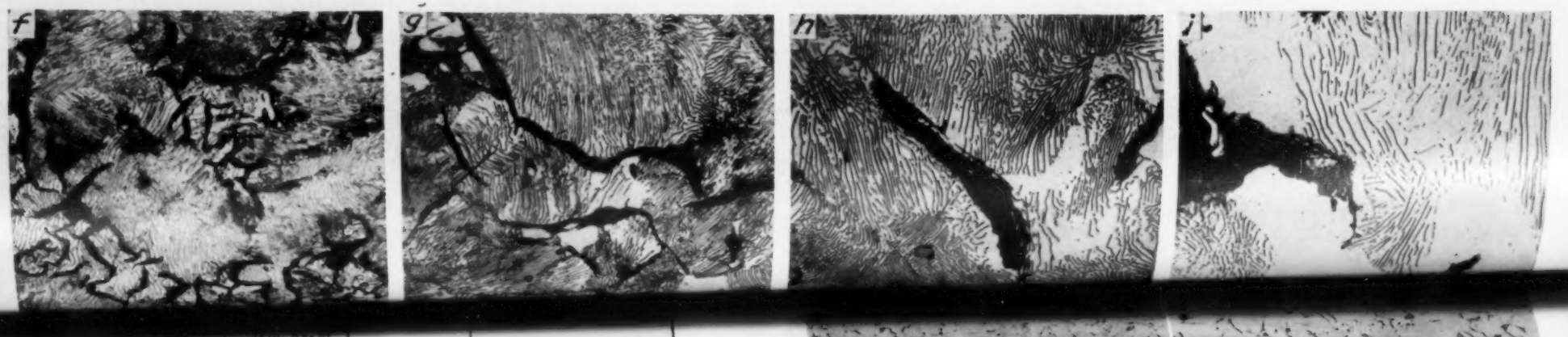
Fig. 4f—Microstructure of Class 20 cast iron from ½-in. cylinder. Specimen contains moderately fine pearlite and interdendritic graphite. Picral etched, 1000X.

Fig. 4g—Microstructure of Class 20 cast iron from 1.2-in. cylinder. Pearlite is slightly coarser than in Fig. 4f and a small amount of massive ferrite has formed. Picral etched, 1000X.

Fig. 4h—Microstructure of Class 20 cast iron from 2-in. cylinder. Specimen contains coarse pearlite and massive ferrite. Picral etched, 1000X.

Fig. 4i—Microstructure of Class 20 cast iron from 6-in. cylinder. The microstructure is composed of very coarse pearlite and a large amount of massive ferrite. Picral etched, 1000X.





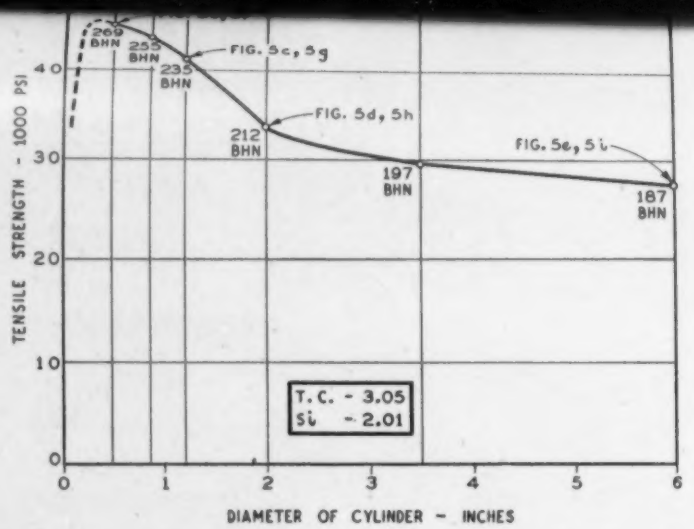


Fig. 5e—The effect of microstructure upon the tensile strength and hardness of a Class 40 cast iron. Test specimens were taken from sand-cast cylinders of the indicated diameters.

Fig. 5b—Graphite distribution of Class 40 cast iron from ½-in. cylinder. Fine graphite flakes of rating 6E, 7D. Unetched, 100X.

Fig. 5c—Graphite distribution of Class 40 cast iron from 1.2-in. cylinder. Distribution of mixed types and sizes of flakes (5A and 6E). Unetched, 100X.

Fig. 5d—Graphite distribution of Class 40 cast iron from 2-in. cylinder. Size 4, Type A random graphite flakes. Unetched, 100X.

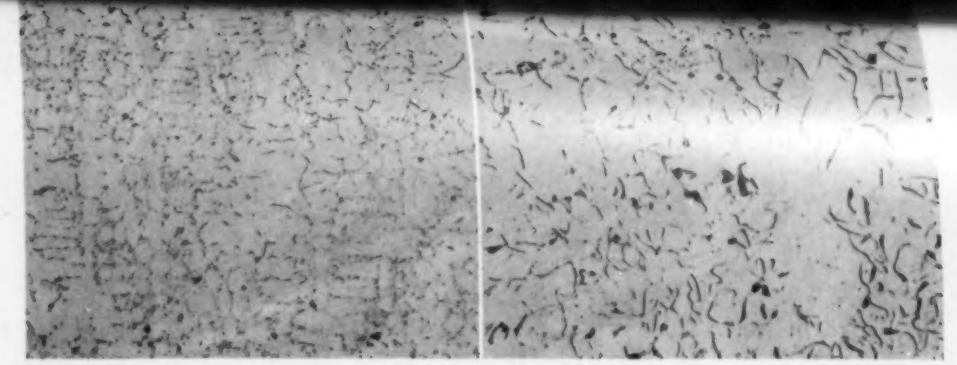
Fig. 5e—Graphite distribution of Class 40 cast iron from 6-in. cylinder. Size 3 random flakes. Note the small size of these flakes as compared to the large flakes found in the Class 20 iron in a 6-in. cylinder (Fig. 4e). Unetched, 100X.

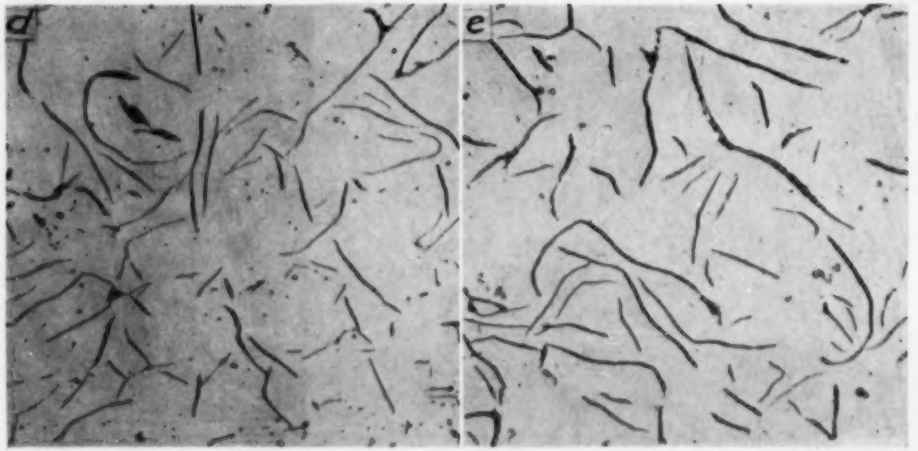
Fig. 5f—Microstructure of Class 40 cast iron from ½-in. cylinder. Matrix is composed of fine pearlite and a small amount of massive cementite. Picral etched, 1000X.

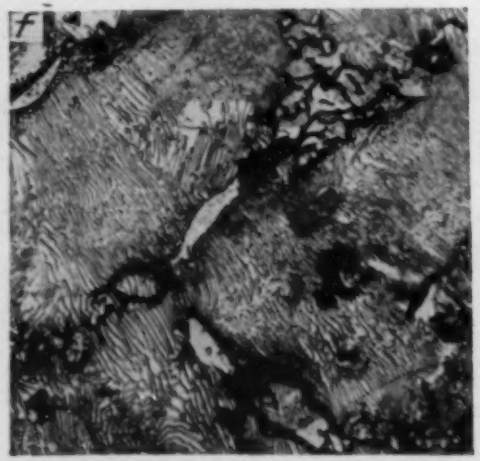
Fig. 5g—Microstructure of Class 40 cast iron from 1.2-in. cylinder. Moderately fine pearlite and traces of massive cementite. Picral etched, 1000X.

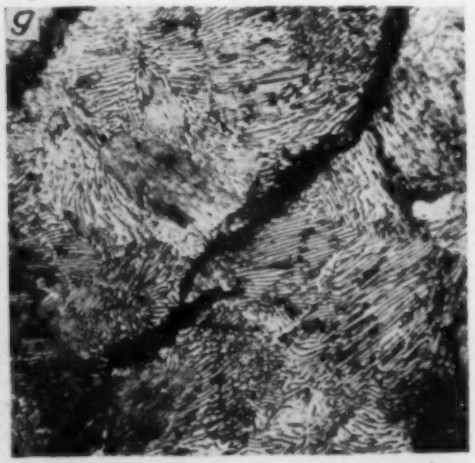
Fig. 5h—Microstructure of Class 40 cast iron from 2-in. cylinder. Matrix is composed entirely of moderately coarse pearlite. Picral etched, 1000X.

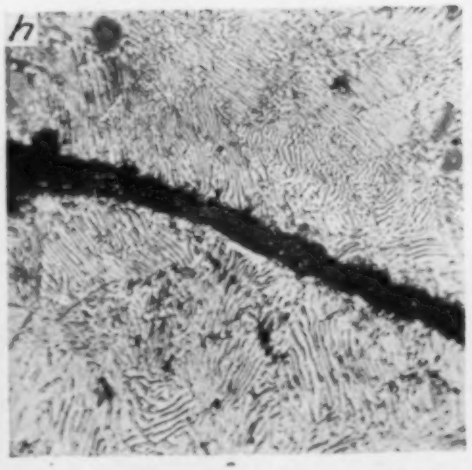
Fig. 5i—Microstructure of Class 40 cast iron from 6-in. cylinder. Matrix consists of coarse pearlite. Note absence of massive ferrite as compared to Fig. 4i. Picral etched, 1000X.

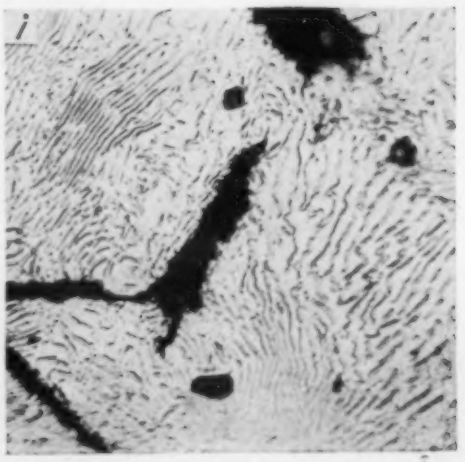












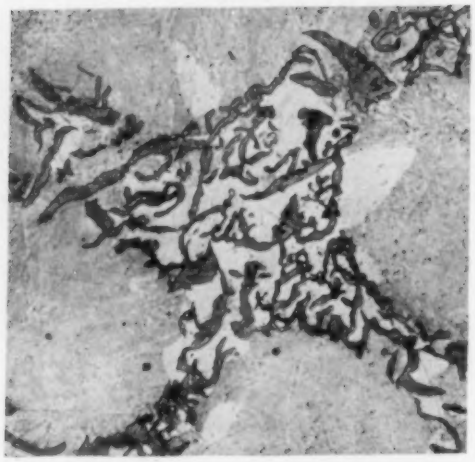


Fig. 7—Massive ferrite as associated with fine interdendritic (Type D) graphite flakes. Gray background is unresolved pearlite. Picral etched, 500X.

can be produced by changes in composition. Increasing the total carbon and silicon contents acts in the same direction as decreasing the cooling rate. Decreasing total carbon and silicon contents or increasing contents of stabilizing alloys such as chromium and molybdenum act similarly to increasing the cooling rate. This composition factor is illustrated by comparing of Figs. 4 and 5.

Effect of Graphite. Considering the matrix of cast iron as quite similar to that of steel, the effects of the graphite flakes upon the matrix properties are shown graphically in Fig. 8 and summarized as follows:

1. Reduce tensile strength, yield strength, and per cent elongation (ductility) due to interruption of the continuity of the matrix.

2. Exert little effect upon Brinell hardness. However, the hardness of a cast iron may be higher than that of a steel of similar combined carbon content due to the effect of solid solution elements such as silicon and phosphorus in strengthening the matrix.

3. Reduce the modulus of elasticity, specific gravity, thermal conductivity, thermal expansion, specific heat, and electrical conductivity.

4. Increase damping capacity because of the cushioning effect of the soft flakes.

5. Reduce notch sensitivity (increase the ratio of fatigue strength to tensile strength in notched bars) due to the stress distribution effect of the dispersed flakes.

6. Improve machinability because of lubricating action and the brittleness of the chip.

7. Decrease galling in bearing applications, again because of the lubricating effect imparted by the graphite flakes.

The amount, size, and distribution of the graphite flakes materially affect the properties of cast iron. However, due to the fact that the same factors that affect the graphite also affect the matrix, it is difficult to evaluate quantitatively the individual effects of the variations of the graphite.

If it were practical to develop various distributions and sizes of graphite flakes in a constant matrix, then it would be possible to evaluate the effect of various graphite dispersions. In the case of malleable iron, where the matrix is entirely ferritic, it has been possible to make such a correlation of the effect of primary graphite flakes upon physical properties.[4]

Therefore, the changes of properties with microstructure (as shown in Figs. 4, 5, and 6) are the result of combined changes of the graphite flakes and of the matrix. An attempt to determine semi-quantitatively the effects of microstructure upon properties has been reported recently by Adams.[5]

Adams considered as variables of microstructure: (1) graphite characteristics, (2) matrix structure, and (3) cell size. The cells studied by Adams are apparently the result of the segregation of lower melting point constituents to austenite grain boundaries.

Although the quantitative effects of changes in the microstructure of graphite on properties have not yet been determined, certain qualitative effects listed below are known:

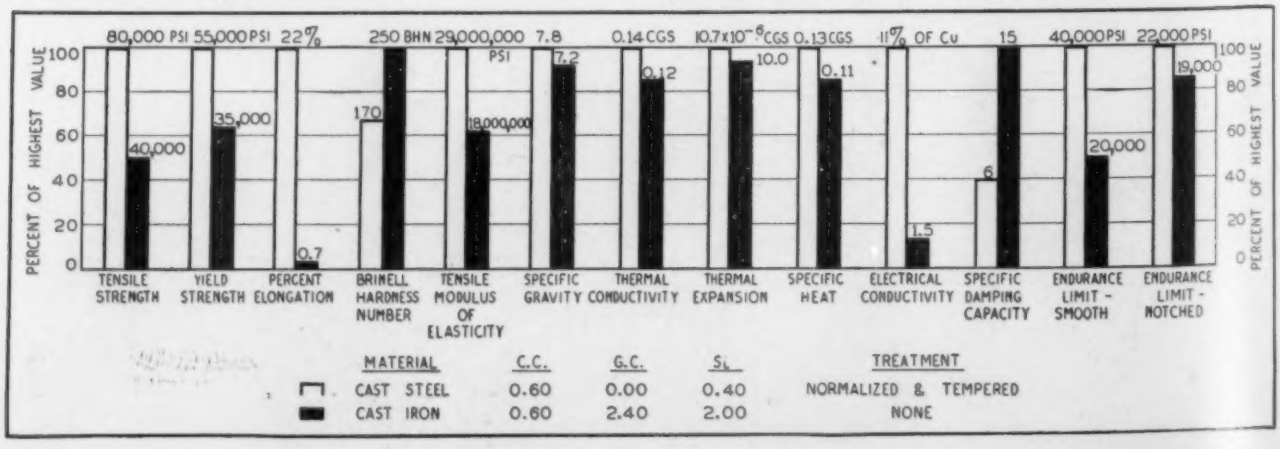
1. Increased amounts of graphitic carbon reduce strength and increase machinability.

2. Graphite flake distribution and size have little effect upon hardness which is controlled primarily by the structure of the matrix. Hardness is a function of composition and cooling rate, both of which affect the structure of the matrix and control the presence of massive cementite.

3. Small random (Type A) graphite flakes are conducive to higher tensile strengths than are small interdendritic groupings or large random flakes. This effect cannot be seen from the curves accompanying this article as the weakening effect of the small interdendritic flakes (Types D and E) obtained by rapid cooling is overbalanced by the effect of faster cooling in strengthening the matrix.

Ladle inoculation is proposed widely throughout the literature as a means of improving strength by

Fig. 8—Schematic comparison of the properties of cast iron and cast steel both having the same combined carbon content. The principal difference in microstructure is the presence of graphite flakes in the cast iron. A small part of the difference in most properties can be attributed to the difference in silicon content.



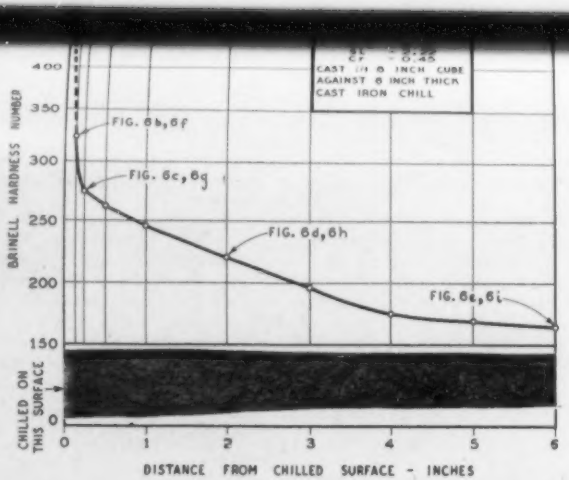


Fig. 6a—The effect of cooling rate upon the microstructure and hardness of a low alloy cast iron. The photograph illustrates the fracture appearance of a section cut from the center of the 6-in. cube. The white fracture next to the chilled surface is due to the formation of massive cementite. The coarsening of the fracture surface is due to the increase of graphite flake size as distance from chill increases.

Fig. 6b—Graphite distribution in alloyed cast iron 1/8-in. from chill. Fine interdendritic graphite; Rating 6E, 8D. Unetched, 100X.

Fig. 6c—Graphite distribution in alloyed cast iron 1/4-in. from chill. Small random flakes; Type A, Size 5. Unetched, 100X.

Fig. 6d—Graphite distribution in alloyed cast iron 2 in. from chill. Random graphite flakes; Type A, Size 4. Unetched, 100X.

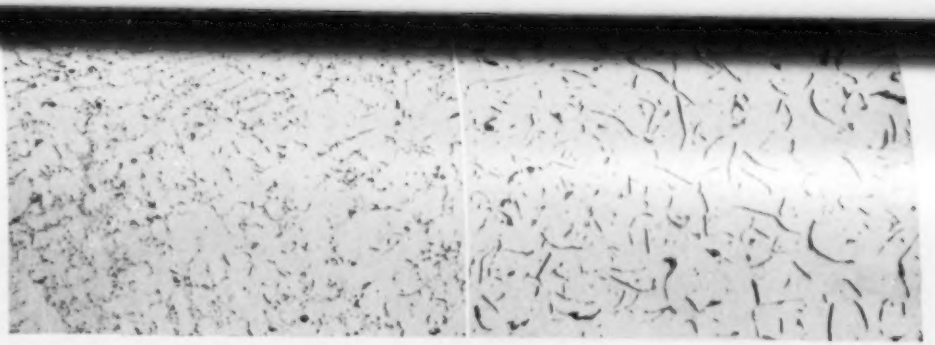
Fig. 6e—Graphite distribution in alloyed cast iron 6 in. from chill. Coarse random flakes; Type A, Size 3. Unetched, 100X.

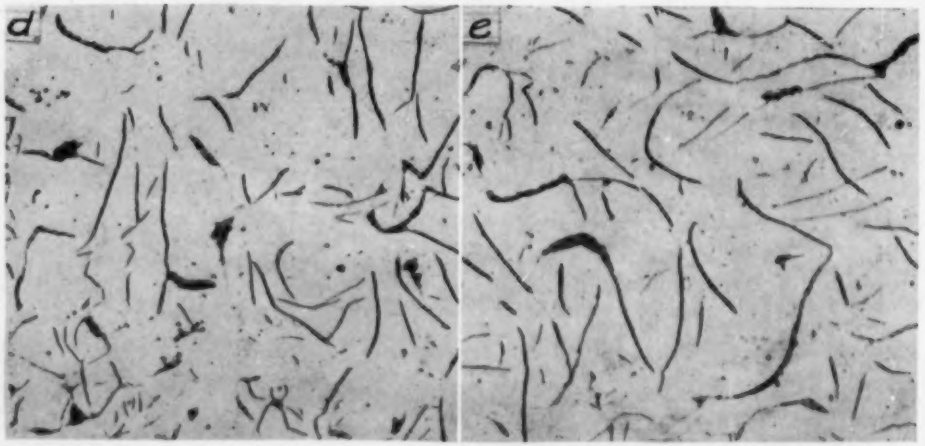
Fig. 6f—Alloyed cast iron 1/8-in. from chill. The microstructure consists of moderately fine pearlite and a considerable amount of massive cementite. Nital etched, 500X.

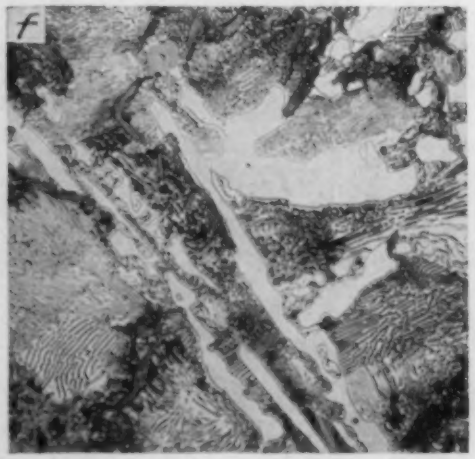
Fig. 6g—Alloyed cast iron 1/4-in. from chill. The microstructure consists of moderately fine pearlite and a small amount of massive cementite. Nital etched, 500X.

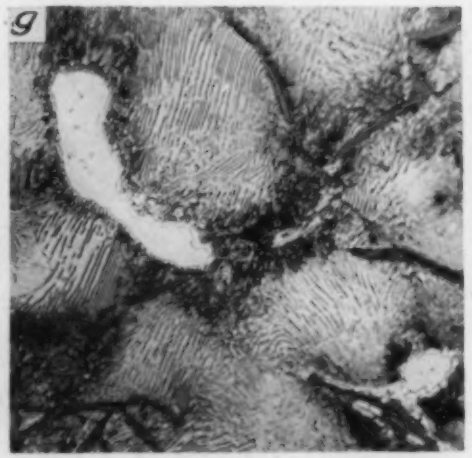
Fig. 6h—Alloyed cast iron 2 in. from chill. The microstructure consists of moderately fine pearlite and traces of massive ferrite. Nital etched, 500X.

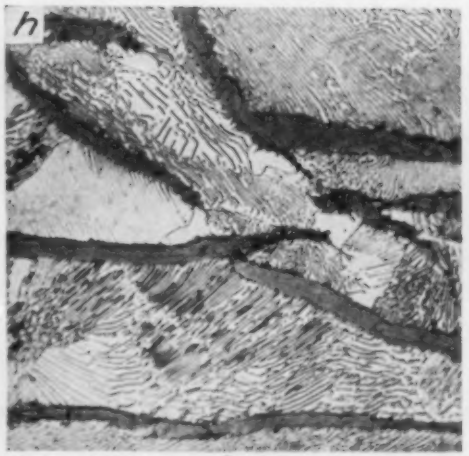
Fig. 6i—Alloyed cast iron 6 in. from chill. The matrix contains moderately coarse pearlite and considerable massive ferrite. Nital etched, 500X.

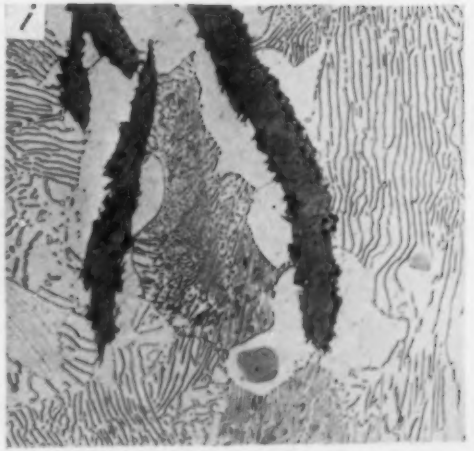


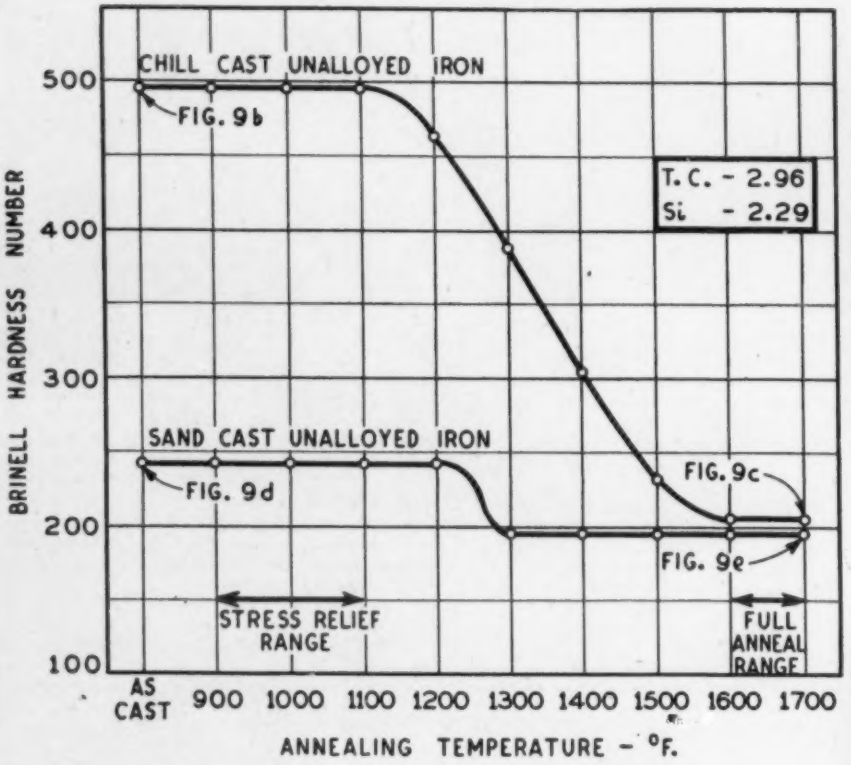


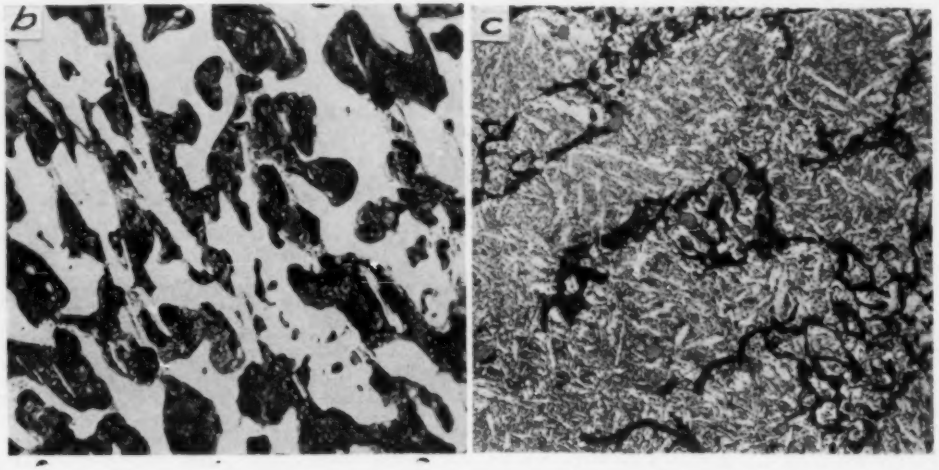












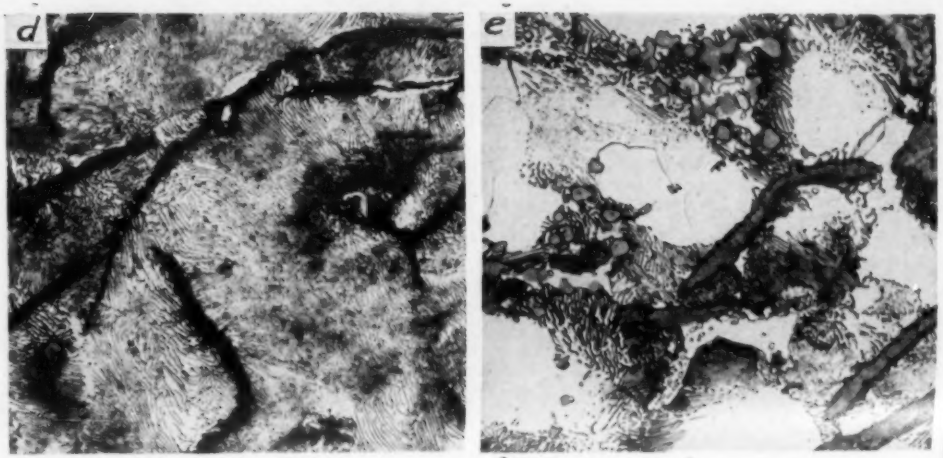


Fig. 9a—The effect of annealing upon the microstructure and properties of unalloyed cast irons. All samples were cast in 1.2-in. diameter bars, held at temperature one hour, and furnace cooled.

Fig. 9b—Chill cast unalloyed iron consisting of large masses of massive cementite in a background of fine unresolved pearlite. Nital etched, 500X.

Fig. 9c—Same iron as Fig. 9b after annealing at 1700° F. The microstructure consists of graphite flakes in a matrix of fine spheroidized pearlite. Nital etched, 500X.

Fig. 9d—Sand cast unalloyed iron consisting of fine pearlite and large graphite flakes. Nital etched, 500X.

Fig. 9e—Same iron as Fig. 9d after annealing at 1700° F. The microstructure consists of coarse graphite flakes in a matrix of massive ferrite and partially spheroidized pearlite. Nital etched, 500X.

reducing the tendency for formation of interdendritic graphite. The theory and effects of ladle inoculation have been summarized in a recent paper.[6]

Massive Cementite Effect. Massive cementite is to be avoided when an iron is to be machined. Although a uniform distribution of cementite particles throughout a matrix may have little effect upon hardness and cause only a small reduction in tensile strength, it may cause an iron to be virtually unmachinable by ordinary methods, or in milder cases to force reduced machining speeds and feeds. Conversely, massive cementite is desirable when wear resistance is the primary service requirement.

Pearlite Effect. As in steel, fine pearlite is conducive to relatively high strength and hardness; coarse pearlite to lower strength and hardness. Even very coarse pearlite, however, is considerably stronger than massive ferrite. All massive cementite-free, pearlitic cast irons are readily machinable. Irons containing coarse pearlite machine somewhat more readily than those containing fine pearlite.

Massive Ferrite Effect. Massive ferrite in cast iron has the same effects as in steel. It is a soft and weak constituent that is readily machinable, especially so in cast iron where the chips break up readily due to the effect of the graphite flakes.

Effect of Heat Treatment Upon Microstructure and Properties

Annealing Effect. Annealing below the critical range (for example, at 1000-1150° F.) has practically no effect upon the microstructure and hence little effect upon physical properties, but may be used to relieve stresses and reduce or eliminate distortion of finished parts.

Full annealing changes the microstructure considerably. Such a heat treatment above the critical range (for example, at 1500-1750° F.) breaks up massive cementite to form graphite and ferrite, thereby increasing machinability. Full annealing, consisting of slow cooling following heating at the annealing temperature, also will soften the pearlitic matrix, especially in castings having relatively thin sections or in chill-cast or permanent mold castings.

Therefore, such treatment will result in a considerable reduction in hardness and strength, but will increase machinability. Figure 9 illustrates the structural changes occurring during the annealing of two typical cast irons at various temperatures.

Quenching and Tempering Effect. Due to the similarity of the cast iron matrix to steel, the structure and properties of cast iron can be changed, as for steel, by quenching and tempering. The limitation imposed by the low ductility of cast iron is that complicated sections sometimes cannot be quenched because of the tendency to form quenching cracks.

Martensitic Matrix

Figure 10 shows the martensitic matrix developed by oil quenching and tempering of the iron shown in Fig. 2. By thus changing the microstructure, the tensile strength was increased from 36,000 to 53,000 psi. In a typical case, quenching was used to increase the hardness of a cast iron from 200 to 500 Bhn. Wear resistance in such cases increases greatly over that of the as-cast structure.

A high initial combined carbon is not necessary for the hardening of a cast iron by quenching as relatively rapid re-solution of carbon from the graphite flakes occurs at temperatures above the critical range. Irons composed entirely of graphite and ferrite may be hardened to high hardness by suitable treatment.

Acicular Structures. Flinn and Reese[7] have shown that suitable alloying of cast iron with nickel and molybdenum, followed by a low temperature heat treatment, will result in acicular structures such as that shown in Fig. 11. Such irons are reported to have tensile strengths up to 105,000 psi.

Summary

The effects of microstructure upon the tensile strength and hardness of cast iron can now be summarized by a closer study of Figs. 4, 5, 6, and 9.

In Figs. 4, 5, and 6, as the section size increases, massive cementite disappears, the pearlite becomes coarser, graphite flakes become larger and free ferrite appears in greater quantities. Each of these changes contributes to the progressive lowering of the tensile strength and to improved machinability.

However, with thin sections or the use of chills, a condition is reached where massive cementite begins to form. As the amount of massive cementite increases, tensile strength and machinability drop rapidly, whereas hardness increases. This formation of massive cementite, therefore, is responsible for the sudden drop in tensile strength occurring for the thin sections illustrated in Figs. 4 and 5.

Lowering of the total carbon and silicon contents of the metal decreases the graphitizing tendency of the iron and increases the stability of the cementite (including the cementite in the pearlite). Such a

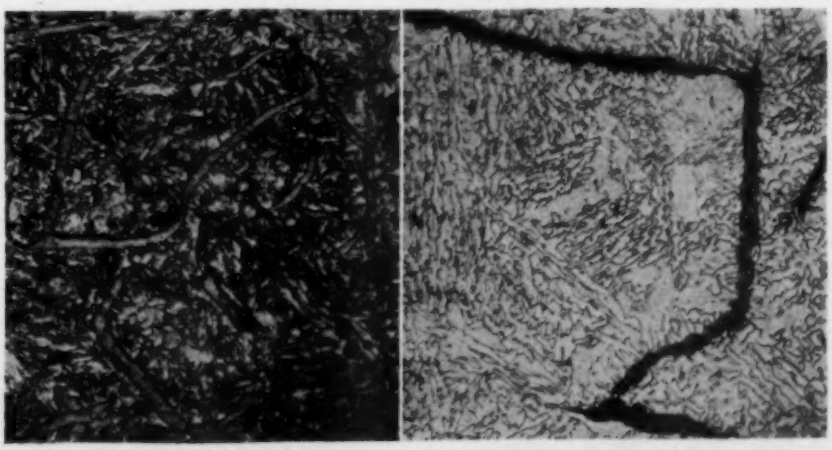


Fig. 10 (Left)—Same iron as that shown in Fig. 2 after oil-quenching from 1500° F. and tempering at 900° F. The microstructure consists of graphite flakes in a matrix of tempered martensite. Nital etched, 500X. Fig. 11 (Right)—Acicular structure formed by certain nickel-molybdenum alloy iron compositions. When tempered at low temperatures this type of microstructure is responsible for very high tensile strengths of the order of 60,000 to 100,000 psi. Nital etched, 500X.

change in analysis, therefore, tends to restrict coarse pearlite, large graphite flakes, and the formation of massive ferrite, and hence results in higher tensile strength and hardness, especially in the thicker sections. A comparison of Figs. 4 and 5 shows this to be true.

Full Annealing

Figure 9 illustrates the fact that full annealing of a cast iron is to be avoided when high hardness and high strength are desired. Full annealing will eliminate massive cementite, however, and will promote the presence of ferrite, so that machinability is increased.

Stress relief annealing results in substantially no change in microstructure and does not adversely affect tensile strength or hardness. Therefore, such treatment generally will be beneficial to any engineering application, especially so in the case of large castings or those with complex combinations of thin and thick sections.

As in the case of steel, the microstructure and attendant properties of cast iron may be changed by the use of quenching and tempering treatments and by the utilization of alloying elements. Quenching and tempering result in improved strength and hardness through formation of martensite, bainite, etc., in the "steel" matrix.

Similarly, alloying elements may be used to develop microstructures of increased strength and hardness and improved wear resistance. Some of the alloying effects may be attributed to the control of graphitization; others may be caused by effects on the matrix.

Considered from the standpoint of casting design, a cast iron for a given application should be selected with consideration to the microconstituents which make up the iron. Section thicknesses should be uniform, or a variation in properties in the different parts of a casting will be obtained.

If machinability and strength are desired in light sections, the type of iron used is especially important. Very light sections must be carefully considered in the metallurgy of the iron used in order to prevent occurrence of massive cementite and resultant difficulties with machining. Annealing will help but leads to low strength. If strength is desired in heavy sections, the analysis of the iron must be selected to give the desired properties.

Because of the variation in microstructure with cooling rate, it is important to notice that a tensile strength determination on a test bar does not indicate the strength of the iron in all sections of a casting. A test bar tensile strength corresponds only to the tensile strength of similar section thicknesses in the casting; sections in the casting that are heavier than the test bar will have lower strength and lower hardness than the test bar because of the effect of microstructure as noted previously.

Likewise, a test bar tested in the as-cast condition will be of little value in estimating the strength or hardness of a casting that subsequently has been given a full anneal, even though the test bar is of a section thickness similar to that of the casting.

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DISCUSSION

Chairman: T. E. EAGAN, Cooper-Bessemer Corp., Grove City, Pa.

J. H. SCHAUM[1]: I would like to ask Mr. Mahin where the tensile specimens were located. It is important to know whether you pulled a 6-in. diameter specimen or a smaller specimen from the center or the edge of the test bars.

MR. MAHIN: They were taken from the center.

J. R. GOLDSMITH[2]: Was it a machined test bar such as a 0.505-in. bar? Was the hardness that you recorded taken on the test bars?

MR. LOWNIE: All test specimens were machined from the center of the test bars and all had threaded ends.

For bars cast to ⅞-in. diameter, ASTM Tension Test Specimen A (0.505-in. diameter in the reduced section) was used. For 1.2-in. bars, ASTM Specimen B (0.750-in. diameter) was used. For the 2, 3½, and 6-in. diameter bars, Specimen C (125-in. diameter) was used. For bars cast to ½-in. diameter, a modification of Specimen A was used but the standard proportions were retained.

Brinell Hardness measurements were made at the center of the tension test specimens.

R. W. LINDSAY[3] *(written discussion)*: It was interesting to note the comparison between Figs. 9b and 9c. The graphite shown in Fig. 9c would be classed as flake-like in nature. Now, the more likely expectation would be that the graphite resulting from annealing at 1700° F. of iron of the structure shown in Fig. 9b would be nodular or temper carbon-like in character.

It was mentioned that the structure in Fig. 9c resulted from the annealing of an iron already containing some flake graphite in the structure, hence an iron that would be classified as a mottled iron rather than a white iron. This would account for the graphite resulting from "breakdown" of the carbide at this temperature appearing as flakes rather than temper carbon nodules. It might be better to show in Fig. 9b a structure including some of the flake graphite in the "as cast" iron.

Annealing of cast iron of completely white structure such as shown in Fig. 9b would lead to development of temper carbon, the form of graphite commonly associated with malleable iron. Thus, if the authors removed a piece from the pure chilled iron part of their sample, so that it would be free of any flake graphite that formed during or shortly after solidification, and annealed this at 1700° F., the graphite would be more like temper carbon in appearance.

It is true that annealing completely white iron at temperatures higher than 1700° F., say 1800° or 1900° F., would produce a flake-like graphite. The approach to a well defiined flake even at these annealing temperatures is governed by the size of specimen and the rate at which they heat through the lower range of temperature (1500° to 1700° F.). Slow heating through this range favors graphite formation that is more like temper carbon.

In closing, it might be emphasized that I have taken exception to the authors' Figs. 9b and 9c on the basis of expected structure development rather than on the grounds of practical results. The undesirable consequences of chilled corners or edges on supposedly gray iron castings would be corrected in any case as long as massive or free iron carbide is eliminated from the structure.

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CORE SAND PURCHASING FACTORS

H. Louette
A. E. Murton
and
H. H. Fairfield

TWO SANDS of widely different properties were investigated in the course of testing core sands for foundry use. Since the test data indicated some of the important properties which should be considered when purchasing core sand, it was thought that the results of the investigation might be of general interest.

The purpose of the investigation was to determine which of two sands was best suited for the production of radiator cores. Figure 1 shows the type of core made. This core is made on a bench. One half is rammed in a core box, and the other half is rammed in a drier. The two halves are "booked" together and the corebox is removed. The drier containing the completed core is placed on the oven rack.

Laboratory Tests on Sand Samples "As Received." The two sands were compared by mixing them with 1 per cent of bentonite, 1 per cent of cereal flour, and 1 per cent of core oil. The optimum water content for each sand was determined by test. The results obtained are shown in Table 1. Screen analyses of the two sands are given in Table 2 and Fig. 2.

Foundry Tests on Samples "As Received." The foundry test consisted of making 64 radiator cores with each type of sand. Details of this test are given in Table 3.

Preliminary Conclusions. From the laboratory tests it was evident that sand No. 1 would give baked tensile strengths three times as great as sand No. 2 with similar core sand mixtures. The foundry tests were not conclusive because the mixtures tested had different baked transverse strengths. Further test work was considered advisable in order to determine why No. 2 sand seemed to be a "better working" sand.

Laboratory Tests on Treated Samples. At this stage of the investigation it was thought that the difference in baked strength between the two sands was due to the difference in grain size distribution (Fig. 2). In order to test this idea, sand No. 2 was carefully screened to the same grain size distribution as sand No. 1. Tests were carried out on the treated sand, and the results are shown in Table 4.

Shape of Sand Grains. Photomicrographs of the two types of sand are shown in Figs. 3 and 4. It will be noted that sand No. 1 had a smooth, rounded grain, and sand No. 2 a rough, subangular to angular grain. These grains are referred to as "rounded" and "angular" in this paper.

Foundry Tests on No. 1 Sand at Different Oil Ratios. A series of

Specification of grain size only in purchasing core sand is not a sufficient precaution to ensure that a suitable sand will be obtained. The pronounced effect of grain shape upon many of the properties of core sand mixtures is shown by tests. The rounded sand grain has many advantages over the angularly shaped grain, including lower core oil consumption, easier coremaking operations, and less gas and smoke in the foundry.

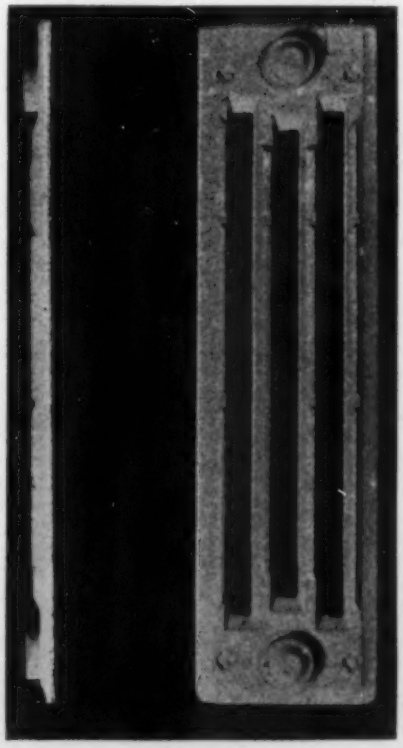


Fig. 1—Core used for radiator casting.

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†Metallurgical Engineers, Bureau of Mines, Ottawa, Ont., Canada.

Presented at a Sand Shop Course Session of the Fiftieth Annual Meeting, American Foundrymen's Association, at Cleveland, May 9, 1946.

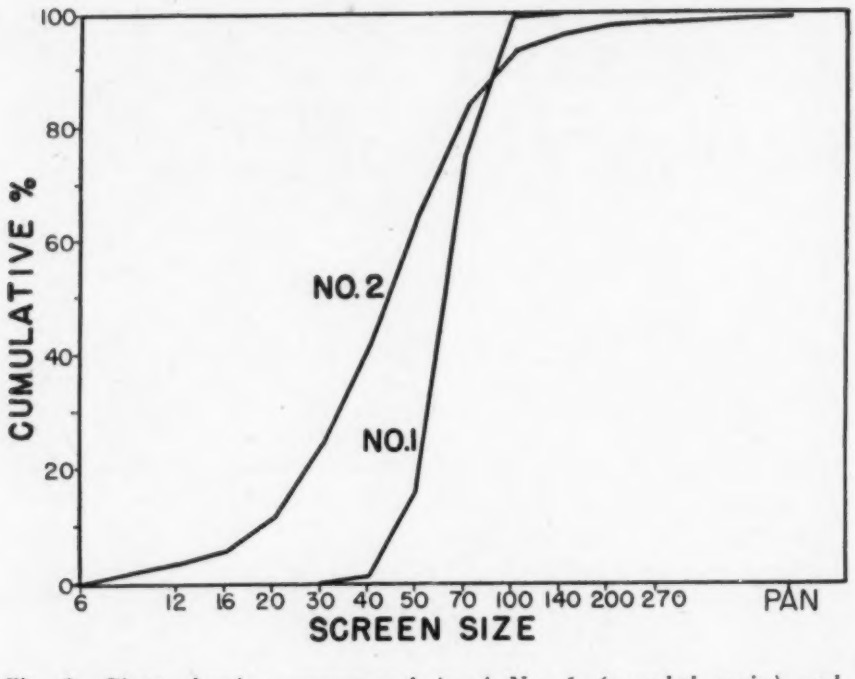


Fig. 2—Chart showing screen analysis of No. 1 (rounded grain) and No. 2 (angular grain) sands.

Table 1

COMPARISON OF CORE SANDS

	No. 1 Sand	*No. 2 Sand*
Moisture required to temper, %	1.3	3.0
Green permeability	156	173
Green bond, psi.	2.6	2.6
Deformation, in. per in.	0.0175	0.0155
Flowability, %	88.5	84.0
Baked permeability	178	235
Baked tensile strength, psi.	164.5	54
Core hardness	55	38
Hot strength at 2200° F., psi.	10	14
Hot strength at 2300° F., psi.	8	4
Hot strength at 2400° F., psi.	7	3

Table 2

GRAIN SIZE DISTRIBUTION

U. S. Screen No.	*Retained, %*		*Cumulative, %*	
	No. 1 Sand	*No. 2 Sand*	*No. 1 Sand*	*No. 2 Sand*
6	0.0	0.2	0.0	0.2
12	0.0	3.2	0.0	3.4
16	0.0	2.4	0.0	5.8
20	0.05	5.9	0.05	11.7
30	0.05	12.8	0.1	24.5
40	1.2	17.9	1.3	42.4
50	14.9	22.1	16.2	64.5
70	59.2	18.7	75.4	83.2
100	24.0	9.6	99.4	92.8
140	0.4	3.0	99.8	95.8
200	trace	1.3	99.8	97.2
270	trace	0.6	99.8	97.8
Pan	trace	1.0	99.9	98.8
A.F.A. Clay	0.1	1.0	100.0	99.8
A.F.A. Fineness No.	49	43		

sand mixtures using sand No. 1 were prepared. The relationship of the core oil content to the baked transverse strength is shown in Table 5.

Cores made with a 170:1 sand-oil ratio gave excellent results in the coreroom and in the foundry. The earlier difficulty experienced with sand No. 1 was attributed to the fact that it was used in mixtures with a higher transverse strength than was desirable. In order to check the effect of transverse strength upon results in the foundry, the following tests were carried out. In these tests (Table 6) the No. 2 sand was screened so as to duplicate the grain size distribution of No. 1 sand.

Up to this time the core sand mixture used for radiator cores had consisted of No. 2 (angular) sand mixed with core oil in a ratio of 60:1. It was decided to conduct a trial run with No. 1 (rounded) sand mixed with core oil in a ratio of 170:1. After the experiment had been carried out for a three-week period, during which time 16,000 castings were made, the general conclusions were as follows:

(a) In comparison to cores previously used, the 170:1 core sand mixture produced much less gas. With the 60:1 core sand mixture, gas from the core vents spurted out into a pressure jet when the mold was poured. With the 170:1 mixture, only a small, lazy flame was noticeable at the vent after pouring. Workers noted the improvement in the foundry atmosphere.

(b) Coremakers reported that the low oil content core sand mixture resulted in cleaner coreboxes, easier coremaking, and less sticking

(c) In the casting cleaning operation, it was noted that the core sand ran freely out of the casting.

(d) The test core mixture gave a smoother casting finish.

(e) The low oil content core mixture required slightly closer control in baking.

Effect of Grain Shape. The work of Davies and Rees* showed that sand grain shape had a marked ef-

*W. Davies, and W. J. Rees, Paper No. 8/1944, Steel Castings Research Committee, Iron and Steel Institute; and "The Effect of Grain Shape on the Moulding Properties of Synthetic Moulding Sands," *Refractories Journal*, March, 1945, p. 98.

Table 3

TESTS OF SANDS "AS RECEIVED"

	Sand No. 1	*Sand No. 2*
Ratio of sand to oil (by weight)	81:1	59:1
Moisture, %	3.4	4.5
Baked transverse strength, lb.	45	34
Number of times core box needed cleaning	10	2
Time to make 64 cores, min	90	55
Coremakers' reports	"sticks to box" "hard to pack"	"better working sand"
Cores broken in removing from driers	none	5
Cores broken in cleaning and handling	3	1

fect upon molding and coremaking properties of sand mixtures. It was concluded that the difference observed between No. 1 (rounded) and No. 2 (angular) sand was due to the difference in sand grain shape.

In order to verify this conclusion, two sands were selected for practical test in the coreroom without consideration as to the grain fineness. The angular and the rounded sands were selected by microscopic observation (Figs. 5 and 6). These specimens were designated as No. 3 (rounded) and No. 4 (angular).

It was apparent from these tests that successful results in the coreroom were obtained with core mixtures which would develop a transverse strength of 25 to 35 lb. when baked (corresponding baked tensile strength is approximately 100 to 175 psi.).

High Baked Strength

Core mixtures with higher baked strengths gave trouble in the coreroom, both in ramming up cores and in removing cores from driers.

The difference between the foundry performances of No. 1 (rounded) and No. 2 (angular) sands is attributed to three conditions. First, the No. 2 sand had clay or other materials adhering to the grains. This was shown by the improvement attained by washing. Secondly, the grain size distribution of No. 2 sand was not as good as that of No. 1 sand. When the fine material was eliminated by screening, No. 2 sand showed better results. *The most important difference between the two sands, however, was that of grain shape.*

Grain Contact

Rounded grains pack together more easily than do angular grains. This enables the oil-coated grains to have better contact with each other. Better contact means higher baked tensile strength.

Rounded grains have a smaller surface area than do angular grains. The smaller surface area results in a thicker coating of oil on each grain; therefore, a higher baked

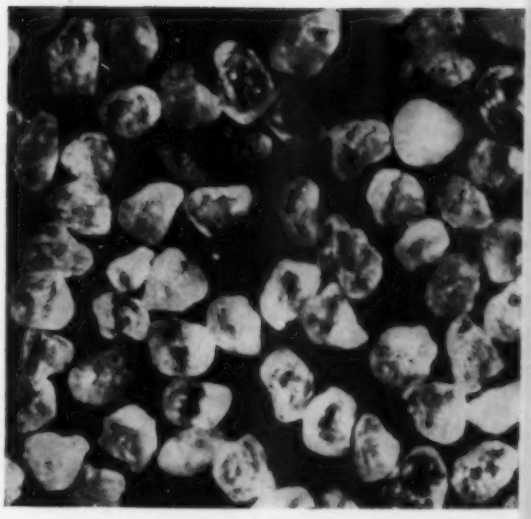


Figure 3
Photomicrograph of No. 1 sand showing smooth, rounded grain. 25X.

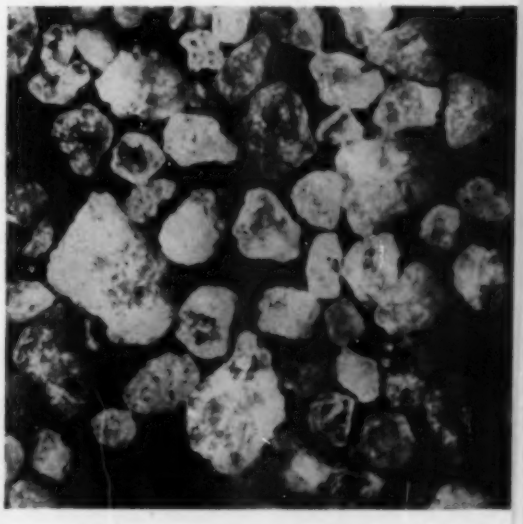


Figure 4
Sand No. 2 has rough, subangular to angular grain. 25X.

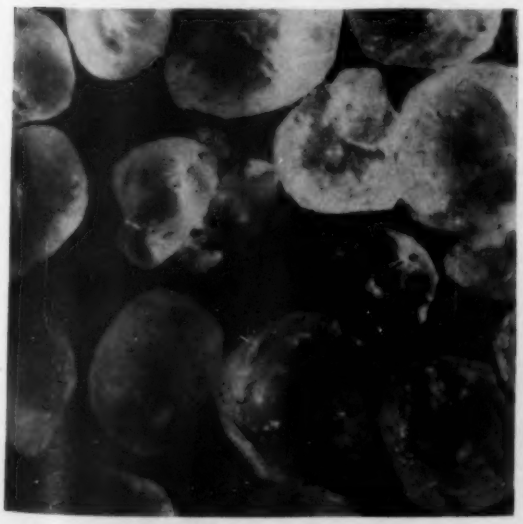
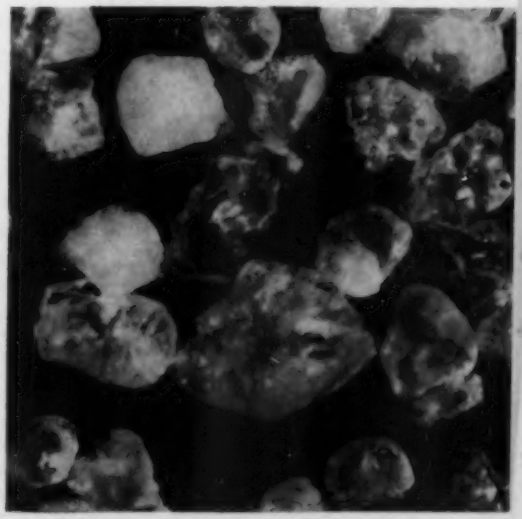


Figure 5
Left—No. 3 sand, rounded grain. 25X.

Right—No. 4 sand, angular grain. 25X.



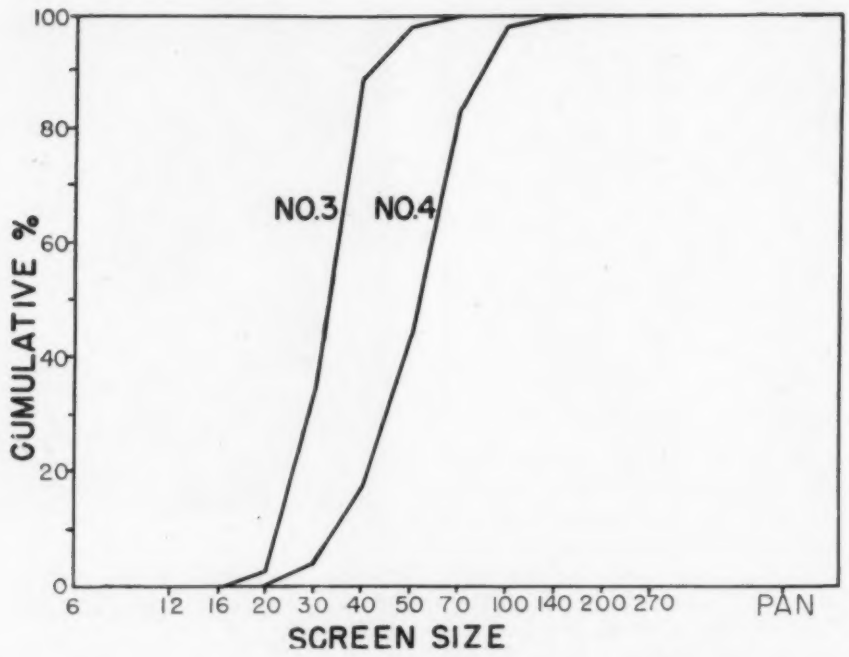


Fig. 6—Chart showing screen analysis of No. 3 (rounded grain) and No. 4 (angular grain) sands.

Table 4

TENSILE STRENGTH, SANDS NO. 1 AND 2

Description of Sample	*Baked Tensile Strength, psi.*
No. 1, as received	165
No. 2, as received	54
No. 2, washed only	82
No. 2, washed and screened to same grain size distribution as No. 1	107

Table 5

EFFECT OF OIL CONTENT ON TRANSVERSE STRENGTH OF NO. 1 SAND

Sand to Core Oil Ratio	*Baked Transverse Strength, lb.*
57:1	54
81:1	47
114:1	34
170:1	26

Table 6

TRANSVERSE STRENGTH TESTS

Sand to Core Oil Ratio	*Sand Type*	*Transverse Strength, lb.*	*Results in Coreroom and Foundry*
170:1	No. 1 (rounded)	23	Good.
170:1	No. 2 (angular)	None	All cores broke when removed from driers.
87:1	No. 1 (rounded)	46	Considerable difficulty in removing cores from driers.
87:1	No. 2 (angular)	20	Fair.

Table 7

SAND GRAIN SHAPE AND CASTING FINISH

	Core Sand	
	No. 3 (rounded)	*No. 4 (angular)*
Core sand to core oil ratio	170:1	170:1
Core hardness	65 to 70	nil
Transverse strength, lb.	24	8.5
A.F.A. Grain Fineness No.	27.5	47.5
Results in coreroom	Excellent	All cores broke when rapping out of driers
Finish on casting	Smooth	

strength is obtained with rounded-grain sand.

The areas of contact are larger with rounded grains than with sharp cornered angular grains. This also results in higher baked strength with rounded-grain sand.

Tests carried out with No. 3 and No. 4 sand served to verify the conclusions concerning the effect of grain shape. It was of considerable interest to note that the sand with an A.F.A. Fineness of 27.5 gave a smooth casting, which could not be distinguished from one made with No. 1 and No. 2 sand (A.F.A. Fineness No. 40 to 50). It is apparent that the shape of the sand grains has considerable effect upon casting finish. (Table 7).

Conclusions

Previous to this project, when core sands were examined for foundry use the properties considered were:

1. Refractoriness.
2. A.F.A Grain Fineness.
3. Amount of clay, dirt, and foreign matter.
4. Grain size and distribution.

As a result of the tests reported in this paper, it was concluded that the shape of the sand grain has a profound influence upon the properties of core sand mixtures. Therefore, preference should be given to a sand with rounded, smooth-surfaced grains. A low-power microscope is a useful tool in estimating the suitability of core sands.

The authors are indebted to E. N. Delahunt, general superintendent, and E. Gervais, sand control supervisor, Warden King, Ltd., for their valuable cooperation in writing this paper. Warden King, Ltd., also wishes to express appreciation of the work carried out by the Physical Metallurgical Research Laboratories of the Bureau of Mines, Ottawa, Canada.

MECHANICAL SHAKEOUT

Design, general layout, sand handling, dust control, costs, and general hygiene are some of the foundry operating factors directly affected by the shakeout installations.

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SHAKEOUT IS A TERM which, to the average foundryman usually has many particular translations: (1) removing castings from the flasks and separating them from the molding sand; (2) means employed in accomplishing the removal and separation (manual, semimechanical or completely mechanical); (3) the most difficult foundry job to man in the prevailing labor market; (4) the source of the majority of foundry dust, which is rapidly becoming the progressive foundryman's foremost problem because of its effect on general foundry working conditions, and the advent of more stringent state laws governing foundry sanitation and hygiene.

Manual Shakeout

Manual shakeout has been practiced for years and there are many places where it still is the most practical way of freeing a casting from the flask; however, it does not afford much chance of controlling the resulting dust, and poses a further problem of manhandling the sand either to clear the floor or prepare the sand for reuse.

Semimechanical shakeout, such as holding a flask in an elevated position and using a vibrator, is a somewhat faster operation for a given size of flask but often a more dusty operation than straight manhandling. Wear and tear on flasks and crane equipment is high and adequate dust control well-nigh impossible. The sand pile occasioned by shakeout has to be handled by hand or clamshell unless some special conveyor system is available. Sand handling, especially from dry sand mold shakeout, by either manhandling or clamshell bucket, further aggravates the dust control problem.

Presented at a Plant Equipment Session of the Fiftieth Annual Meeting, American Foundrymen's Association, at Cleveland, May 6, 1946.

Mechanical Shakeout

Mechanical shakeout, as the author understands it and further discusses it in this paper, is a device, usually a vibrating deck, upon which flasks are placed and the sand vibrated free from the flask and the casting and deposited in boxes or on a conveyor after passing through openings in the vibrating deck.

A mechanical shakeout system, to the author, is a combination of the mechanical shakeout device itself; the foundation upon which it rests; the dust control system necessary to control dust at the point where made; the conveying apparatus handling the sand shaken out and either storing it for future use or conveying it to other apparatus for reconditioning or refuse; and such other devices as are necessary to adequately feed the shakeout (cranes, etc.) and remove castings and flasks from the mechanical shakeout deck.

The design of a mechanical shakeout system presents five problems:

1. The shakeout device itself.
2. Foundations supporting the shakeout as well as other parts of the system.
3. Dust control apparatus.
4. Equipment for handling the sand from shakeout.
5. Crane, conveyor or other equipment to bring flasks to the shakeout and to take the castings and flasks away after shakeout operation.

Each of these problems has a definite relation to a particular foundry layout and, usually, a change of one element—size, shape, location, or otherwise—affects the other elements of the system.

Assuming that the majority of foundrymen are interested in the application of mechanical shakeout systems to existing foundries, the author attacks the problem from that angle. Treatment of the subject of mechanical shakeout from the standpoint of the design of a new foundry would of necessity be different if one expected to benefit from the improvements made in foundry practice in the last decade.

Shakeout Systems Vary

No one design of a mechanical shakeout system will fit every requirement of existing foundries. Many foundries—such as jobbing foundries—make a wide size variety of castings and require more than one design of mechanical shakeout system.

As to the mechanical shakeout device itself, there are two principal types—each of them much alike and differing principally in the manner in which they are motivated for deck vibration. One type has a straight mechanical drive through an eccentric (Fig. 1) and the other obtains its motion through an unbalanced weight acting on a spring support.

Both types are widely accepted

and both do an excellent job when properly applied. They are available in many sizes, from as small as 2x3 ft. to as large as 8x10 ft., and when used in multiples they are limited only by the means of getting rid of the sand after it is freed from the flask.

The design of the shakeout device itself has been well worked out by the equipment manufacturers. The foundrymen need not be concerned about mechanical design, but he must be careful to select a size in keeping with his requirement and his pocketbook because the device itself usually represents but a small part of the cost of a mechanical shakeout system.

Grate Openings

In addition to the matter of shakeout size, there is the matter of size of openings in the shakeout grate. There are two schools of thought on this matter, one of which insists upon large openings and passes the work of breaking the lumps to other parts of the equipment, and the other which advocates the smallest practical opening.

With large openings the shakeout deck clears rapidly but usually passes gaggers, tramp metal, sprues, gates, etc., which must be separated from the sand at some other part of the system. Usually, the larger the flask the larger the lumps, especially in dry sand work. The number of gaggers likewise increases, quite often out of proportion to the increase in flask size.

By the same token, large flasks are not handled as rapidly as smaller flasks, and the time required for one flask through the shakeout operation ordinarily is sufficient to allow the sand to flow through small holes. Sand lumps not passing through the small openings are crushed when the next flask is placed on the shakeout.

On the basis of personal observation and experience, the author leans to the smallest practical opening. With a maximum opening 1¼x 4 in. on a vibrating deck 12x33 ft. with a shakeout area of 396 sq. ft., and handling lumps of dried sand 8x10x6 in. that will support an average man's weight, it has been found that practically all of the lumps will break up and go through the deck by the time the crane crew can remove and land the flask, pick up and land the casting and deposit another flask on the shakeout. Any lumps that remain are quickly crushed when the next flask is landed on the shakeout deck.

Green Sand

On shakeout decks 36x42 in., it has been found that green sand readily passes through openings ¾x3 in. On large shakeouts, say 5x6 ft., openings of 1x3 in. have proved quite satisfactory.

In handling small, dry sand flask shakeouts, again with reasonably hard lumps 3x3 in. openings are satisfactory; as a matter of fact, the original openings have been reduced from 3x3 to 1¾x3 in. It has been found that fewer gaggers pass through the smaller openings, with correspondingly less trouble in the conveying system and less wear and tear on conveyor belts.

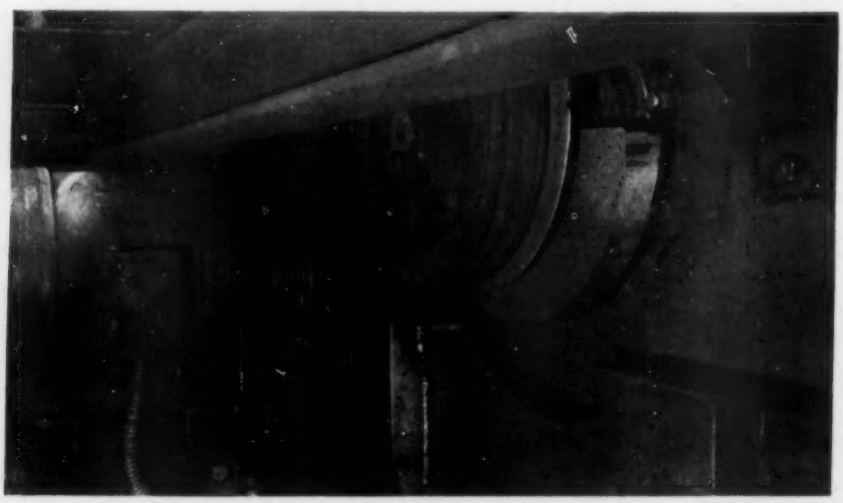


Fig. 1—Twenty hp. multiple "V" belt drive for one of the 6x10-ft. sections of the shakeout. Motivating counterweight and spring supports in the center.

Foundations

Foundations for shakeout devices, especially those shakeouts intended to handle extremely large flasks, i.e., flasks 10x30 ft. and weighing upwards of 50 tons, require careful consideration of many factors:

1. Location of the shakeout with relation to building supports.
2. Character of the ground on which the foundation is to rest.
3. Piping and waterproofing.
4. Proximity of the shakeout to adjacent buildings or property owned by others.
5. Shakeout location with reference to chemical or metallurgical laboratories where sensitive balances are used.
6. Location with respect to machine tools that turn metal, either by cutting or grinding.
7. The usual considerations for the location of heavy equipment in relation to crane facility, material flow, chipping room, sand storage and proper placement to suit general foundry layout.

In general the foundation is found to be appreciably larger than the shakeout deck area, and often is supported on steel which spans the foundation walls in order to provide room for the hoppers receiving the sand as it cascades through the shakeout deck, room for the feeders and conveyors taking the sand from the hoppers, and working room for the operator who must grease the conveying apparatus and tend such other devices as are in the pit. Furthermore, a means of drainage should be provided because some water will get into the pit in spite of all preventive measures.

Pit Construction

On large shakeout installations, the pit over which the shakeout device is mounted is constructed with at least 6 ft. of headroom and, when allowance is made for pit floor and pit roof, the excavation is down 8 ft. or more below the foundry floor level. This depth is greater than that of the average crane column footing, and quite often means underpinning the building column or making the pit wall heavy enough to take the load due to the column footing; therefore, as previously

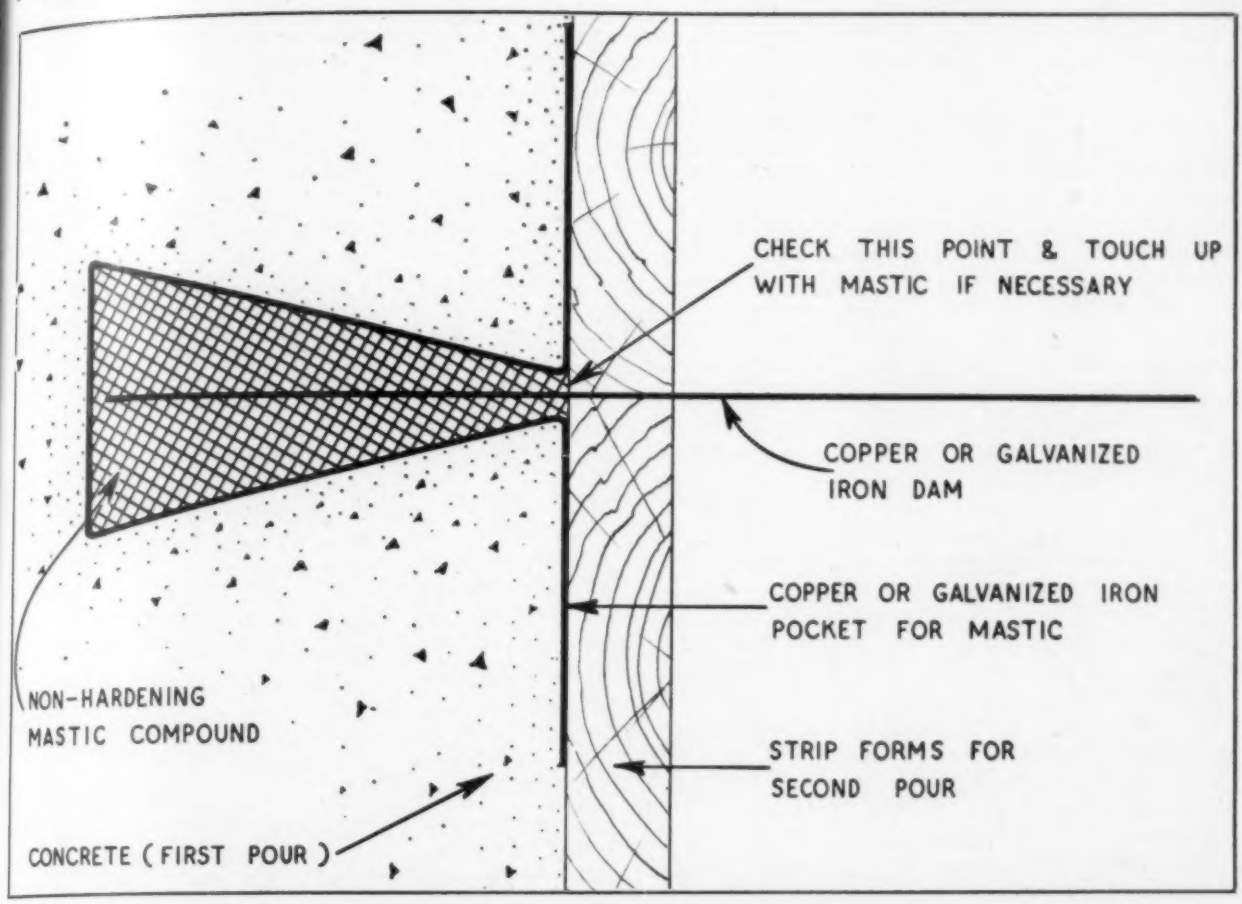


Fig. 2—Sketch showing method of making vertical construction joints in concrete pit of shakeout installation.

pointed out, it is important to consider shakeout location with reference to building columns or supports.

Consideration of ground or soil characteristics is most difficult, unless one is familiar with them through previous foundation experience in the immediate area. Any condition from hardpan to quicksand may be encountered. However, good foundations can be placed in any sort of ground, provided that precautions are taken to make the foundation design suit prevailing conditions.

All vibrating shakeout devices have shaking forces, some more so than others, and there are times when the minimum shaking force to be absorbed by the foundation is an important consideration in shakeout device selection from the standpoint of foundation cost.

The location of underground piping is important because it is costly to reroute piping of any appreciable size. Storm sewers and other drainage lines disturbed during excavation often cause no end of trouble by leaking during pit construction, or afterward by seepage through construction joints.

Extreme vigilance on the part of the owner or his agent is required during pouring of concrete if the pits for shakeout installations are to be made watertight. Short cuts taken by contractors often result in water seepage into pits 6 months to a year after a shakeout is put into operation.

The practice of placing metal dams (flat pieces of sheet copper or galvanized iron 6 in. wide) at all horizontal construction joints so that they are half width in the first pour and half width in the next pour is well worth while. Vertical joints require careful treatment, and the metal dam shown in Fig. 2 has worked out well. The best possible waterproofing is none too good when shakeout foundations are placed in normally wet areas.

Next in importance is the location of the shakeout with regard to adjacent buildings, especially dwellings. Shakeouts do make some noise and there are times when it is objectionable, especially if prevailing winds are such as carry the noise to adjacent dwellings. This is especially true if the shakeout works at night when the usual factory or foundry noises are at a low ebb.

Shaking forces which may not be noticeable immediately around the shakeout may cause vibration at distances of 100 ft. or more from the shakeout location and, while they may be of a magnitude difficult to measure, they are sufficient to rattle dishes and other objects and create unpleasant situations for all concerned.

Chemical and metallurgical laboratories with their delicate "balances" and other instruments and chemical glassware are readily disturbed and sometimes made inoperative due to vibration transmitted through the ground, and it is well to keep large shakeouts at a reasonable distance from them. This is especially true in areas where ground conditions are poor.

Experience has shown that

ground-borne vibration of small magnitude will seriously affect the finish on parts that are either ground or turned. There have been cases where machine tools have had to be insulated by non-vibration-transmitting or vibration-absorbing materials in order to prevent foundry-made vibration from affecting the finish on work being machined.

Location of the shakeout determines the location of the foundation, but there are other considerations such as crane facility, material flow, relative chipping room location, sand storage, and proper placement to suit the general foundry layout, and these will be discussed later.

The availability of a crane over the foundation will expedite the excavation, placing of forms and pouring of concrete as well as setting up the equipment which rests in and on top of the foundation.

Refuse foundy sand makes excellent fill around the foundation and can be placed after the concrete has set. If the foundations are outside in the weather and set on clay, it is well to set a tile drain around the foundation in a manner similar to that used in dwelling basement construction due to the fact that, for at least a few years after building, surface water from a considerable area will accumulate in the broken ground around the foundation.

The design of the foundation itself has to suit so many variables that it is difficult to set down any hard and fast rules. In general, foundations are constructed of reinforced concrete, often in conjunction with rolled sections ("I" beams, channels, etc.) set to suit the superimposed equipment. They must be carried down to a solid footing or the equivalent. In all cases, it is well to isolate shakeout foundations from building or crane column footings, building walls, and foundations for other equipment. In special cases, it is well to consider vibration-absorption materials of the rubberized fabrics, rubber or steel spring type, in order to prevent transmission of vibration.

Shakeout Dust Control Apparatus

One of the most important reasons for installing mechanical shakeout is that it affords an opportunity for better control of the resultant dust. Many variables enter the picture and each shakeout installation presents an individual problem.

Shakeout dust usually leaves the flask at a temperature higher than that of the average temperature for the building in which it is located; therefore, it has a tendency to rise and to follow any air movement in its immediate vicinity. In many shakeout operations, steam rises with the dust and further aggravates the problem.

The tendency for dust to rise suggests overhead dust pickup. Unfortunately, overhead space just above the shakeout grate is at premium because it is in the area of flask movement. In general, foundrymen prefer to have no obstruction in the way of flask handling, particularly when overhead cranes are used.

Therefore, it is necessary to compromise between ease of casting handling and the difficulty of handling shakeout dust in directions other than natural flow. Repeated tests have proved that the best dust control is accomplished by arresting and collecting it close to the "point where made."

There are three generally accepted basic methods of dust control:

1. Overhead hoods or partial enclosures.
2. Down draft.
3. Side hoods.

A fourth method, complete enclosure, with openings only large enough to pass the required control air, is being tried in connection with what is thought to be the world's largest completely live-deck shakeout, and it may be that this method will set a pattern for the larger shakeouts.

The "overhead hood" is limited

Fig. 3 (left)—Close-up of 28x9-ft. flask on shakeout during vibrating period. Roof closed. Entire shakeout deck in motion.

Fig. 4 (center)—Close-up of 72-in. diameter by 40-in. deep cope being vibrated. Roof closed. One-third of shakeout deck in motion.

Fig. 5 (right)—Hot mold being transported to flask yard for cooling prior to shakeout.







to smaller shakeouts where the flasks can be upset by hand, or by mechanical, hydraulic or pneumatic devices. It can be made to suit a wide variety of conditions as it is a modification of the complete enclosure. The amount of air (cu. ft. per min.) required to adequately exhaust this type of hood can be figured quite accurately by measuring the area open to indraft in sq. ft. and multiplying by 200. A double-check on this figure can be made by taking the area of the shakeout grate and multiplying by 200. It is advisable to use whichever figure is the greater. The exhaust connection should be at the top of the hood, preferably at the apex of sloping top plates.

Dust Control Systems

Downdraft, as the name implies, is a method of drawing dust, steam and fumes through the shakeout grate by means of exhaust connections in the shakeout hopper. This type of shakeout dust control appeals to the foundry operator because it requires no hood above the floor level and gives the shakeout crew and crane operator unobstructed access to the shakeout grate. Occasionally, downdraft systems are successful, but these installations are the exception rather than the rule.

Some of the reasons for difficulties with downdraft shakeout dust control systems are:

1. The natural path of dust and gas travel is upward, directly opposite to the direction of control air travel; this is particularly true when castings are shaken out hot.

2. The maximum concentration of dust usually is well above the level of the grate and much of the control (exhaust) air bypasses from the floor through the grate with little or no effect in the area of maximum dust concentration.

3. Temporary coverage of the entire grate with sand may completely shut off the flow of control air at the moment of maximum dust concentration.

4. On large shakeouts, conveying equipment usually is of such capacity as to require more time to clear the shakeout hoppers than it does the shakeout to fill them. Consequently, the hoppers act as surge chambers and often are so fully loaded with sand as to completely block the control air connections.

Table 1

RATE COMPARISON ON BASIS OF SEMIMECHANICAL AS AGAINST FULLY MECHANICAL SHAKEOUT

Part	Flask Size, in.	Depth, in. Cope	Depth, in. Cheek	Depth, in. Drag	Castings per Flask	Old Standard, units	New Standard, units
Liner	72 (diam.)	16	28	40	5	214	132
Sheave	72 (diam.)	16	—	16	1	121	88
Engine Frame	92x276	35	—	35	1	997	313
Engine Frame	92x276	17	—	35	1	642	383
Cylinder	60x136	33	—	33	1	370	166
Cylinder	72x90	30	—	30	1	284	113

Table 2

DETAILED RATE COMPARISON BREAKDOWN, SEMIMECHANICAL AND FULLY MECHANICAL SHAKEOUT OF CASTINGS MADE FROM THE SAME PATTERN AND IN THE SAME FLASK

(Engine frame, finished cast weight, 25,000 lb.)

	Cope (35-in. depth) Old Standard Quantity	Cope (35-in. depth) Old Standard Units	Cope (35-in. depth) New Standard Quantity	Cope (35-in. depth) New Standard Units	Drag (35-in. depth) Old Standard Quantity	Drag (35-in. depth) Old Standard Units	Drag (35-in. depth) New Standard Quantity	Drag (35-in. depth) New Standard Units
Runners	2	18	—	22.8	—		—	
Risers	9	12	—	4.5	—		—	
Bolts	47	56.4	47	56.4	47	56.4	39	46.8
Plates		—		—		8		—
Pick up and store flask		49		35		49		35
Vibrate		556		10		84		4.4
Rollover		—		—		15		11.8
Gaggers and rods		53.3		33		—		—
Casting (removal and handling)		—		—		40		40
Open and close shakeout doors		—		6.6		—		6.6
Total		744.7		168.3		252.4		144.6

Old Rate, Total Units.....997

New Rate, Total Units.....313

5. As all of the sand from shakeout must cascade through the air being exhausted, much usable sand is picked up in the air stream, tending to overload the control system and waste usable sand.

The side hood offers an excellent compromise for the average shakeout that cannot be provided with an overhead hood. It is readily fitted to almost all sizes of shakeouts and usually blocks only one side of the shakeout. Side hoods can be so placed that the crane operator's view of the shakeout deck is unobstructed, thus aiding in the movement of flasks to the shakeout.

Side hoods usually are of heavy construction, often with reinforcement to stand abuse from careless handling of flasks. Detail design varies considerably, with a great difference of opinion as to the method of distributing air flow into the hood through the hood face; however, there are certain points which are in general agreement, viz:

1. The side shakeout hood should be installed parallel to the long side of the grate.

2. The hood should extend over the shakeout grate as far as possible without interferring too much with flask handling, say, within 6 to 18 in. of the centerline of the grate, depending upon the type of crane hook and slings used. An overhang of at least one-third the grate area is a good starting point, and each additional in. of coverage improves dust control by extending the "reach" of the control air.

3. The length of the side hood is extremely important; it should be longer than the long side of the shakeout in order to prevent bypassing of control air around the end of the hood. Experiments tend to indicate that the wings (the parts of the vertical face of the hood which extend beyond the ends of the grate) need not be provided with air inlet passages; as a matter of fact, they often are more effective if left blank.

End shields at right angles to



the extreme ends of the hood are most helpful if it is possible to use them.

Side hoods are affected by cross drafts and convection currents in the shakeout area, but to a much lesser degree than downdraft shakeout control.

Air volumes of 400 cu. ft. per min. per sq. ft. of gross grate area is a good starting point for air volume required, and some installations run as high as 500 cu. ft. per min. per sq. ft.

Double side hoods offer good shakeout dust control with somewhat less control air, say, 300 to 400

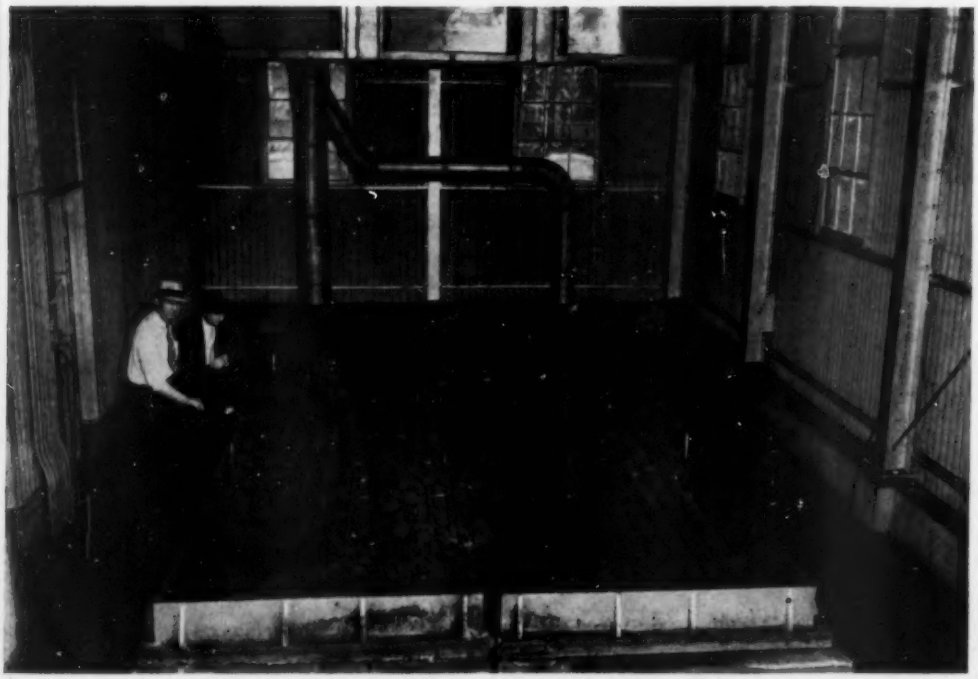
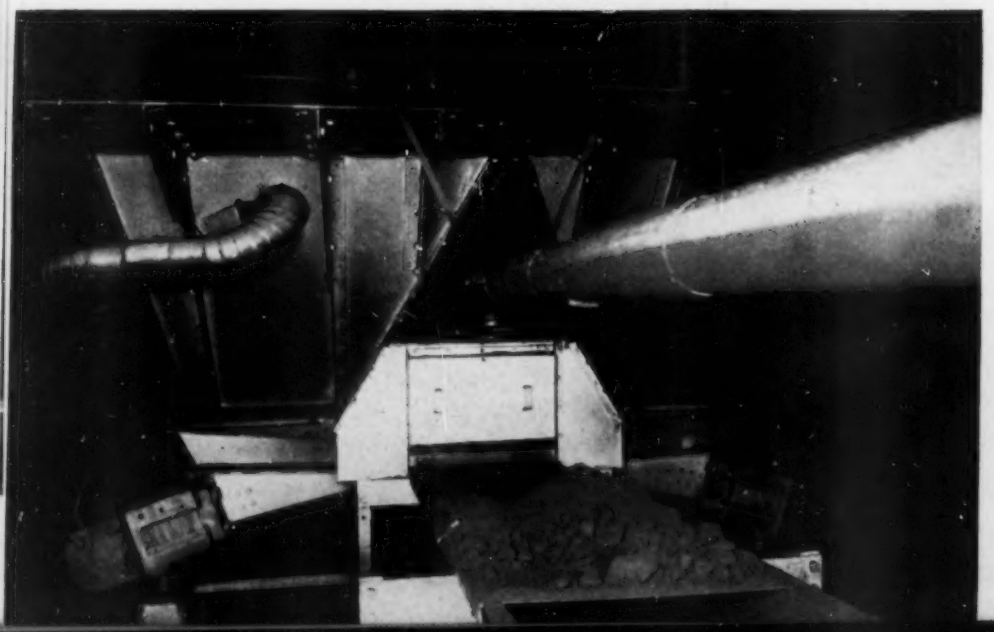


Fig. 6 (top)—General view of main bay showing large molds in the making. Taken before foundry was painted white and large shakeout installed.

Fig. 7 (center)—Interior of shakeout cubicle showing shakeout deck and dust-collecting manifold.

Fig. 8 (bottom)—View in pit under shakeout showing two of the six hoppers and two of the six vibratory feeders discharging to the apron feeder. Pipes shown are downdraft dust collection from hoppers to the left and dust pickup from vibratory feeder discharge and apron feeder housing.



cubic feet per minute.

Canopy type hoods are sometimes used because they fit certain conditions. Picture a situation where castings from shakeout must be delivered to the main bay; the area immediately adjacent to the shakeout is used for making large cores and is swept by a jib crane; the shakeout is fed from a roller conveyor; the sand from shakeout is deposited in a container below, which is removed by overhead crane; and further, that any hood must be movable in order to allow all of the aforementioned functions. Here is

one of the few places where a canopy type hood offers a solution.

Complete enclosure of the shakeout operation has been tried with varying degrees of success. Recently a novel form of extremely large shakeout, which shows every promise of being successful, has been put into operation. This shakeout has been placed in a separate building at the end of the flask storage yard. It has a completely live deck 12 ft. wide and 33 ft. long.

The shakeout enclosure is divided into two compartments, one housing the dust-collecting fan and dust-disposal equipment, and the other the shakeout itself with the dust-collecting manifold.

The enclosure is arranged with an electrically controlled roof section which exposes approximately 75 per cent of the grate area to the overhead crane, and an electrically operated two-section sliding front door on one of the short sides of the enclosure, which provides an opening wider than the shakeout. Both doors and sliding roof are fitted with limit switches to control movement in both directions.

Dust control is accomplished by a manifold on the end opposite the door (Fig. 3) and air velocity



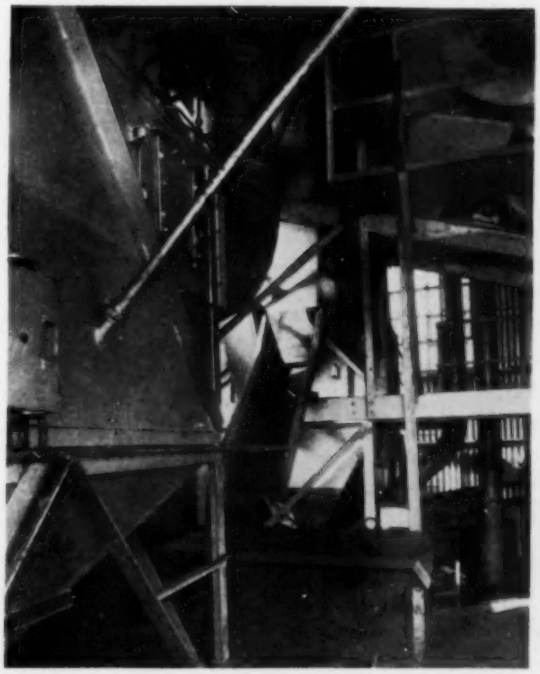


Fig. 9 (center) — Sand handling equipment room. Starting from the left: sludge tank under rotoclone with mud bucket; tramp metal chute and tramp metal box; hexagonal screen housing and fines control; refuse belt at lower right; apron feeder discharge at top right.

Fig. 10 (top)—Sand distribution belt with plows for filling silos. Elevator at far end.

Fig. 11 (bottom)—View showing relation of main foundry to sand storage and handling facilities. Crane door in end of foundry is open.



through the room is controlled by the amount of opening left between the doors when the shakeout is operating. This is variable and, under certain wind conditions, the doors can be left wide open without any escape of dust to the outside atmosphere. When other than the largest flasks are being handled, the sliding room section is opened only far enough to give sling clearance, thus obviating the necessity of removing the slings during shakeout.

In addition to dust control overhead and across the length of the shakeout grate, there are ventilation connections to the hoppers under the shakeout to prevent, insofar as possible, and "dusting" from the hoppers into the equipment tunnel below.

These dust-control connections in the hoppers, constitute a form of downdraft dust control, which tends to control dust at the extreme sides and ends of the shakeout grate. Also, these connections tend to prevent the "upsurge" of dust-laden air, displaced by the large volume of sand released when the sand in the flask "breaks," from the hoppers.

Dust Concentration

Experience with this large shakeout shows that the highest concentration of dust (Fig. 4) is at the time the sand in the flask "breaks," i.e., becomes loose in the flask and, on flasks that are not unhooked from the slings, is the time at which the flask can be lifted clear and taken to the storage pile. By the time the flask has been deposited in storage, the major portion of the sand has passed through the grate and the shakeout enclosure is free of dust. Then the hooker or crane follower can hook up the casting for removal as soon as the crane is back in position. Depending where the type of castings, flasks with the bottom grates removed are shaken out without further dismantling. With other types, the flask is split along the parting line and the casting rolled out before being placed on the shakeout.

Gaggers, rods, risers, gates, shrink bobs and the like are removed from the grate, either by hand or by electric magnet, as required. An attempt has been made to prevent, insofar as possible, any gaggers or other metal going through the grate. Previous experience shows that the majority of forced shutdowns of handling systems and much of the wear and tear on conveyor belts are directly traceable to gaggers in the sand. Much of this difficulty occurs before the sand goes over the magnetic pulleys, and can be prevented by properly sized openings in the shakeout deck.

Handling Sand After Shakeout. Up to this point in laying out a shakeout system, the task is comparatively simple. Beyond this point lies most of the headaches. The principal reason for this is the rapidity with which the modern mechanical shakeout devices release sand from flasks.

Sand is released in a matter of seconds, much faster than it can be taken away except on the smaller shakeouts, where the handling systems must be large enough to take the sand away as fast as it is released from the flasks. Fortunately, the time element in handling the flasks on extremely large installations is such as to allow the sand-handling system to clear the sand from one flask before another is placed in position for shakeout.

Handling systems are tailor-made for a given test set of conditions and vary widely. One of the simplest forms is that in which the shakeout sand is directed into a container that can be lifted by overhead crane and transported to the reconditioning system or other point of treatment.

Another method, and most popular, is that in which the sand is deposited in a hopper under which runs an apron feeder. The feeder in turn deposits the sand on a magnetic belt to remove all iron and steel (gaggers, bars, stools, gates, risers, etc.), and then transfers it to an elevator or other feeding device for transport to storage, mixers or stock pile, as desired.

Shakeout Problems

Large shakeout installations present a number of serious problems.

1. Provisions necessary for rapid handling of huge quantities of sand.
2. Storage facility to keep the sand out of the weather and still immediately available for use.
3. Facility for getting the sand back into the system.
4. Accomplishing all of the foregoing at low cost, considering both cost and operating cost.

As a specific example, shakeout practice and foundry conditions in a large foundry prior to the installation of a large shakeout may be cited.

Large flasks (28 ft. long, 10 ft. wide, and 6 ft. high) were shaken out at night by hanging them in a crane sling and using pneumatic vibrators attached to the flasks, plus the efforts of men with bars and sledges. After shakeout the sand was piled by clamshell bucket and, as required, handled by the bucket into a hopper feeding a synthetic sand preparation system.

The molds, after pouring, were taken to a flask yard to cool (Fig. 5), brought back after cooling for shakeout, and the flasks returned to the flask yard after shakeout to await reuse. All molding in this foundry was in flasks, regardless of size or weight. A sizable portion of the main bay of the foundry, under a heavy craneway, was occupied by a combination of sand pile and shakeout (Fig. 6).

Problem. Cut down shakeout time, rid the foundry proper of sand piles, save wear and breakage of flasks, clean up the foundry atmosphere, and make the area used for sand storage and shakeout within the foundry available for more productive use.

Solution. (a) Mechanical shakeout; (b) mechanical sand handling; (c) outside sand storage.

Large Unit Required

The size of shakeout required was larger than any single unit shakeout on the market, making it necessary to combine six units in order to get the required area.

As it works out, a unit made up of a multiplicity of smaller units has flexibility that is definitely advantageous to the jobbing shop type of foundry (Fig. 7). True, there is the problem of making individual units work alone or in unison with others, but this has been successfully worked out.

The question as to hole size through shakeout grates has been settled on the basis of the smallest holes possible that will allow the grate to clear in a reasonable time without passing gaggers, etc. Openings of 1¼x4 in. have proved capable of handling sand lumps 10x10x8 in. and hard enough to support an average man's weight

Each of the six 6x10-ft. units that make up the fully live deck or grate is individually controlled so that any one or any combination, including all six simultaneously, can be operated. A 9-in. ledge around the outside edge of the shakeout prevents spilling of sand, regardless of how the shakeouts are operated.

Sand Hoppers

Sand from the shakeout drops into individual hoppers, each fitted with an exhaust connection to control dust generated within the hoppers themselves, and further arranged to deposit sand onto vibrating feeders which, in turn, discharge onto a steel apron conveyor (Fig. 8).

Each vibratory feeder is individually controlled through a half-wave rectifier, and so arranged that the rate of feed onto the apron can be controlled by the operator. This arrangement prevents sand sticking in the hoppers, uncontrolled overloading and jamming of the apron feeder, and further allows accelerated movement of sand when shaking out smaller flasks.

From the apron feeder the sand discharges onto a magnetic pulley, and then into a revolving, hexagonal screen. Tramp metal is discharged into a box on a narrow-gauge car so that it can be pushed out under the overhead crane (Fig. 9).

Coke, cinders, and lumps which will not pass through the screen are discharged to a refuse belt and conveyed to a box under the main crane.

The hexagonal screen is provided with an air curtain which can be adjusted to control fines pick-up as required to maintain proper permeability of the sand before it goes to storage or to the mixers.

After cascading through the air curtain the sand lands on another magnetic belt, and small pieces of tramp metal and a considerable quantity of iron oxide in the form of flask scale, etc., are taken out. The magnetic belt discharges onto a cross conveying belt under ground, which runs under two 250-ton capacity salt-glazed tile silos to the boot of an elevator.

The top of the elevator (Fig. 10) is arranged to discharge either to storage in the two 250-ton silos, or into the sand feed bin directly over the muller in the foundry, simply by changing the position of the flop gate.

Sand from either of the storage silos can be tranferred to the foundry by opening gates over the cross conveying belt and again using the elevator with the flop gate set to discharge to the foundry conveyor (Fig. 11).

Dust control of the shakeout is as previously described, with wet rotoclones handling the dust and automatically discharging it as a wet sludge. Control of the system dust, from transfer points, fines control, tunnel, and etc., is handled by a second rotoclone, again with the dust being discharged as a wet sludge.

Considering that the air drawn into the shakeout enclosure may be at freezing temperatures and that water is sprayed into the rotoclones to collect and wash out the dust, provision has been made to supply heated water to the shakeout rotoclone. The shakeout cubicle is not heated, and no trouble has been experienced in taking air directly from the atmosphere at temperatures down to 10° F.

First and foremost result of the installation is the cleaner foundry atmosphere, which has prompted the use of white paint on all interior walls of the foundry, with a gray dado. All equipment and bins are painted aluminum, jib cranes aluminum with yellow stripes, and all overhead cranes yellow.

Overhead lighting is seldom necessary during daylight hours. Windows, skylights, reflectors and light bulbs require less cleaning. Light reflection from white paint is greater and the whole atmosphere brighter.

Washroom Facilities

Up-to-date washrooms, in keeping with the generally cleaner foundry, make the foundrymen the cleanest and best street-dressed group coming into and leaving the plant.

Mechanical shakeout has contributed heavily in making this foundry a pleasant place to work. Furthermore, it improves foundry production and decreases cost by eliminating much of the laborious work and making it possible to do the remaining work efficiently.

Examples taken at random from rate cards are listed in Table 1 and show over-all shakeout savings made possible through the use of mechanical shakeout. These figures show only reduction in rates for a given operation and not the benefits from the large number of intangibles, all of which tend to improve other operations and contribute to greater production and lower costs.

Large Shakeout Advantages

Further, it is apparent that by and large a mechanical shakeout can handle molds faster than they can be fed to it; therefore, in projecting expected savings through the use of mechanical shakeout, they must be predicated on the slowest link in the chain, which usually is transportation of the flask to and from the shakeout.

An example of this is shown in Table 2, which gives a comparison of rates on semimechanical as against fully mechanical shakeout of the same flask and casting. The actual shakeout time is vastly reduced, but other factors, which are part of the complete cycle of operations, remain the same.

DISCUSSION

Chairman: JAMES THOMSON, Continental Foundry & Machine Co., East Chicago, Ind.

Co-Chairman: C. P. GUION, W. W. Sly Mfg. Co., Chicago.

MEMBER: How much area do you have through the door opening?

MR. YATES: We have not yet determined what the minimum is going to be. We have a 36-in. rotoclone handling that room. According to the New York State Code, we would have to have one twice or three times as large. Fortunately, however, the New York State Code does contain a clause which makes it possible for a properly engineered job that comes out with the desired result to use other than the stipulated velocities and the cfm. per sq. ft. of grate area as given in the code.

In this particular installation we are under very close scrutiny. There are dwellings immediately across the street from the shake-out. On the other side, there is a public park. Any dust or any smoke that gets away from this plant is immediately spotted and it does not take 15 min. before the Smoke Abatement Inspector calls it to our attention.

On the operation of the shake-out, we can set the doors as we see fit and believe we can adequately control the dust with a comparatively small amount of air. This shake-out figures at about 125,000 cfm. We are doing it with something less than 50,000 cfm.

MALLEABLE CORE MAKING PRACTICE

In the mechanized foundry of today, the sand used in the coreroom is the basis for the sand in the foundry. Proper selection of sands for cores is of utmost importance as these sands will largely determine the casting finish obtained and permeabilities of foundry sands.

E. C. Zirzow
National Malleable & Steel Castings Co.
Cleveland

CONTROL OF CORE SANDS has been neglected by the foundry industry as a whole. Great strides have been made in the proper control of foundry sands, but little has been done with core sands and mixtures, and it is the author's intention to present a brief summary of general coreroom practice in the malleable foundry with which he is associated.

Sands are received in the plant in open hopper cars. These sands are unloaded and placed in the various bins in the storage bay. From the bins the sands are taken by means of a grab bucket and crane to the sand dryer. At the dryer the moisture is reduced to about 0.3 per cent and sand conveyed by means of an elevator and belts to the storage hoppers in the mixing room.

Sand Buggy

Sands are drawn from these bins through hand-operated gates and placed in a buggy which holds 800 lb. of sand. The sand buggy is pushed by hand to the electrical hoist at the sand mixers. At this point the sand and buggy are hoisted above the sand mixers and the sand flows into the mixer. The buggy is lowered and is ready for the next batch.

Clay and cereal binder are received in box cars and transported by truck to the storage room in the basement. From storage it is conveyed by hand truck to the mixing room. The mixer operator then dumps the bags into bins located at the two mixing machines.

Core oil is blended in the core oil plant in the foundry and pumped into large storage tanks for aging. From these aging tanks the core oil is pumped to smaller tanks in the mixing room.

After the sand has been placed in the mixer as described, the green bonding material is added. These binders are measured and added by volume. This mixture of sand and green binder is mixed for one min. Core oil is then added, followed immediately by the water, and mixed an additional 4 min.

Core Sand Distribution

The core sand mixture is discharged from the mixer and conveyed by elevator to the distribution bins located in the mezzanine floor above the coremakers. From these distribution bins the mixture is taken by cart or buggy and dumped into the hoppers of the coremakers.

Coremakers are grouped about four vertical-type continuous ovens (Figs. 1 and 2). About 90 per cent of the work is made on core blowing machines, the remainder being bench and roll-over. After the cores have been made they are placed directly in the oven by the coremaker. The cores are unloaded on the opposite side of the oven, placed on gravity roller conveyors and permitted to cool (Fig. 3).

After cooling the cores are finned, gauged and mudded and placed in boxes. These boxes of cores are transported by buggy to the various molders' stations as needed. The coreroom endeavors to keep a 4-hr. supply of cores ahead of the foundry.

All sands as received are practically clay free. These sands can be divided into three classifications, coarse, medium and fine. A fourth sand is used for special work requiring a high permeability. Typical screen analyses of the four sands are listed in Table 1.

Medium-grained sand is the basis sand and is the sand upon which the entire sand control in the foundry is based. From this sand the majority of the cores is made. If it is necessary to obtain more permeability, either in the core or the foundry sand, coarse sand is blended with this medium-grained sand. To decrease the permeability it is the practice to blend in the fine-grained sand. Every effort is made to do this blending in the cores in order to save additions of the coarser- or finer-grained sands in the foundry.

Defective Cores

In most cases, this is the most economical procedure. However, there are exceptions. If it is found that with additions of the finer- or coarser-grained sands defective castings are produced due to defective cores or, in other words, the core has not met the required specifications, then the core sand is changed so that the core will do the job required of it in the foundry.

Sand grains of the three regular sands are of sub-angular shape. The special sand is composed of sub-rounded grains. All four sands have minimum fusion points of 2850° F.

This paper was presented at a Malleable Foundry Practice Session of the Fiftieth Annual Meeting, American Foundrymen's Association, at Cleveland, May 8, 1946.



Fig. 1—View showing coremaking machines grouped about the core ovens.

and go as high as 3100° F. maximum.

With the exception of the fine sand, these sands withstand thermal and mechanical shock. The slight breakdown of the grains due to these shocks is well above the permissible amount.

All sands are subjected to screen analysis before unloading. Tolerances are set up, and the sands must come within these tolerances or they are rejected.

For the development of green bond in the core sands, corn cereal and bentonite are used. The cereal is tested on a weight per volume basis and must meet the specifications. The bentonite is tested in the foundry and must meet the foundry specifications.

Core oil consists of a blend of a vegetable drying oil, rosin and kerosene. These ingredients are blended with heat and aged in storage tanks.

At the present time paddle-type mixers of an ancient design are used. These mixers first mix the sand and green binders, which are in the dry state. The sand grains then are coated with the oil and, finally, the water develops a paste from the green binders. The coating of the grain with the oil and the development of the green bond are accomplished simultaneously.

Various mixtures of the sands, green binders and core oils are used to produce cores with various physical properties. These physical properties are varied to meet specific foundry conditions. Specific examples of these mixtures are listed in the following paragraphs.

Mixture No. 1 is used for cores requiring maximum permeability, or cores which, because of design, it is impossible to vent by mechanical means. The scratch hardness of these cores will run between 55 and 65.

Mixture No. 2 is used for cores requiring high permeability and slightly higher green and baked strengths. The scratch hardness of the cores made from this mixture will run between 60 and 70.

Mixture No. 3 is used for cores which require an exceptionally rapid collapsibility. Baked scratch hardness of these cores will run between 60 and 70.

Mixture No. 4 is perhaps the most widely used of all the mixtures. This mixture must be sufficiently open to permit ready escape of the gas, and have sufficient strength in the baked condition to prevent cutting by the metal. Green strength must be sufficiently high to prevent distortion on the form dryer. Baked scratch hardness of these cores will run from 70 to 80.

Mixture No. 5 is used for cores requiring a high green strength. These cores usually are made without the aid of form dryers. Baked scratch hardness will run between 60 and 70.

Mixture No. 6 is a hard sand and is used for ram-up, gate cores, and on all cores which must withstand abuse in the molding operation. Cores of this mixture will have a baked hardness of 80 to 90.

All cores are baked in vertical-type continuous ovens. These ovens are equipped with a variable speed control so that the baking cycle may be varied as the need arises. The burners are designed so that either natural gas or fuel oil may be used. Natural gas is the usual fuel.

Temperature may be varied and is controlled by means of a gas-filled

TABLE NO. 1

RETAINED ON	COARSE	MEDIUM	FINE	SPECIAL
12	0.00	0.00	0.00	0.00
20	9.30	2.20	0.15	0.00
30	20.65	2.80	0.50	1.30
40	28.95	6.35	1.50	59.25
50	23.00	23.00	4.50	36.35
70	10.20	40.00	16.00	3.10
100	6.00	20.70	41.50	0.00
140	1.70	4.10	21.05	0.00
200	0.10	0.60	7.70	0.00
270	0.05	0.20	2.45	0.00
PAN	0.05	0.05	4.65	0.00

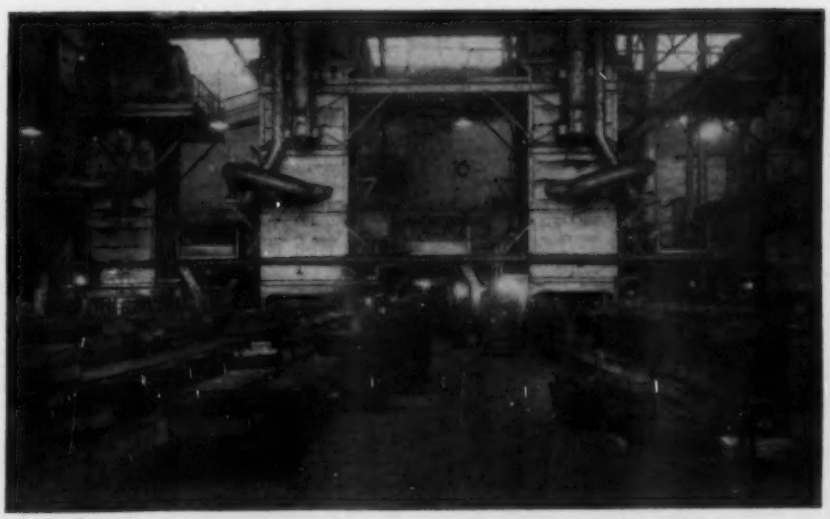


Fig. 2—Battery of vertical-type continuous core ovens.



Fig. 3—Photograph of operation at core oven unloading station.

tube extending through the oven at approximately the hottest point. A portion of the gases and products of combustion or recirculated through the oven by means of a recirculating blower. Natural draft is used to exhaust the moisture and volatile combustible matters formed in the baking operation through a series of adjustable flues on each oven.

Load-to-unload cycle will vary from 60 to 70 min. Graphs (Figs. 4, 5, 6 and 7) show the results of running a traveling thermocouple through the four ovens. Temperatures were read for every foot of travel.

On oven No. 1 the moisture evaporation point is reached in about 10 min. and 20 sec. Baking temperatures are reached in 17 min. and 34 sec. Actual time in the baking temperatures is 39 min. and 16 sec. This leaves a cooling zone of 9 min. and 18 sec.

Moisture evaporation point on oven No. 2 is reached in 9 min. and 18 sec. Baking temperatures are reached in 15 min. and 30 sec., and the temperature remains above the baking level for 38 min. and 14 sec. Cooling zone is the last 12 ft. of travel, or 12 min. and 24 sec.

Oven No. 3 reaches the moisture evaporation point in 11 min., the baking temperature about 30 sec. later. The cores remain in the baking zone for 40 min. and 30 sec. and are then cooled in the last 12 ft., or 12 min. of travel.

The graph (Fig. 7) for oven No. 4 shows that the moisture evaporation point is reached in 13 min. and 52 sec. Baking temperatures are reached in 18 min. and 8 sec., and the cores remain at or above the baking temperature for 38 min. and 12 sec. The remaining 11 ft. or 11 min. and 44 sec. comprise the cooling zone.

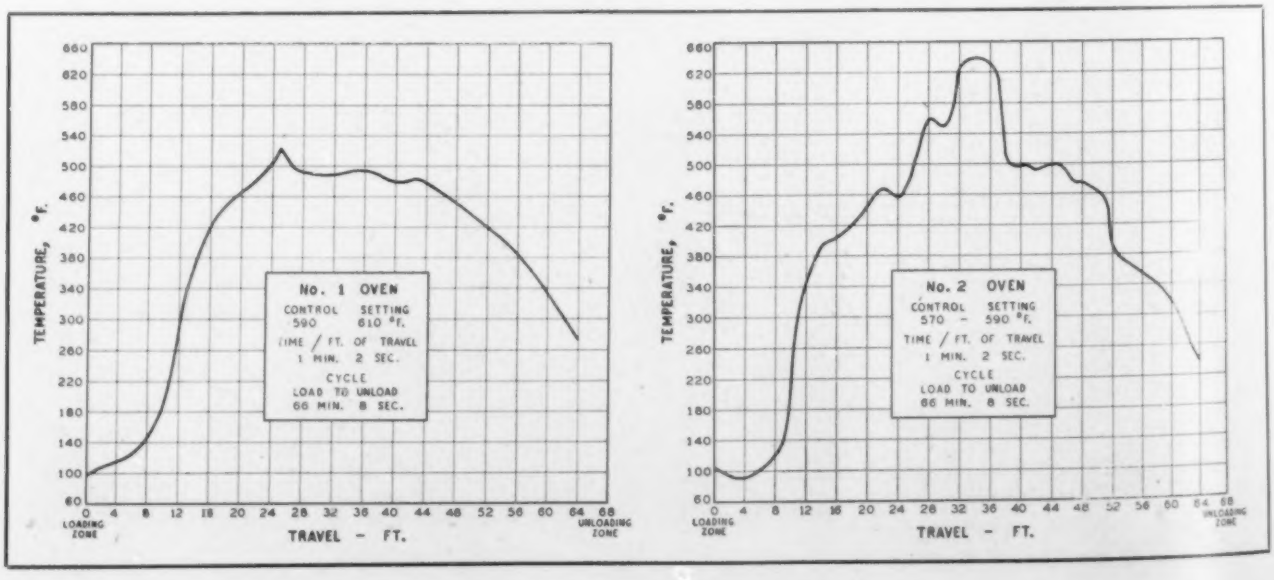
Temperatures mentioned in the foregoing are oven temperatures and not core temperatures. No doubt, the outside of the core will reach the temperatures shown in approximately the final 15 ft. of travel, but it is with considerable difficulty that temperatures much above 212° F. are attained in the core centers, especially in large cores. At present, this plant is in the course of determining the temperatures reached in the center of various sizes of cores, using the oven temperatures and cycles given in the foregoing.

A relation between the temperature control setting and the actual temperatures attained in the oven is not apparent. This may be due to the gas tube not being located in the proper place in the oven, or to the gas tube going out of calibration.

Gas-Filled Tube

It is possible that the traveling thermocouple may not be giving as good an average as the gas-filled tube which actuates the burner. However, the great variation in the control settings of the various ovens would indicate that the fault lies with the gas-filled tubes. Regardless of what the temperature control setting may be, the final criterion

Fig. 4 (left)—Chart showing temperatures through a cycle of No. 1 core oven. Fig. 5 (right)—Temperatures through a cycle of No. 2 core oven.



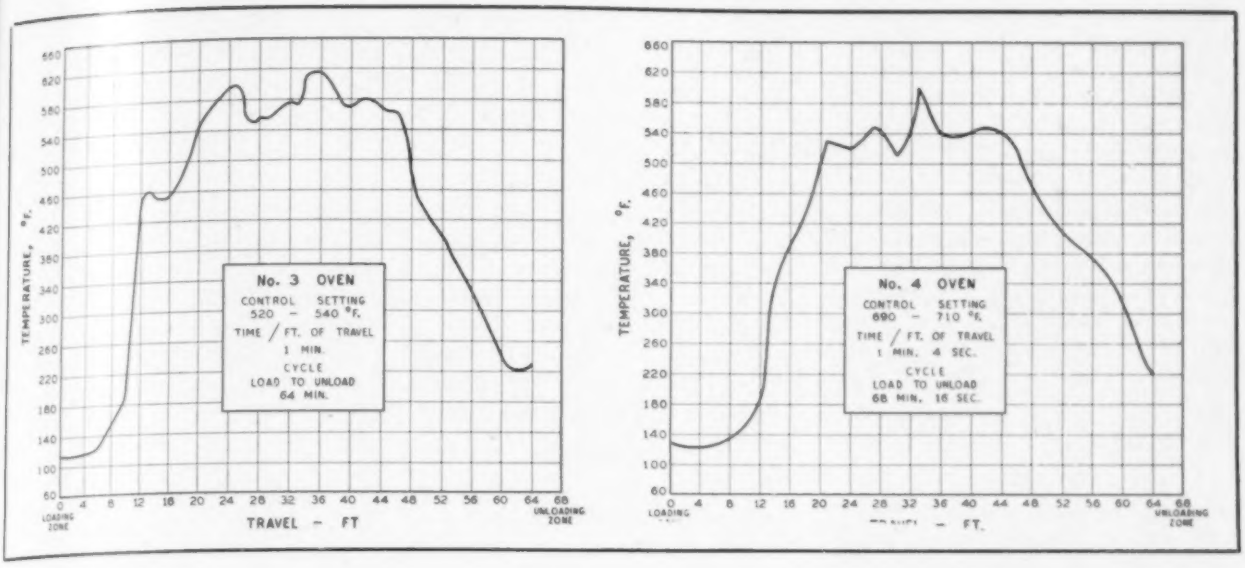


Fig. 6 (left)—Temperatures through a cycle of No. 3 core oven. Fig. 7 (right)—Temperatures through a cycle of No. 4 core oven.

should be the actual results obtained on the cores.

While on the subject of core ovens and core baking, it may be advisable at this time to review briefly the mechanics or steps involved in the process of baking cores.

Binders may be roughly divided into four classes: (1) those becoming solid or firm upon freezing; (2) becoming firm at room temperature; (3) becoming firm upon heating; (4) clays. In other words, cores requiring a subnormal temperature, cores requiring room temperature, cores requiring heat and, of course, clays in a class by themselves.

Malleable Cores

As a whole, the malleable industry does not use cores requiring a subnormal temperature or room temperature to become solid. However, use is made of binders which become firm upon heating, and of clays. The binders which find the greatest application in the making of malleable cores are those which become firm upon heating.

This class can be further subdivided into the binders that occur in the solid state, such as pitch and rosin, and which, upon the application of heat, become a viscous liquid which surrounds the sand grain. Upon cooling this mixture of sand and binder resumes its original state and forms a solid mass of the required shape.

The second subdivision includes the cereals, sulphites and proteins, and may be classified as adhesives. Adhesives, upon the introduction of water, become gelatinous and produce the green strength in the mixture. Upon application of heat these binders are dehydrated and become hard, thus holding the sand grains together.

A third subdivision is that of the core oils, which group includes the vegetable oils, wood oils, fish oils and the polymers, either vegetable or mineral. In the formation of the baked strength of this group are five distinct steps. It is true that some of these steps may overlap each other, but they do occur and must occur for proper baking of a core containing these ingredients.

1. Evaporation of moisture.
2. Evaporation of volatiles.
3. Oxidation.
4. Polymerization.
5. Decomposition.

Moisture evaporation is the first step and requires the greatest amount of time. The core temperature will not rise above 212° F. until all water has been evaporated. Silica sand is a notoriously poor conductor of heat, and temperature must be attained through the entire body of the core for proper baking. The most common defect in cores baked in continuous ovens is that of green center.

The core surface is heated sufficiently to drive out the moisture, but ample time is not allowed to bring the center of the core to the temperature necessary to permit the evaporation of moisture to proceed. However, temperature alone is not the only method available for evaporation of moisture. Evaporation depends upon the changes of air. It is much easier to dry clothes on a windy day than upon a day of little or no breeze. In the drying of clothes temperatures of 212° F. are not used but, nevertheless, the clothes do dry.

Moisture Evaporation

Hence, in the evaporation of moisture from cores it is necessary to have (1) a movement of dry air through the oven and (2) to avoid saturation of the air, a sufficient temperature. It is a known fact that it is harder to bake cores on a humid day, when the air is saturated with moisture, than upon a day when the air is dry.

In the graphs (Figs. 4, 5, 6 and 7) showing the temperature and cycles of the various ovens, about 50 per cent of the total time the temperatures of the ovens are at or above the baking temperature, or about 35 to 40 min. This is ample time to permit of evaporation of the volatiles, and for the oxidation and polymerization to proceed provided that the moisture has been removed. Evaporation of the volatiles proceeds along with the evaporation of the moisture. The oxidation will proceed as soon as the oil has been freed of moisture and the oxygen in the air can have free access to the oil.

The principle of evaporation can be used to prevent the decomposition or burning of thin sections of cores by saturating these thin sec-

tions with moisture, and is a common practice.

Use of clay as a binder in cores is quite common. Various green cores are made of molding sands, and not only rammed but blown.

Cores in the malleable foundry are made on the bench, with a blow machine, and with a jolt or roll-over machine (Fig. 8). Cores may be either hand rammed or rammed with a pneumatic hammer. To the best of the author's knowledge, malleable cores are not made with a sand slinger.

The method used in making cores will depend upon:

1. The number of cores required.
2. The available equipment.
3. The form or shape of the core.

Generally speaking, if only a small number of cores are required on the order, from a cost standpoint it would not be economical to prepare a blow plate and vent the core box. In other cases, the cores could be made in the time required to set up the machine.

Make Machine Cores

However, if the order is large and the equipment is constructed of material that will stand up under the stresses encountered in blowing a core, it is more economical to make the cores on the blow machine. Core boxes constructed of wood or plaster do not stand up under the continuous erosive action of sand and air.

Many types of cores, due to their shape and size, do not readily adapt themselves to the blow machine. In many cases it is more economical to make these cores on the bench.

In this plant about 90 per cent of the work is made on the blow machine. Between 80 and 90 lb. air pressure is used. The average size of the blow hole is about ¾-in., with a minimum of ½-in. and a maximum of one in. The usual practice is to start off with a ½-in. hole and work from that point. The objective in blowing a core is to get the sand and air into the box, and then get the air out as fast as possible. The proper venting, blow holes, not only size but location, and air pressure determine to a large extent the uniformity of the core produced.

Core Work

Some cores after baking do not require any further work other than unloading and placing in proper receptables. Others merely require finning, which may be due to worn equipment and improper clamping.

Still other cores must be coated to prevent metal penetration. Many commercially prepared coatings which meet these requirements are available. Cores should be coated while they are still warm in order to insure that the coating will be dry. If this is not practicable, the coated cores should be placed in a drying oven and all moisture driven from the coating.

Many times defects are caused by the improper use of coatings. Core wash manufacturers have trained technical men who will recommend the proper wash and give instructions regarding the proper application. Cores may be coated while green by dipping or spraying. Dry cores may be coated by dipping, spraying, swabbing or brushing.

Many cores are of such design or shape that it is impossible to make them as a single unit. This requires the assembly of two or more cores, and the usual practice in the malleable shops is to paste the various parts together, using a jig or specially prepared fixture. Many satisfactory pastes are on the market. Literature on the proper use of these pastes is supplied by the manufacturer. Here again the paste must be thoroughly dry before the core is sent to the foundry.

Usually, when two or more cores are pasted together the joints must be mudded up. Mudding compounds may be merely molasses and sand, clay and sand, or any of the commercial mudding compounds found on the market. Their function is to seal the joint against metal penetration. The mudding compounds must not spall off during the pouring operation and they must not produce an excessive amount of gas which may cause a casting blow.

Core Properties

Properties desired in a core are largely determined by its use in the foundry. A core must have sufficient strength to withstand the abuse to which it will be subjected in unloading, transporting to the foundry and handling by the molder. The core must be capable of resistance to erosion and deformation by the molten metal.

Ram-up cores must have sufficient strength and resilience to stand up under the jolting, ramming and squeezing encountered in making the mold. Cores must have sufficient



Fig. 8—Photograph showing variety of boxes used in making cores.

TABLE NO. 2

INGREDIENTS	MIX 1 %	MIX 2 %	MIX 3 %	MIX 4 %	MIX 5 %	MIX 6 %
FINE SAND					22.83	6.90
MED. SAND		47.43	52.14	53.13	46.75	63.22
COARSE SAND			23.18			
SPEC. SAND	95.74	47.43		40.84		
SCRAP SAND			18.54		22.83	22.99
CEREAL BNDR	1.07	1.40	1.74	1.40	1.03	1.03
BENTONITE					.17	
CORE OIL	.12	.24	.35	1.16	1.37	2.18
WATER	3.07	3.50	4.05	3.47	5.02	3.68

TABLE NO. 3

TEST NO.	MIX NO. 1 A CORE	MIX NO. 2 B CORE	MIX NO. 3 C CORE
1	16.0 CC GAS/GR.	29 CC GAS/GR.	12.8 CC GAS/GR.
2	19.5 "	13.6 "	6.8 "
3	15.0 "	18.5 "	13.4 "
4	14.1 "	14.3 "	10.0 "
5	15.4 "	13.8 "	10.5 "
6	18.0 "	12.7 "	8.6 "
7	19.6 "	13.9 "	8.8 "
8	20.0 "	15.3 "	10.6 "
9	19.4 "	13.0 "	8.3 "
10	18.7 "	16.9 "	10.4 "

TABLE NO. 4

TEST NO.	MIX NO. 3 A CORE
1	36.4 CC GAS/GR.
2	25.6 "
3	23.8 "
4	38.7 "
5	40.7 "
6	18.0 "
7	27.1 "
8	40.0 "
9	39.9 "
10	40.0 "

TABLE NO. 5

TEST NO.	MIX NO. 4 A CORE
1	24.1 CC GAS/GR.
2	25.2 "
3	23.2 "
4	24.2 "
5	26.3 "
6	20.6 "
7	30.5 "
8	21.2 "
9	30.4 "
10	26.7 "

TABLE NO. 6

TEST NO.	MIX NO. 5 A CORE
1	52.7 CC GAS/GR.
2	55.7 "
3	58.7 "
4	29.8 "
5	43.2 "
6	55.6 "
7	59.1 "
8	39.5 "
9	36.8 "
10	65.7 "

TABLE NO. 7

TEST NO.	MIX NO. 6 A CORE
1	26.3 CC GAS/GR.
2	27.1 "
3	39.5 "
4	29.8 "
5	26.4 "
6	25.6 "
7	25.5 "
8	33.0 "
9	38.2 "
10	39.0 "

permeability to permit ready escape of the gases formed in the pouring operation. Residual gas must be low to prevent gas entering the metal and to allow the metal to remain quiet.

Cores must disintegrate rapidly after the initial set of the molten metal to prevent strains in the castings and to minimize the shake-out effort.

Core sands should be tested at regular intervals for:

1. Green permeability.
2. Green strength.
3. Moisture.
4. Baked permeability.
5. Baked strength.
6. Fineness.

In addition to the foregoing, many special tests are available. These include:

1. Gas content and gas pressure.
2. Retained strength.
3. Collapsibility.
4. Deflection.
5. Hot strength.
6. Hot permeability.
7. Expansion and contraction.
8. Blowability.
9. Sintering.
10. Spalling.
11. Warm strength.
12. Hardness.
13. Strength of core paste.
14. Moisture absorption in humid atmosphere.
15. Stickiness.

As previously stated, all sands in this foundry are tested for fineness. Typical screen analyses of these sands are given in Table 1.

Tests are run at regular intervals for green strength and green permeability. It is the practice to use green tensile rather than green compression. Green tensile strength will run from 6 to 18 grams per sq. cm. The average green tensile strength of the core sands is from 8 to 10 grams per sq. cm.

It is debatable whether green tensile or green compression is the most desirable test. Although the green tensile test is used in this foundry, perhaps green compression would tell the technician more about his cores in the green state than would green tensile. The most common defect encountered in the green state is the failure of the core to hold its shape. This defect would be caused by sagging or low green compression. However, the testing procedure makes the green compression test rather difficult to run.

Green Permeability

Green permeability of core sands is determined in the same manner as green permeability of foundry sands, and is found to run from 50 to 300, the average being 120.

Moisture is determined by the conventional oven method, and on core sand mixtures will run from 2.5 to 5.5 per cent. The six core sand mixtures given in Table 2 show moistures of 3.00 to 5.00 per cent.

Baked permeabilities vary in about the same ratio as green permeabilities. Baked permabilities usually are higher than green permeabilities. However, if the green and baked permeabilities are determined on a given mixture the difference between the two permeabilities will always be approximately the same. Once this difference has been determined the baked permeability can always be estimated from the green permeability.

Baked Strength

Three commonly used methods for determining baked strength are available; compression, tensile, and transverse. Of these three, baked tensile is perhaps used by the majority of foundries. The Core Test Committee A.F.A. is now in the course of determining which of these three tests will give the most information. This committee may find that the transverse test plus deflection may tell more about the stresses to which the core is subjected than either compression or tensile.

This malleable foundry has been using the baked tensile test, but the results have not been entirely satisfactory. The baked tensile strength of the cores will run from 50 to 350 psi. These strengths are determined not only by the percentage of binder present but by the fineness of the sand, the shape of the grain, mixing procedure, moisture content, baking procedure, length of time elapsing before cores are placed in the oven, atmospheric conditions and the presence of impurities or fines.

In the mixtures given in Table 2, baked tensile strengths obtained are:

Mixture No.	*Baked Tensile Strength, psi.*
1	60
2	80
3	60
4	120
5	100
6	275

Of the special tests previously listed, scratch hardness and gas content tests are used in this foundry's standard testing procedure.

Scratch hardness testing replaces the old method of determining hardness by scratching with the fingernail. The test is easily run and can be used on regular run of cores. Furthermore, it does not require an expert to use it. Tests have shown that no direct relation exists between hardness and strength. In a given mixture there is a direct relation between ramming energy and hard-

ness, provided that the baking has been uniform.

Hardness of the mixtures shown in Table 2 are as follows:

Mixture No.	*Hardness*	*Core Oil, per cent*
1	55-65	0.12
2	60-70	0.24
3	60-70	0.35
4	70-80	1.16
5	60-70	1.37
6	80-90	2.18

Hardness varies from 55 to 90. Generally, the hardnesses show a direct relation to the percentage of oil, but the pattern of this relation is not regular. Mixture No. 5 show a low hardness because of the presence of clay in the mixture.

Considerable work has been done on the gas contents of the core sands. Two important factors must be considered when determining the gas content of core sand mixtures, and these are time and temperature. The apparatus available for determining gas content consists of a resurrected carbon combustion furnace with rheostat temperature control, a vitreosel tube, drying tube, and a gas burette.

Vitreosel Tube

The vitreosel combustion tube is firmly packed for about half its length with asbestos, and the end sealed. The drying tube is filled with glass wool and silica-gel. It is preferred to dry the gas. The gas burette is filled with an acid solution of 10 per cent potassium sulphate.

A weighed sample is placed in a boat and this boat is placed in the furnace, which is held at a temperature of 1800° F. The sample is permitted to burn in the atmosphere of the tube for 8 minutes. The volume of gas produced is then measured. The gas content is recorded as cu. cm. of gas per gram of sample.

The sample should be representative of the core. It is found necessary to run at least five determinations and take the average. Tables 3, 4, 5, 6 and 7 list the gas contents of the six sand mixtures given in Table 2.

Gas Content

In Table 3, the gas contents are those of an assembly of three cores using sand mixtures No. 1 and No. 2. These two mixtures run fairly constant in gas content, variations being due to the degree of bake. The lower contents are caused by the start of decomposition in the bake.

Gas contents of Table 4 are not as uniform. This variation is caused by poor baking and the introduction of scrap sand into the mixture.

Variations in gas contents shown in Table 5 indicate the various degrees of bake. This is a core which seems difficult to bake properly. It reflects poor oven control. Sand mixture No. 4 is one that should bake readily.

TABLE NO 8

SECTION	DYE	VISUAL EXAM OF BAKE	CC GAS/GRAM	SCRATCH HARDNESS
½"	INVISIBLE	MEDIUM	26.6	60
1"	"	"	30.6	67
1½"	"	"	30.4	66
2"	VISIBLE	MEDIUM LIGHT	33.0	70
2½"	"	"	34.9	58
3"	"	"	33.8	59

Gas contents of sand mixture No. 5 listed in Table 6 show considerable variation. This mixture has a high percentage of scrap core sand. The moisture content also is high due to the addition of clay to the mixture. This mixture reflects the degree of bake and the variation in the scrap core sand.

Table 7 gives the gas content of core sand mixture No. 6, which is a hard sand. This mixture must be baked thoroughly. The variations in gas content are due to the variation in the scrap core sand used in the mixture and to the degree of bake.

Gas vs. Cereal Binder

It has been found that three conditions are responsible for the various differences in gas contents in the core sand mixtures. A direct relation exists between the gas content and the percentage of cereal binder. The use of any scrap core sand in any core sand mixture will cause a variation in the gas content This is self-explanatory as this scrap sand may be any one of the mixtures or a combination of all of the mixtures given in Table 2. Finally the gas content shows a relation to the degree of bake. It has been found that gas contents have helped the sand technician to determine whether the right sand mixture is being used, to detect variations in mixing, and enables him to determine whether the cores are receiving the proper bake.

Two additional tests are used for checking the bake. The first method is to introduce a dye into the mixture. The dye will disappear if the core has reached the proper baking temperature. This method has been used by several foundries not only for determining the bake but for means of identification of the vari-

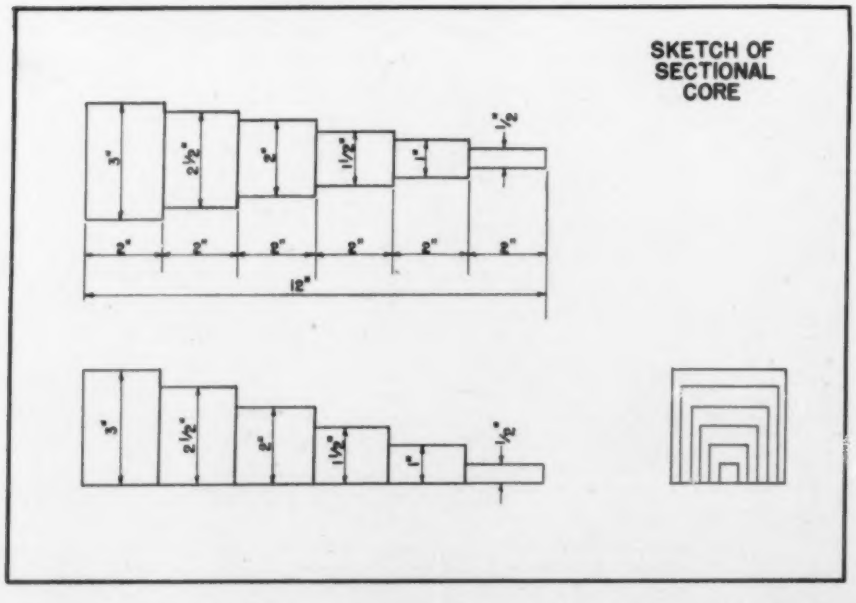


Fig. 9—Sketch of sectional core used to determine baking properties of various core sections from ½ to 3 in.

ous core sand mixtures. This foundry has found it too expensive to use in every mixture.

The second test is one that any practical coreroom man can use to determine how various core sections are baking, and is similar to Woodliff's method. It has been used in this foundry for years. A core of various sections is used instead of a one-section core. This core consists of six sections as shown in the sketch (Fig. 9). These sections vary from ½ to 3 in., approximating the sections of the various foundry cores. It is not necessary to stop at the 3-in. sections. These sections can be made to conform to the minimum and maximum core sections of the foundry.

For visual examination cores are made from any or all core sand mixtures. Sometimes it is necessary to select one specific core sand mixture and make one core for each oven. In this manner the bake on the different ovens or shelves of the ovens can be determined.

After the cores are baked the core foreman or some designated person breaks the cores into the various sections and determines the bake by visual examination. The test is carried one step further. Visual examination is checked by determining the gas content and hardness of the various sections.

Table 8 gives the gas content and hardness obtained on the sectional test, using sand mixture No. 4 from Table 2. On this test the dye was used. The dye had been completely volatilized on the ½, 1 and 1½-in. sections, but was still visible in the 2, 2½ and 3-in. sections. Hardness was fairly uniform, showing a uniform ram.

Visual examination, gas content and dye test all showed that the sections of from ½ to 1½ in. were completely baked, but the sections of from 2 to 3 in. were not completely baked. Interpreting these results, it could be expected that all cores of ½ to 1½-in. sections made from this sand mixture and baked in the same oven would be properly baked. However, if the sections exceeded 1½ in. it is probable that the cores would not be thoroughly baked.

Summary

In summarizing, the author has attempted to show that it is possible to control core sand mixtures and cores. Due to the number of variables in the making and processing of cores, the sand technician must be constantly on the alert. It has been shown that simple ways of checking some of these variables are available. The importance of the gas content test has been stressed, not only from the standpoint of producing solid castings but as a check on coreroom practice.

DISCUSSION

Chairman: W. D. McMillan, International Harvester Co., McCormick Works, Chicago.

Co-Chairman: D. F. Sawtelle, Malleable Iron Fittings Co., Branford, Conn.

Chairman McMillan: Is the oil-rosin mixture a 60-20 or a 60-17 proportion?

Mr. Zirzow: It is a 50-30 mix.

Chairman McMillan: Did you use rosin oil?

Mr. Zirzow: No, we used dry rosin and we heated the oil.

Harold Henderson[1]: Did I understand you to say that cereal was a great producer of gas?

Mr. Zirzow: Yes, in a core it is.

Mr. Henderson: Would you elaborate on that? What proportion of gas is created by the amount of cereal you use in comparison with your oil?

Mr. Zirzow: In comparison with your oil, it runs between 70 and 80 per cent of the gas produced in the core.

David Tamor[2]: Are the dyes described in your paper?

Mr. Zirzow: No. The Venetian dye is the best dye. There is also an acid green that works very well. There has not been enough work done on the volatilization point of dyes to make them absolutely reliable.

Mr. Tamor: Are they water soluble dyes?

Mr. Zirzow: Yes, you use an ounce of dye with a half ton of sand. That is all that is needed, but it is a little too expensive. We use the dyes more or less as a test. We do not use them all the time.

[1]H. C. Macaulay Foundry Co., Berkeley, Calif.

[2]American Chain & Cable Co., York, Penna.

Mr. Henderson: In taking the scratch hardness test, how might we obtain uniformity of results?

Mr. Zirzow: Do you mean in the manner in which you take the tests?

Mr. Henderson: Yes. Should only one man perform the test and should you depend entirely on his sense of touch? I notice we get large reading variations.

Mr. Zirzow: I think it is definitely advisable to have one trained man perform the test.

Mr. Henderson: There must be an art connected with the science then.

Mr. Zirzow: Do you take just one reading on the core? With only one reading you would not get accurate results.

Mr. Henderson: We take several readings.

Mr. Zirzow: If you take several readings and average them you will find they correspond. I take the inside hardness as well as the surface hardness. It is entirely possible that you are not getting uniformly rammed cores, hence the hardness variations.

Member: I ran tests with various types of lead wires and encountered considerable difficulty.

Mr. Zirzow: I use the asbestos covered lead wires.

Member: Continuous wire? All the way through?

Mr. Zirzow: That is right. Asbestos-covered wire gave us the best results.

Member: In line with that question on the thermo-couple in the vertical oven, how do you go about clamping the lead wires so that the wire will not foul the oven when it goes over the upper sprocket?

Mr. Zirzow: We clamp the wire on each length of the drive chain, and that curves it over the top sprocket. You have to be very careful so that you will not foul your oven especially when the wires go over the top. At 3-ft. intervals we clamp the wires to the drive chain of the oven on the outside links, so the wires will not fall on the big sprocket going over the top. We use them once or twice.

Member: Did I understand you to say that you do not attempt to control the temperatures in that oven too closely?

Mr. Zirzow: We definitely control the temperatures, but what I tried to imply was that the pyrometer settings are meaningless. I could set the oven to bake cores at 500° or 400° F. That does not mean a thing as to the actual temperatures in the oven.

Member: Do you find that much vibration of those ovens gives you trouble?

Mr. Zirzow: On two of our ovens we have vibration, but the trouble we encounter is negligible.

ALUMINUM CASTING ALLOY 3 PER CENT CU : 5 PER CENT SI

Methods for recovery of airplane aluminum scrap have been developed, and have made available an excellent source of material for the production of general purpose aluminum alloys.

O. Tichy
Metallurgist
The National Smelting Co.
Cleveland

IN RECENT MONTHS the literature has contained treatises on the utilization of the enormous amounts of airplane scrap. Technical magazines of late have brought such headlines as "From War Planes to Kitchenware," the articles describing a suggested method of converting the available aluminum into alumina and then by electrolysis back into pure aluminum.

Dr. N. H. Engle[1] referred to this process as one which probably would be as expensive as the production of aluminum from bauxite. One of the possible advantages is that of conservation. However, this would not appeal to those who desire a less expensive source of the metal. In the same article are a few comments about a relatively simple process which consists of careful melting of sorted scrap in a special type of furnace, the end product of which is said to be suitable for cast and wrought purposes.

This latter reference is one of the few to a modern method of reclaiming the useful aluminum alloys from obsolete, battle-damaged and war-weary planes, although it has already been applied on a production scale and useful alloys have been recovered from such scrap materials.

Another reference is given by E. J. Hardy[2] who describes this process as the Navy recently has put it into practice. Some precautions as to procedure are mentioned, which are expected to result in good, useable alloys at a reasonable price.

Airplane scrap was put to good use by the smelting industry during the war when the resultant alloys had to conform to the most stringent specifications. Following is a partial list of alloys, familiar to foundrymen, which were produced and utilized during the war for the production of various types of castings: 4 per cent copper, 2 per cent silicon; 4 per cent copper, 3 per cent silicon; 4 per cent copper, 5 per cent silicon; 4 per cent copper, 9 per cent silicon; and aluminum-copper-magnesium-manganese-silicon alloys containing 5, 8, and 10 per cent of silicon.

Reclaiming Scrap

As has been mentioned previously, the reclamation of airplane scrap has been done on a commercial scale with good results. A sloping hearth is used for this process. Sorting to remove any lead present keeps this element at a sufficiently low concentration so that it does no harm. The aluminum is "sweated" from any iron or other high-melting alloys present, and the latter raked out before they can alloy with the aluminum.

Analyses of several heats of metal produced with this procedure are reproduced in Table 1. These data represent a cross-section of metal recovered from many types of airplane scrap, and it can be seen at a glance that useful alloys can and are produced with such material.

During the war the aluminum industry had to cast alloys of the conventional types due to rigid standardization. The price of an alloy was not too significant a factor then, but peace-time competition places a different aspect on this problem.

For a considerable period the No. 12 alloy, which is basically an 8 per cent copper alloy, was used to produce a large variety of castings of both the sand and permanent mold types. During the war period, due to the large amount of wrought material scrap made available, a large quantity of alloys containing copper and silicon were cast. This type of alloy has better castability and exhibits better mechanical properties than the No. 12 alloy. It can now be readily produced from the

Table 1

ANALYSES OF METAL PRODUCED FROM AIRPLANE SCRAP

Lot No.	*Cu*	*Fe*	*Pb*	*Mg*	*Mn*	*Ni*	*Si*	*Sn*	*Zn*	*Cr*	*Al*
B 943	0.35	0.29	0.01	0.94	0.09	0.14	0.70	0.01	0.05	0.23	97.30
B 944	0.32	0.33	0.01	0.94	0.09	0.09	0.61	0.01	0.05	0.26	97.31
B 946	0.12	0.90	0.01	2.22	0.05	0.01	0.19	0.01	0.12	0.23	96.17
B 951	3.86	0.61	0.09	1.40	0.52	0.01	0.47	0.02	0.22	0.03	92.83
B 957	1.56	0.49	0.06	1.41	0.09	0.03	4.53	0.05	0.23	0.08	91.56
B 959	3.58	0.70	0.44	1.03	0.50	0.03	0.56	0.01	0.14	0.07	92.98
B 962	4.20	0.53	0.10	1.45	0.58	0.01	0.23	0.01	0.02	0.03	92.86
B 963	4.15	0.51	0.16	1.24	0.58	0.01	0.14	0.01	0.02	0.02	93.17
B 965	3.98	0.45	0.06	0.74	0.69	0.02	0.51	0.01	0.06	0.03	93.48
B 970	1.61	0.48	0.03	1.36	0.03	0.01	4.10	0.04	0.05	0.10	92.18

Presented at an Aluminum and Magnesium Session of the Fiftieth Annual Meeting, American Foundrymen's Association, at Cleveland, May 7, 1946.

types of materials recorded in Table 1.

It has been recognized for some time that duralumin is not a good casting alloy. In order to utilize this raw material, two general methods were available to improve its castability. One method was to raise the copper content to 7 or 8 per cent, as in the old conventional No. 12 alloy, and the other was to add silicon which, as is well known, increases the fluidity of most aluminum alloys[4]. Since copper also was scarce during the war and silicon afforded the easier casting alloys, it was logical for the silicon additions to be the first choice.

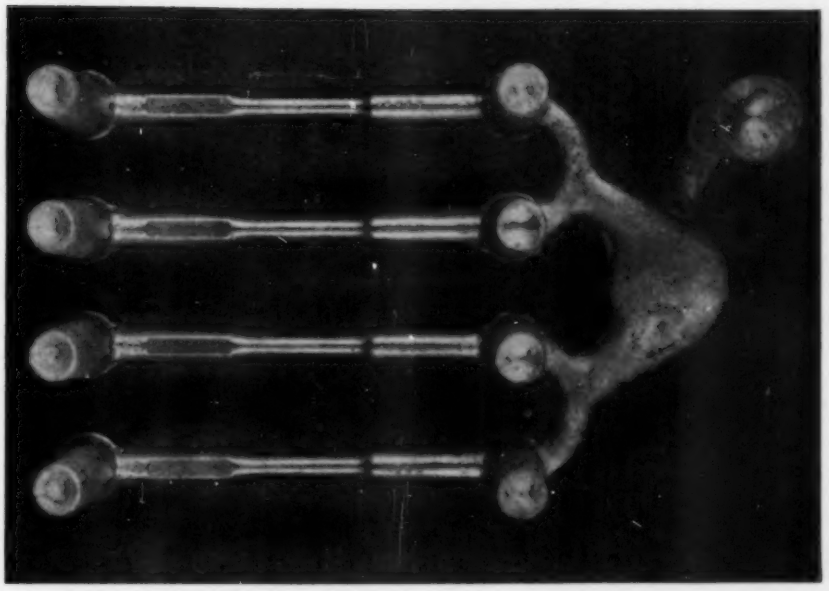
Emergency specifications, such as the AXS series, were developed by the armed services with the aid of specification-writing societies. The limits in silicon content were exceptionally wide and permitted several alloy series within the specifications. Many specific alloys were developed within these series by various users to suit their methods of casting.

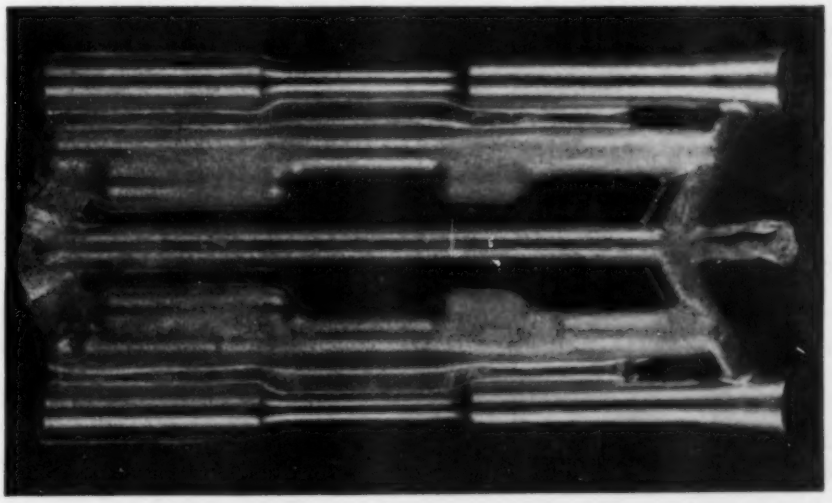
Study Best Composition

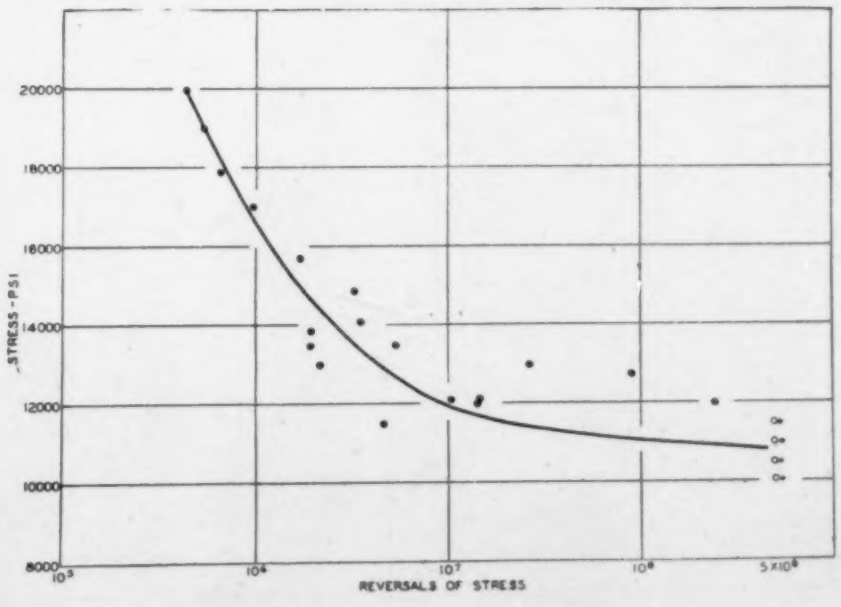
In the production of these various alloys, a field was found which indicated the possibility of interesting properties. Studies were made in various directions to find the best composition which would give an alloy easily produced from existing raw materials.

It was also desirable to have an alloy possessing good casting and mechanical properties without resort to solution heat treatment and still be able to materially enhance these properties by heat treatments if necessary. As a result of this work, an alloy was developed having basically a typical composition of 3 per cent copper and 5 per cent silicon.

Good mechanical properties are developed in this alloy in both the as-cast and heat-treated conditions, and it may be cast in either sand or permanent molds. As will be seen







Reading from top to bottom: Fig. 1—Four-bar aluminum test specimen cast in green sand at temperatures of 1260-1280° F. Fig. 2—Type of permanent mold aluminum test casting. Fig. 3—Endurance limit of sand-cast aluminum alloy—solution heat treated and aged.

Table 2

PROPERTIES OF SAND CAST ALUMINUM ALLOYS

Heat No.	*Yield Strength, 0.2% Offset, psi.*	*Tensile Strength, psi.*	*Elongation, % in 2 in.*	*Brinell Hardness, 500-kg. load*
AS CAST[1]				
X-273 B	14000	26600	3.3	62
X-274 B	14100	26500	3.5	61
SOLUTION HEAT TREATED AND AGED[2]				
X-273 B	18200	33800	4.5	72
X-274 B	18400	33200	4.2	69
355-T 6	25000	35000	3.5	80
356-T 6	22000	32000	4.0	70
STRESS RELIEVED[3]				
X-273 B	16100	29100	2.8	64
X-274 B	16400	28900	3.0	64
355-T 51	23000	28000	1.5	60
356-T 51	20000	25000	2.0	60

[1] One day at room temperature.
[2] 960° F.—12 hr., hot water quench, 300° F.—3 hr.
[3] 440° F.—8 hr.

Table 3

PROPERTIES OF CHILL CAST ALUMINUM ALLOYS

Heat No.	*Yield Strength, 0.2% Offset, psi.*	*Tensile Strength, psi.*	*Elongation, % in 2 in.*	*Brinell Hardness, 500-kg. load*
AS CAST[1]				
X-57 B	15500	30100	3.2	70
X-58 B	14800	30600	3.5	69
SOLUTION HEAT TREATED AND AGED[2]				
X-57 B	18500	38400	6.2	78
X-58 B	17900	39400	7.0	75
355-T 6	26000	43000	4.0	90
356-T 6	24000	40000	5.0	90
STRESS RELIEVED[3]				
X-57 B	19000	33500	2.3	73
X-58 B	18300	34500	2.8	73
355-T 51	24000	30000	2.0	75

[1] One day at room temperature.
[2] 960° F.—6 hr., hot water quench, 300° F.—3 hr.
[3] 440° F.—8 hr.

later, the properties developed are substantially the same as those of AN-QQ-A 376, SAE 322, and ASTM SC 21, which contain basically 1.3 per cent copper, 0.5 per cent magnesium and 5 per cent silicon and shall be referred to later as alloy 1; and AN-QQ-A 394, SAE 323, and ASTM SG 1, which contain basically 0.3 per cent magnesium and 7 per cent silicon and shall be referred to as alloy 2.

In order to obtain properties under conditions similar to commercial practice, the test specimens were cast at a temperature range of 1260 to 1280° F. into green sand molds. The type of casting made is shown in Fig. 1. Average properties obtained with this procedure, consisting of from four to eight individual tests, are listed in Table 2 along with typical values of alloys 1 and 2 taken from the literature[3].

In the as-cast condition this copper-silicon alloy gives tensile strength as high or higher than No. 12 alloy but has much greater elongation. Typical elongation for sand-cast No. 12 alloy is about 1 to 1.5 per cent. In the solution heat treated and aged state (Table 2) this alloy surpasses No. 12 alloy, which can not be markedly improved by such treatment, and it becomes equal or better than No. 2 alloy in strength and elongation, and better in elongation than No. 1 alloy. The only marked difference is found in the low yield strength; No. 2 alloy exhibits higher yield strength and No. 1 alloy still higher.

In the stress-relieved condition the alloy gives a tensile strength equal

Table 4

PHYSICAL PROPERTIES

Heat Treatment	*Appr. Thermal Conductivity*[1]	*Appr. Electrical Conductivity*[2]
As Cast	0.25	27
Solution Heat Treated and Aged	0.27	29
Stress Relieved	0.28	31

Thermal Expansion, in./in./° F. (68-212° F.)	12.2×10^{-6}
Specific Gravity	2.76
Weight, lb./cu. in.	0.100

[1] Thermal conductivity values calculated from electrical conductivity. C.G.S. units are calories per square centimeter per degree centigrade per second.
[2] Electrical conductivity, per cent International Annealed Cu. Standard at 68° F.

Table 5

PROPERTIES OF CHILL CAST (0.05-0.11% MG.) ALUMINUM ALLOYS

Heat No.	*Yield Strength, 0.2% Offset, psi.*	*Tensile Strength, psi.*	*Elongation, % in 2 in.*	*Brinell Hardness, 500-kg. load*
AS CAST				
5112	20600	35000	2.7	84
5116	19400	34600	3.2	80
5122	19700	35000	3.3	79
5125	20300	35200	3.4	78
SOLUTION HEAT TREATED AND AGED				
5112	24100	40800	4.2	91
5116	23700	41500	4.6	91
5122	22300	42400	5.9	88
5125	22400	41600	5.3	89
STRESS RELIEVED				
5112	22900	36900	2.7	81
5116	25200	38200	3.2	83

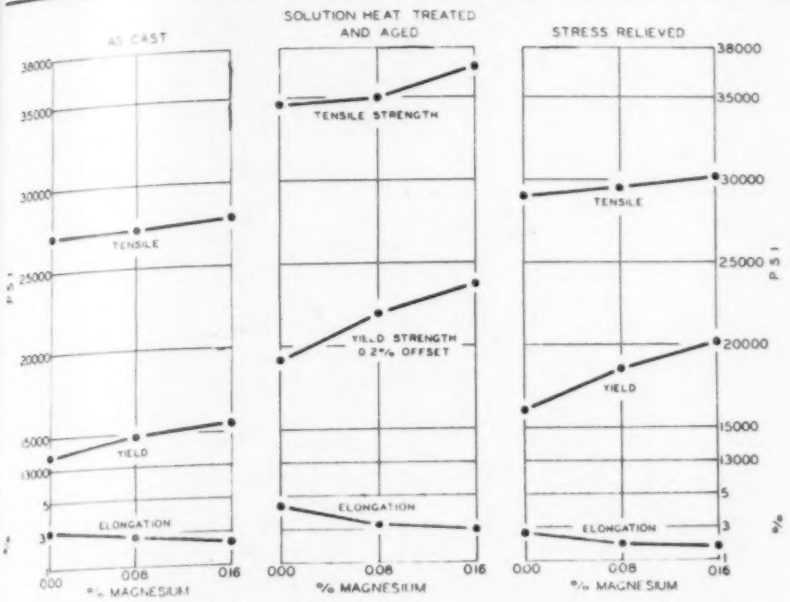


Fig. 4—Effect of magnesium on properties of sand-cast aluminum test specimens.

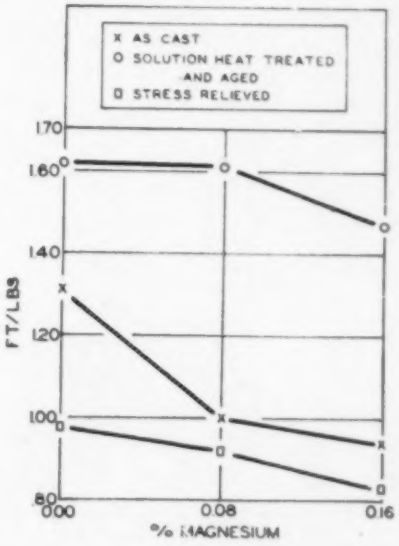


Fig. 6—Effect of magnesium on sand-cast aluminum impact test speciments.

to or greater than that of the No. 1 alloy and greater than that of the No. 2 alloy, while the ductility is greater than in either. Here again the yield strength is somewhat lower. This is easily explained by the larger amount of magnesium present in the latter two alloys. However, the yield strength in this new alloy can be altered, as will be shown later.

Other heats were made using the same melting and casting technique, and permanent mold castings of the type shown in Fig. 2 were made. Average properties of these chill-cast specimens, constituting four to six bars in each case, are shown in Table 3. The as-cast condition develops a good tensile strength with a fair amount of ductility.

Solution heat treated and aged, this alloy has a lower tensile strength and yield strength than No. 1 and No. 2 alloys, but has somewhat higher ductility. In the stress-relieved condition the tensile strength and ductility exceed that of No. 1 alloy, with the yield strength again being somewhat lower.

Fatigue strength values of alloys are considered essential by designing engineers today. Specimens were cast into green sand molds of a design similar to those used for the tensile specimens and tested with R. R. Moore type fatigue machines.

At present, the only test com-

Table 6

ALLOY COMPOSITION

Element	*Per Cent*
Copper	2.5-3.5
Silicon	4.5-6.0
Magnesium	0.15 max.
Iron	1.0 max.
Nickel	0.2 max.
Manganese	0.5 max.
Zinc	0.7 max.
Elements employed to obtain special properties	0.5 max.
Other elements	0.3 max.
Aluminum	Remainder

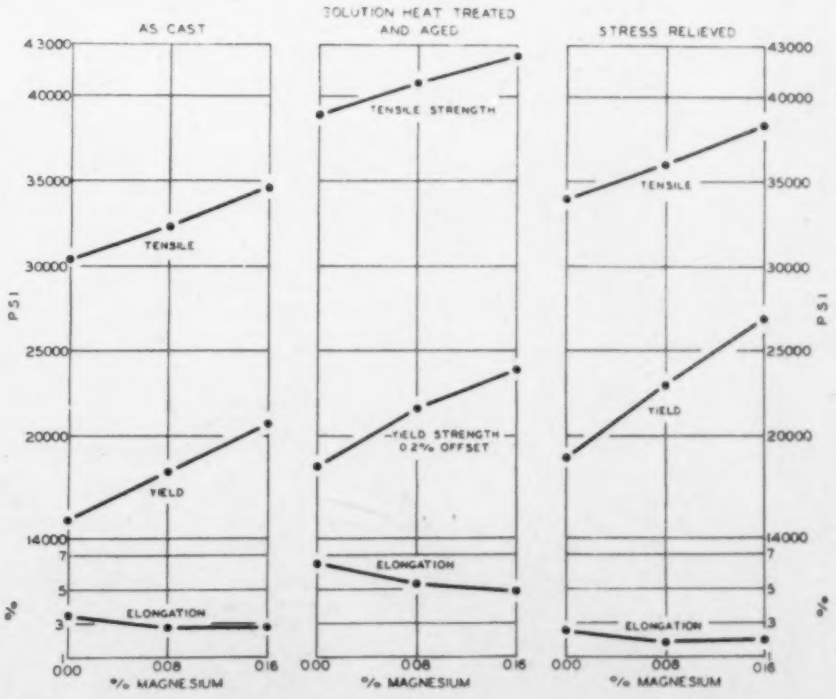
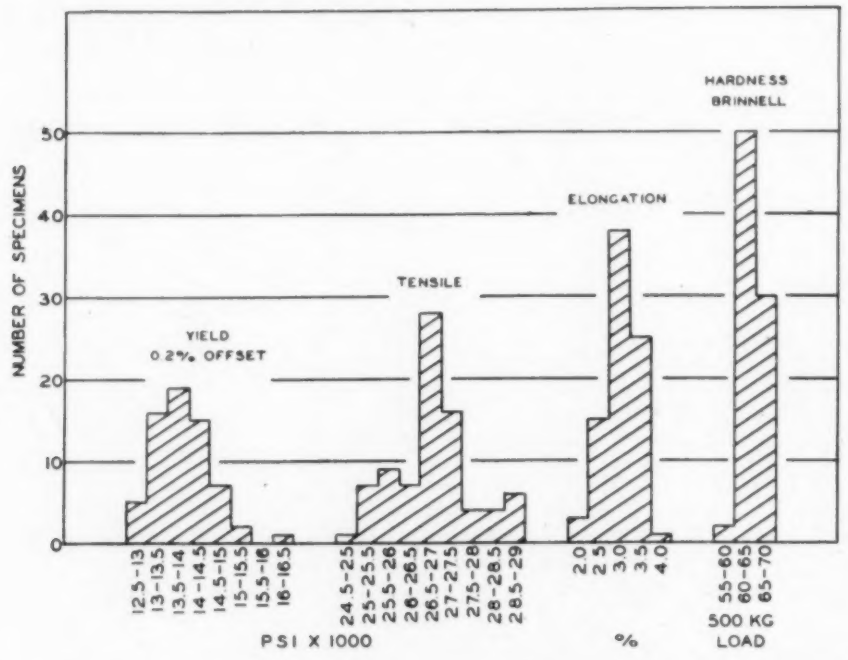
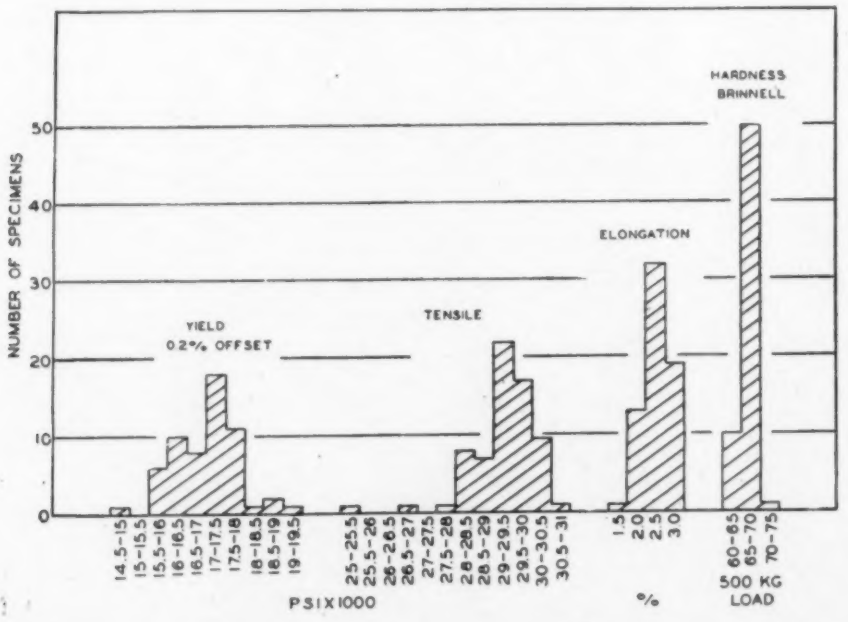
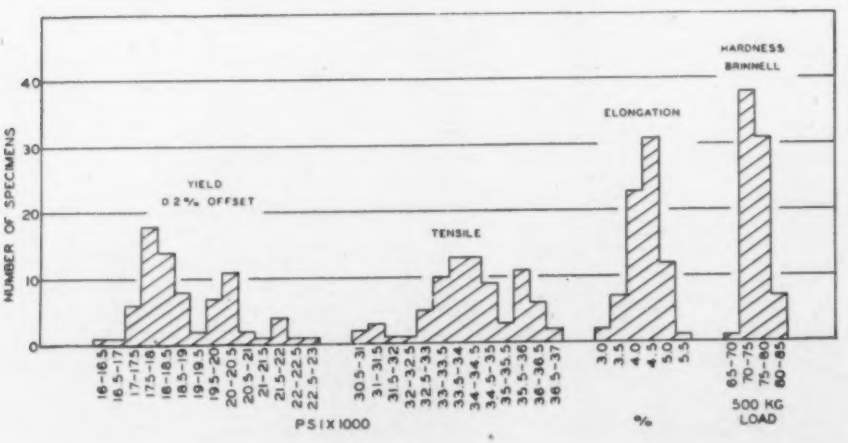


Fig. 5—Charts showing effect of magnesium on properties of chill-cast aluminum test specimens.







pleted was in the solution heat treated and aged condition. The results obtained, which are unusually high for this type of alloy, are given in Fig. 3. Indications of progressing tests are that the alloy also has a high fatigue strength under other conditions.

It is desirable to know some of the physical properties in some applications. A few of the results obtained are recorded in Table 4.

For some applications it is desirable to obtain higher values of the yield strength. As mentioned, this alloy is of the type in which the yield strength can easily be altered by small concentrations of magnesium, so an investigation of this type was made. A quite noticeable effect is obtained in the increase of the yield strength, with a somewhat lesser increase in tensile strength, while an appreciable amount of ductility still remains.

Data obtained from sand-cast specimens are shown in Figs. 4 and 5, respectively. The effect of magnesium on sand-cast impact specimens, machined to ASTM (E 23-41T) Type A Specifications, is shown in Fig. 6. As would be expected in cases where the ductility has been lowered, there is an accompanying drop in impact. In applications requiring higher impact value the magnesium content should be kept at a minimum, while applications requiring a higher yield strength should contain some magnesium.

Mechanical properties were checked on reverberatory heats of approximately 35,000 lb. to procure results expected in commercial practice. In Table 5 are recorded some results obtained from chill-cast specimens with a variation of magnesium content from 0.05 to 0.11 per cent.

Reading from top to bottom:

Fig. 7—Frequency distribution of properties of sand-cast aluminum alloy (as-cast, 13 heats).

Fig. 8—Frequency distribution of properties of sand-cast aluminum alloy (stress relieved, 13 heats).

Fig. 9—Solution heat treated and aged—13 heats. Frequency distribution of properties of sand-cast aluminum alloy.

Data secured from sand-cast specimens of 13 heats containing less than 0.02 per cent magnesium and nine heats containing 0.05 to 0.11 per cent magnesium are shown in Figs. 7 to 12. These data agree favorably with those of the preceding test.

Since composition is only a guide to the foundry trade for the type of alloy used, it should be considered as a problem secondary to such matters as casting technique, castability, and physical and mechanical properties. Therefore, the composition has not been mentioned in detail previously is now given in Table 6.

Summary

In summarizing, it may be said that, due to one of the latest methods for the recovery of airplane scrap, an excellent source of material is available at a reasonable cost for the production of alloys which can be utilized for several purposes. One such alloy has been discussed here. Alloys of this type, because of their good castability and properties, are general purpose alloys, and their usage could eliminate much of the confusion arising in foundries due to the segregation required when several alloys are cast.

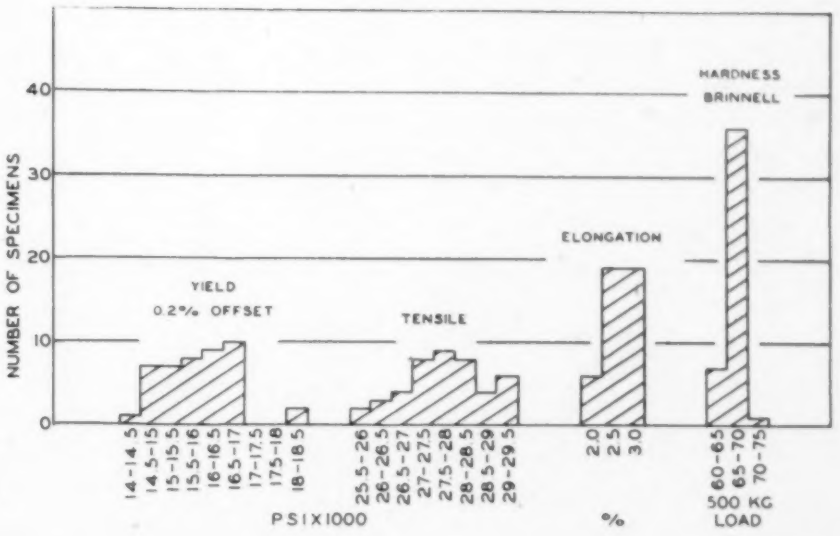
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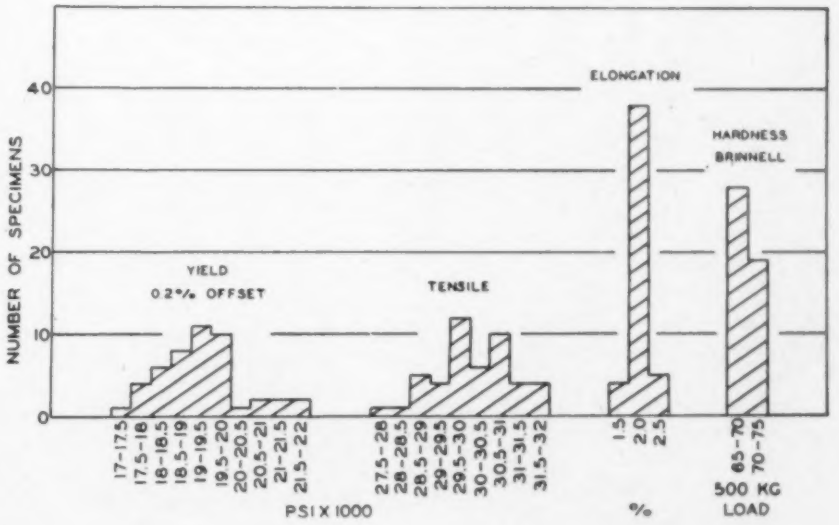
1. N. H. Engle, "The Influence of Scrap Aluminum on the Post-War Market," *Aluminum and Magnesium*, Nov., 1945, pp. 18.

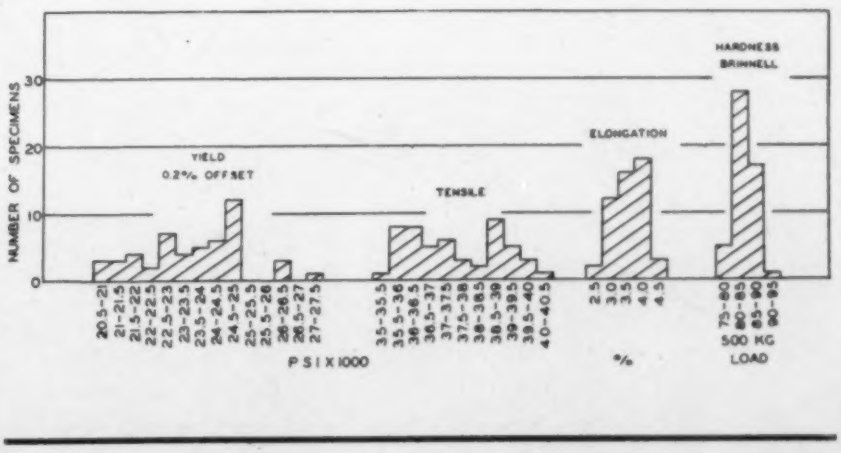
2. E. J. Hardy, "Navy Disposal of Aircraft Scrap to Produce 15 Million Pounds of Secondary Metal," *The Iron Age*, Dec. 20, 1945, pp. 104.

3. "Aluminum in Aircraft," Aluminum Company of America, 1943.

4. W. Bonsack, "Effects of Minor Alloying Elements in Aluminum Casting Alloys," ASTM *Bulletin*, Aug., 1942.







Reading from top to bottom:
Fig. 10—Sand-cast aluminum alloy—frequency distribution of properties. As-cast, 0.05-0.11 per cent Mg—9 heats.
Fig. 11—Stress relieved, 0.05-0.11 per cent Mg—9 heats. Frequency distribution of properties of sand-cast aluminum alloy.
Fig. 12—Solution heat treated and aged, 0.05-0.11 per cent Mg—9 heats. Frequency distribution of properties of sand-cast aluminum alloy.

Table A

SOLUTION HEAT TREATED AND AGED

Yield Strength, 0.2% Offset, psi.	*Tensile Strength, psi.*	*Elongation, % in 2 in.*	*Brinell Hardness, 500-kg. load*	*Days Aged*	*Hours Aged*
AGED AT ROOM TEMPERATURE					
25400	37600	2.9	87	1	—
25500	37100	2.6	87	2	—
26300	37900	2.6	90	3	—
26100	38000	2.6	91	4	—
26500	38700	2.5	91	7	—
27200	38500	2.4	91	14	—
AGED AT 300° F.					
21800	35000	3.3	81	—	2
22400	35600	3.3	82	—	3
23000	36300	3.6	81	—	4
22200	35300	2.9	82	—	5
24800	37200	2.9	86	—	8
27300	38700	2.6	91	—	11
29400	39600	2.1	94	—	13
32100	41500	2.0	97	—	15

DISCUSSION

Chairman: WALTER BONSACK, National Smelting Co., Cleveland.

Co-Chairman: L. W. EASTWOOD, Battelle Memorial Institute, Columbus, Ohio.

C. V. COOPER[1]: We have used this alloy and note that in the solution and aged condition the elongation is higher than in the solution treated condition only. Do you have any comments on this behavior?

MR. TICHY: We have no explanation for this behavior. We have also encountered this same thing as can be observed in Table A, which shows the average results of four to eight bars in each case.

[1]Acme Aluminum Alloys, Dayton, Ohio

MOLDING SAND for BRASS AND BRONZE

L. B. Osborn
Houghland & Hardy
Evansville, Ind.

EFFORTS HAVE BEEN MADE in recent years to duplicate, synthetically, some of the widely known natural sands of repute in non-ferrous foundries. French sand from Normandy is a notable example. Although it is possible on paper to duplicate, from the sands of the Middle West or combinations of Eastern sands, specifications that resemble the famed French sand, few practical foundrymen will agree that actual casting results are equal.

In discussing French sand both older and younger molders almost invariably refer to it as a very fine sand, as if it were a 0 or 00 instead of a No. 2 in the A.F.A. classification (149 average fineness after being ground). The strength of the bond is baffling, especially when it has such flowability that the fairly coarse grains adjust themselves under pressure to produce a smooth casting surface. The sand and bond combination also holds moisture exceedingly long. French sand is an example of a natural product with quality performance that has been acknowledged around the world, and few foundrymen will admit that modern sand science can duplicate it.

Synthetic sands have their adherents, but even the largest brass and bronze foundries are, in general, continuing with natural sands with their merits of easy control, easy patching of molds, good casting finish, prevention of dried and broken edges of molds, elimination of excessive ramming pressures and other detrimental factors that go hand in hand with synthetic sand.

Correlation of Laboratory Work with Foundry Results. During the past ten years of sand research, much effort has been made to find the rhyme and reason for foundry troubles. Many special visits have been made on invitation from different foundries which were experiencing casting troubles. Other visits were at routine intervals, checking results and gaining information of mutual interest. In many instances, sand analyses were made during periods of trouble and again at a later date when better casting results were known to be in effect. As much as we would like to believe that definite results could always be shown by analytical methods, such is not the case. Frequently, only a slight change of permeability, clay content, strength or moisture could be found after the troubles had been eliminated.

These instances were noted especially in foundries where no central sand conditioning equipment is used. In some foundries laboratory results did not confirm the extensive change of results that actually took place in the foundry. Part of this condition probably is due to metal or molding factors other than sand conditions.

A number of the medium and small size foundries undertaking drastic sand changes experienced severe trouble. For example, in such cases where a change of permeability was attempted from, let us say, 8 to 30, the results sometimes were bad, especially in foundries where the individual skill of the molder largely determines the outcome.

In one foundry operating a continuous system, a study of sand conditions was made during a period of abnormally high losses, and again 5 weeks later after large new sand additions had been made, with a resultant large increase of natural clay content. In this plant a 7 to 8 psi. green compressive strength was found to be most efficient. During a period when the consumption of new sand had been cut off almost entirely, the bond strength was regulated by synthetic additions. After

Surveying molding sands for the non-ferrous industries, natural bonded sands continue to be the popular choice for both large and small foundries. Postwar casting production focuses greater attention upon casting finish and the economies necessary in obtaining finish under competitive conditions. Molding and core sands themselves assume a greater responsibility to this end, with many foundries seeking to reduce the use of mold coatings, special facings and other procedures that may have been more profitable under wartime systems of operation. For operating efficiency and simplification, it is advisable to have the whole heap or the entire system equivalent to facing sands; and the fine texture and favorable workability of many natural sands makes this conveniently possible for most foundries.

resuming additions of natural sand, the workability of the sand improved and it was no longer extremely critical to moisture content. Cracked molds and damaged edges due to drying out of sand and rough mold handling were eliminated. A complete analysis of this system sand at these two stages of operation is given in Table 1.

Relationship of Dry Strengths. Brass and bronze foundries using natural sands often have difficulties because of inadequate dry strength. A range of from 25 to 75 psi. usually is suitable, but some of the natural sands that have been so extensively used in the past have lower dry strengths and need reinforcement.

Many sands produced in the belt from southern Ohio to southeastern Kansas have dry strengths within these desirable ranges. The tendency in the plants where dry strength is inadequate is to spray the molds or use special shake-ons and facings. In some cases portland cement is used as a shake-on to increase dry strength. However, such remedies can often be eliminated by proper sand selection in the first place. Most foundrymen will agree that water is one of the necessary evils in a foundry, yet it is one of the molders' favorite working materials. Moisture of heaps on the floor is difficult to control, and influences other sand properties of permeability, green and dry strength, hot strength, retained strength, etc. Most foundries using well selected natural sands obtain good results regardless of large moisture variations in heap sand. The superintendent of a good brass and bronze foundry frankly admitted that the heaps vary from 6 to 10.5 per cent moisture.

Table 1
ANALYSIS OF SYSTEM SAND

Properties	*Additions* — *Synthetic Sand*	*Natural Sand*
Moisture, per cent	6.8	7.6
Permeability	11.4	18.4
Green shear, psi.	1.7	1.9
Green compression, psi.	7.6	7.8
Dry shear, psi.	18.0	21.0
Dry compression, psi	85.0	93.5
Retained on Mesh, per cent		
12	1.40	—
20	0.80	2.60
30	0.80	0.40
40	1.40	1.60
50	4.20	4.00
70	13.40	12.80
100	16.20	20.20
140	14.60	13.20
200	21.20	12.20
270	2.80	4.20
Pan	20.00	15.60
Total Screen	96.80	86.80
Clay	3.20	13.20
Deformation	0.012	0.018
Tensile	3.7	4.2

Relationship of Permeability. The subject of sand permeability has had widespread discussion and study. It was one of the first sand properties to be brought into open discussion and it continues to interest nonferrous as well as other foundrymen, technicians, and metallurgists. During the past 15 years the tendency has been to increase the permeability of sands for non-ferrous practice. Recently, however, some leveling off in this program has been noticed and, occasionally, some reversals.

High Lead Bronzes

The finer sands of low permeability have a great advantage over other sands, natural or synthetic, for high-lead bronzes. H. W. Dietert and his associates have presented interesting information on permeability and back pressure in the mold, and many foundrymen in the course of actual production have learned that there is a limit to the advantages of permeability and that moderation is sometimes the best policy.

A relationship between permeability and the texture and working properties of molding sand exists which undoubtedly accounts for the popularity of natural bonded sands for non-ferrous castings. The typical brass and bronze foundry usually obtains good production and sound castings within the range of 10 to 50 permeability, and 7 to 25 permeability for average aluminum castings. One thing should be remembered, however, and that is that no certain specifications can be set up for successful operation in every foundry.

However, the permeability ranges just mentioned are in wide use in both average size and larger foundries. In some cases the permeability may have to be kept as much as two to three times the maximum previously mentioned; nevertheless, by avoiding unreasonably high permeability better sand handling conditions and casting finish are obtained and, at least where hand molding is concerned, the molder usually is much happier when his sand is not too coarse and open.

Relationship of Green Strength. Green strength should be ample to provide flexible molding conditions, fast production with a minimum of gaggering, nailing, and patching. Light to medium weight bronze castings usually are produced efficiently in heap sand ranging from 6 to 9 psi. green compressive strength. Sands having either natural or artificial bond and which are too sticky or too colloidal usually lack flowability and result in poor casting finish. The tendency of highly colloidal clay bonded sands to dry rapidly often causes difficulty in mold patching.

This is one of the serious handicaps of synthetically bonded nonferrous sands, and some natural bonded sands are more objectionable in this respect than others. It is also desirable to have a relatively low strength per unit of clay in order that a heap clay content may be maintained at about 10 to 16 per cent clay for brass and bronze, or 8 to 12 per cent clay for aluminum castings. This extra clay provides padding and filler between the grains, giving greater moisture tolerance, better casting finish, and reduced thermal expansion to protect against seams, streaks, buckles, and other difficulties caused by sand movement under change of temperature.

Sintering Point Range

Relationship of Sintering Point. The sintering point of sands for nonferrous castings, with the exception of nickel-silver and monel metal, may be much lower than that of sands for ferrous castings. Sands having *"B"* sintering points from 2100 to 2350° F. usually are considered adequate for brass, bronze, and aluminum. This range embraces sands from New York state as well as several Midwestern states.

A number of foundries have attempted to improve casting finish by switching to higher sintering sands for non-ferrous metals. The result sometimes has been an increase in mechanical penetration, veining, fencing and other such difficulties peculiar to the brass and bronze castings. This unsuitability is not due entirely to the sintering point but has a relationship to flowability, hot

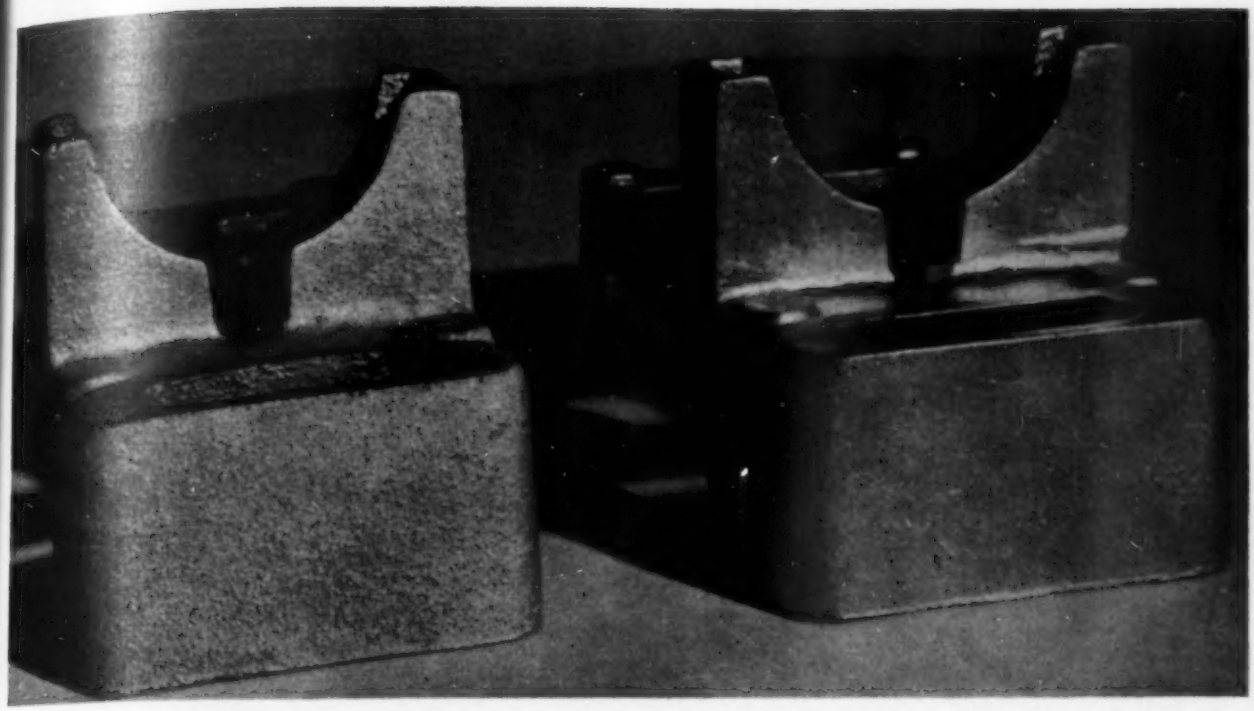


Fig. 1—Aluminum castings made in sand with sands having permeabilities (left) of 20 and (right) 7.

strength, and sand expansion and contraction. In many instances, the higher sintering sands of synthetic composition and a few natural bonded sands are deficient in flowability and have too much hot strength in the temperature ranges from 1000 to 2000° F., which is a handicap of such sands.

Also, the clay content of high sintering sands is more colloidal and less flowable, and this is partly responsible for the failure of high sintering sands to give as good non-ferrous casting finish as lower sintering sands under the same conditions of ramming, tempering, grain size, and permeability. From the point of view of refractoriness, it does not seem worth while to take a sand that is suitable, let us say, for producing steel castings and to pour metal in it at only one-half to three-fourths of the temperatures used in steel foundries.

Relationship of Hot Strength. Years of experimentation on hot strengths of various sands, domestic and foreign, brings out the fact that this feature is important in brass and bronze molding sands. Inadequate hot strength often is responsible for fins and a wrinkled surface. Careful attention must be given to the type of sand used, its hot strength characteristics, and its tempering and ramming, in order to minimize this defect. Additions of silica flour, iron oxide, fly ash, cereal binder, bentonite, and other fillers and binders have been experimented with in an effort to overcome this annoying defect. It sometimes happens that a sand, which in a dilatometer shows cracks and spalling, nevertheless gives satisfactory performance in the foundry. Many foundry supervisors and molders automatically compensate for sand deficiencies without knowing why or how they do it.

Controlled Conditions

For example, one of the popular sands which has an abnormally low hot strength and rapid rate of contraction which is conducive to sand spalling and cracking does, under carefully controlled conditions, perform satisfactorily. Such a sand must be worked at a higher temper than other sands which have better high temperature stability. The low hot strength sands are worked with as much as 10 to 12 per cent moisture content in order to achieve sufficient hot strength and retard collapse. Such excess moisture frequently results in other troubles, pinholes, blows, etc., resulting in rejection of castings before or after machining.

Some foundries have successfully achieved a desirable balance by blending sands to obtain the correct medium hot strength and thus get satisfactory results for brass and bronze castings. The lower hot strengths are more satisfactory for aluminum. Many sands, natural and synthetic, have too much hot strength at 1000° F. for good aluminum performance.

Non-ferrous foundrymen have been puzzled by the different degrees of sand adhesion or penetration in comparing one sand with another. Some concluded that chemical reaction during the casting process is the prime cause. On the contrary, recent experiments by W. M. Ball, Jr.,* give evidence that the penetration and wrinkled surface defects are physical and mechanical. Analysis of the metal comprising the wrinkled surface has been found to be the same as that of the casting proper. This is further evidence that penetration is principally a physical reaction, and the variations in ramming inevitable under hand molding conditions bring about greater penetration difficulties.

Low bonded, flowable sands capable of carrying rather high temper contents often have been observed to be effective under such conditions. The extra water content creates back pressure as well as chilling the metal

*Magnus Brass Div. of National Lead Co., Cincinnati.

Table 2
ANALYSES OF NATURAL SANDS

Properties	*Sand No. 1*	*Sand No. 2*
Moisture, per cent	7.0	7.0
Permeability	18.4	49.0
Green shear, psi.	1.8	2.7
Green compression, psi.	6.9	11.2
Dry shear, psi.	3.0	5.1
Dry compression, psi.	22.0	29.0
Retained on Mesh, per cent		
30	——	0.60
40	0.80	0.60
50	0.60	0.80
70	1.00	7.80
100	6.60	40.60
140	28.60	26.20
200	40.00	4.40
270	3.40	0.60
Pan	2.20	0.40
Total Screen	83.20	82.00
Clay	16.80	18.00
"B" Sintering Point, °F.	2606	2651
Deformation	0.016	0.014
Tensile	6.8	9.8
A.F.A. Av. Grain Fineness	125	82

and inhibits penetration. Recently, in foundries that are more highly mechanized and where sand is rammed to a more uniform density, the appearance of the penetration defect is less frequent.

Advantages of Synthetic Sand. Synthetic sands appeal to foundrymen who want open and permeable sands. The choice of fine grades of bank sands or silicas suitable for brass, bronze, and aluminum castings is relatively limited. Nevertheless, some foundries have insisted on synthetic sands largely for the purpose of maintaining high permeability. This was true during the war by the makers of some aluminum castings and in other war work.

Some aluminum foundries insisted, rightly or wrongly, on a permeability as high as 100, and conventional molding sands ordinarily used in non-ferrous foundries do not cover this range. Therefore, non-ferrous synthetic sands usually are made from the higher priced washed and dried silicas. However, the bonding materials used often have excessive hot strength in the temperature ranges between 1000 and 2000° F., resulting in cracked castings, leakage, seams and buckles, and other difficulties.

Full synthetic sands are also tricky to control and are critical to slight changes of moisture. It is well known that where such sands may work satisfactorily with 4 per cent of moisture, they are not at all satisfactory at 3 or 5 per cent. Rapid drying out of the molds is also experienced under these conditions.

Various expedients have been used to prevent this annoyance, such as chemical additions, oil or cereal binder, and additions of natural sand to synthetic, none of which are as dependable as the proper natural sands would be. Some natural sands have the peaked grain distribution and high permeability which characterizes synthetic sand, and at the same time have clay bond that is not critical to reasonable variations of moisture. The analyses shown in Table 2 indicate two such sands that have a marked resemblance to synthetic sand, and yet are entirely natural sands and with clay bond characteristics capable of satisfactory performance over a moisture range of 5 to 11 per cent. These sands are also suitable for foundries which wish to pursue a semi-synthetic program where natural bonded sand is used with only occasional additions of re-bonding material to conserve and prolong the use of sand.

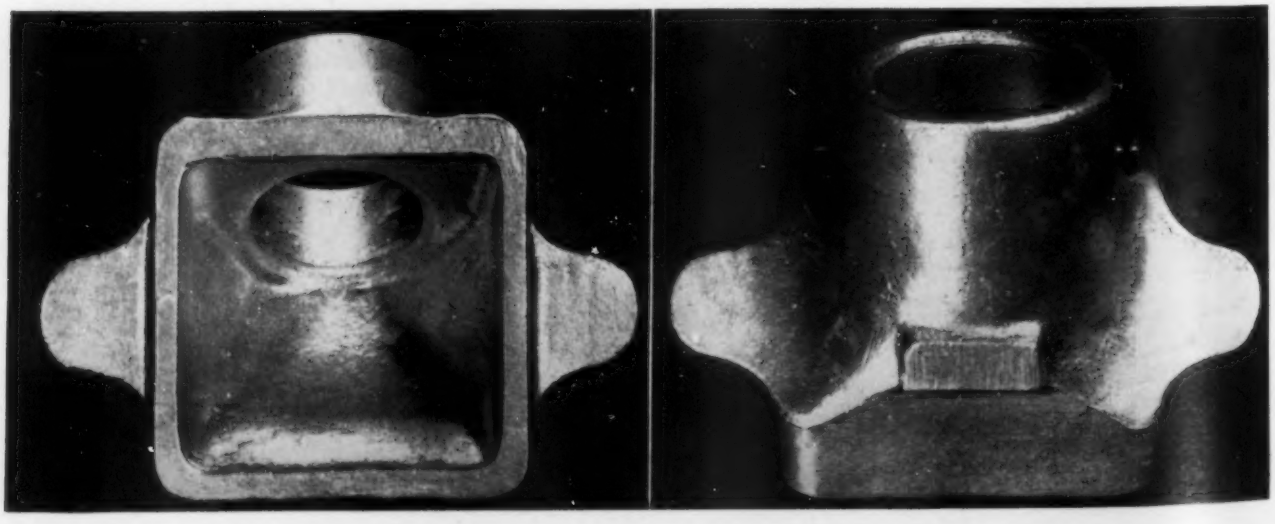
Metal Penetration

Too often foundrymen speak loosely of burn-on or metal penetration as if the conditions were due to inadequate refractoriness of the sand, whereas the true cause often is due to inadequate flowability, causing voids to be left in the mold surface which are open for metal penetration during the casting process. The highly colloidal bond used in synthetic sands is, under practical conditions, likely to impair flowability, bringing about penetration and sandpaper finish, especially in making the bronzes that have an inherent or persistent tendency to flow in between the sand grains.

For this reason, it is necessary to have fairly high permeability and high flowability, permitting hard ramming, or to have a soft bonded sand of sufficiently fine texture, in which case medium ramming usually is used. Following the latter policy, many foundries like the so-called No. 1½ and No. 2 Eastern sands or the soft textured Middle Western sands of 10 to 40 permeability range.

Suitable grain size and good flowability are quite essential in producing smooth surface aluminum castings. Figure 1 (left) shows an alumi-

Fig. 2—Casting made in a high clay content, extremely fine molding sand.



num casting made in a sand of 20 permeability, 7 psi. green compressive strength and 7.5 per cent moisture. Figure 1 (right) shows the same casting made in heap sand of 7 permeability, 6 psi. green compressive strength and 7.5 per cent moisture. The sand for the latter casting is not only finer in texture but is more flowable. Figure 2 shows a casting made in a high clay content, extremely fine molding sand.

The new sand has 270 average fineness and 20 per cent clay content. The heap sand tests 6.5 permeability, 7.5 psi. green compressive strength and 8.5 per cent moisture content. The green sand core and the mold were made of the same sand.

DISCUSSION

Chairman: B. A. MILLER, Baldwin Locomotive Works, Cramp Brass & Iron Foundries Division, Philadelphia.

Co-Chairman: C. LOUIS LANE, Florence Pipe Foundry Co., Florence, N. J.

H. M. ST. JOHN and N. A. TODOROFF[1] *(written discussion):* Natural bonded sands and synthetic sands occupy different but overlapping fields; neither type of sand can be considered universal. In his enthusiasm for the natural sands Mr. Osborn has perhaps failed to do complete justice to the synthetic type.

Natural bonded sands are easier to work and find greater use in small foundries that cannot economically use expensive sand handling equipment. Synthetic sands are more difficult to work and control; they find their application in high-production foundries which can profitably use elaborate mixing equipment. There are many borderline cases.

Mr. Osborn states that brass and bronze foundries obtain good results with sands having a permeability of from 10 to 50. This seems much too wide a range if we are considering only naturally bonded sands. At the lower permeabilities difficulty will be experienced with blows, scabs, pinholes, rat tails and gas seams; at permeabilities near 50 a good finish can hardly be expected if a natural sand without additions is used.

H. W. Dietert and his associates have shown (Mold Surface Gas Pressure, A.F.A. TRANSACTIONS 1944) that permeability measurements must be correlated with the amount of steam and gas generated when molten metal is poured into the mold. The amount of gas generated may be so great that the mold permeability is zero in the early stage of pouring.

Water is the principal source of mold gas and the moisture content of the sand cannot safely be permitted to wander widely no matter what type of sand is used. Nor can the chilling effect of high moisture be ignored, particularly when pouring castings of thin section.

[1]Crane Co., Chicago, Ill.

Moisture control is probably the most important single factor in maintaining the desired properties of a molding sand, whether natural or synthetic. The fact that some natural sands *can* be used at a moisture content as high as 10 or 11 per cent is hardly an advantage if the results of doing so are undesirable.

It is true that certain fine natural sands have no equal if an exceptionally smooth skin finish is desired on small castings. It is also true that, for small castings which are to be polished and plated, a smooth skin finish is of no importance and that these require a high permeability sand of low moisture content if gas porosity is to be avoided.

Poor workability is perhaps the most serious disadvantage of synthetic sands. Intensive mixing and mulling are essential if they are to be used successfully. If the foundry is provided with the proper mechanical equipment for this purpose, the problem need not be a troublesome one.

Synthetic sand molds are more difficult to patch, but their green strength is so high that little patching is necessary. Dry and broken edges can largely be avoided if the proper bond is selected. Synthetic sands admittedly must be controlled more closely than natural sands, but fortunately they lend themselves to control much more readily than do the natural sands.

The foundryman will do well to choose his sand as carefully as he does his metal and to plan his control procedure with equal care. Whether the sand is natural or synthetic (and each has its place) the best possible results will not be obtained without both good sand and good practice.

MR. OSBORN: It is true that my enthusiasm for natural sands is perhaps quite strong. We have to admit that synthetic sands do have considerable value, especially where they are properly used. However, we feel that the natural sands, when they are given the proper selection and given the careful attention synthetic sands must be given, will then perform quite efficiently for brass and bronze castings.

In other words, we feel that natural sands have really a greater place for the non-ferrous metals than they may have for some of the other metals poured at the higher temperatures. The permeability does not have to be so high for natural sands and it is not necessary to watch so closely the chemical purity and the sintering point of the sands as it would be in some other foundry operation.

As to the comment about the permeability range from 10 to 50 being too wide, it is rather wide as a general statement. For the general foundry operating sand in the heaps and hand-ramming or machine-ramming it under average conditions, there are many foundries today that are operating on 10 permeability and having sound castings. I would prefer to see approximately 15 to 20 permeability for most bench work but many foundries are working much lower than that.

As for 50 permeability being too high for satisfactory finish in natural sand, I do not agree with the author of that statement. There are a number of natural sands that will give a satisfactory finish at 50 permeability. I am thinking of a foundry that does a great deal of aluminum-bronze work and they discarded synthetic sand in favor of natural sand of about 90 permeability, which they work in the heaps. After the breakdown of sand in the heaps, they work the heaps at about a 60 or 70 permeability and get what they consider a good finish, in fact, a better finish than they were getting with synthetic sands.

W. M. BALL, JR.[2]: This sand question is a broad question but I still think it has to be discussed in conjunction with what you are going to pour against it. In this non-ferrous group we have probably two or three thousand alloys and two or three thousand conditions and when you try to buy one sand to match those two or three thousand conditions, you are in a quandary.

The variance of the alloy is going to determine the kind of sand to be used and that is where you get into much trouble. The foundryman who is trying to find a sand that is going to cover all these conditions in a brass foundry is asking for the impossible. He might as well make up his mind that he is going to have trouble trying to seek the condition that will fit the thing that he is trying to do.

If he is changing these alloys, he has to think of a sand for each type of work. That is what we try to do. Sometimes we use a rough sand and the next time we have to use a fine sand. You have to get that skin closed up. I have seen the metal go clear through the core. So if you have that condition, you have to get down to that fineness of sand, or the metal will go through it. You have to change the condition of the sand to suit the material that you are going to pour against it.

MR. OSBORN: Mr. Ball's statements I think are very well made and in nearly every foundry except those that are mechanized and with continuous conveying units, I think it is necessary to use two or three different types of sand. When I spoke of the aluminum-bronzes, I referred particularly to that type of metal, that type of casting. That same foundry, however, is also using sand of about 18 permeability for general work and they have an even finer sand for name plates and memorial tablets and other ornamental work.

We must absolutely select the sand for the job. That was another reason why this paper was left rather vague. We do not want to pin ourselves down too closely to any general program, because it is just impossible to do it. It is better to see the foundry's operation, to see what they are doing with the sand, what they expect, what their past experience has been, and take everything into consideration before you make a recommendation as to the type of sand to be used for certain jobs.

CHAIRMAN MILLER: I have discussed

[2]Magnus Brass Div., National Lead Co., Cincinnati.

the term permeability with the steel, gray iron and non-ferrous foundrymen and they all seem to disagree on the word permeability and its application to foundry sand. Mr. Ball put it mildly when he said, we must consider what we are pouring metal against. I think in a few years hence we will have a better knowledge of what permeability means.

H. F. TAYLOR[3]: I would take rather vigorous but good-natured exception to Mr. Ball's statement and turn it around and say that the man who tries to find a different sand for every possible condition that he runs into in the foundry is in for a big disappointment.

We in the Naval Research Laboratory were faced with the problem of finding what we call a universal sand. The reason for that was that in repairing ships at advanced bases, the business of getting a sand for every purpose presented quite a problem in inventory, in shipping, and in procurement. So we tried to find one sand which would serve all these purposes. We compounded this sand and had the molder ram four molds from a single batch of sand. He did not know what would be poured into those molds and we did not know either.

We proceeded to pour aluminum, bronze, cast iron and steel in those four molds that he made up, and they were all excellent castings. To be sure, the aluminum casting was not as smooth as some people would like, but it was a good casting nonetheless, and with a little work in the blending of the sand, not necessarily a mixing of fines, that will give resistance to penetration and smoothness.

It is a blending, the combination of these sand sizes that give this to you, and I am sure that if we took this one base sand and blended it more expertly, we would find that we could get nice-looking, smooth castings. At the Research Laboratory, we used one sand. At the Norfolk Navy Yard they use a single sand for the various methods and I am sure that in time the trend will be towards simplification rather than trying to make the picture more complex.

As to this business of permeability, the permeability of an ingot mold is zero. Yet they make nice, smooth ingots. The reason for that is that you have not put into the mold a lot of things that are going to create gas, and that is the key to a universal sand. Keep your moisture at a minimum, keep your other materials of a nature that will not form a large volume of noxious gases, and when we do know more about this thing, I believe we will be able to simplify the picture.

The natural sands have a place, a good place, in the foundry. I feel we need to know a lot more about them. A fundamental study should be made, more than we have ever made in the past, even more than we have made in the synthetic sands. I feel that one advantage that the synthetic sand has which perhaps the natural sand will have later is that it is a little easier to control at the present time. When it is controlled the same as the synthetic sands and perhaps a new sand facing used regularly, I do not see how you can control the sand that you use over and over. If you used a new sand and natural sand facing and controlled it like you do the synthetic sands, you would have a pretty good deal.

[3]Massachusetts Institute of Technology, Cambridge, Mass.

EXOTHERMIC MATERIALS

Exothermic materials in concentrated form, used as riser additions, provide the heat necessary for proper directional solidification in the casting, reducing riser size and increasing yield.

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J. P. Hickey
and
Michael Bock II
Boston Naval Shipyard
Boston

SHRINKAGE IN CASTINGS, of all metal types, is the most troublesome defect encountered in present foundry practice. The problem of shrinkage, coupled with the attempt to obtain as high a percentage of yield of good casting to metal melted as possible, has resulted in many of the recent developments in gating and risering technique.

Such devices as blind risers, external chills, insulating sleeves, and liquifiers or pipe eliminators have been used with varying degrees of success. Steel companies use insulating fire brick for "hot tops" and, at times, have gone so far as to attempt to produce heat in the top of ingots by applying an electric arc.

Liquifiers Applied

Use of liquifiers on open risers has been adopted as common practice in most steel foundries. Usually they are of the insulating type or the mildly exothermic type, or a combination of the two. Liquifiers are also used in bronze foundries, but in this application the liquifier generally provides insulation against loss of heat rather than lowering the freezing point of the metal through absorption of carbon, as is the case with steel. Many foundries also employ dry floor sand to minimize loss of heat.

Usage of new types of exothermic materials in concentrated form for addition to risers, thus providing great heat at the proper place for producing directional solidification, are dealt with in the present paper. Results of tests are given for steel, monel metal and corrosion-resistant steel. As one of the products of these exothermic materials is metal, the composition must match the metal of the casting.

In exothermic reactions the properly sized powders of oxides of metals react with sized aluminum powder to produce the desired metal or alloy, aluminum oxide and heat. The chemical formulas for three general reactions, i.e., for steel, monel metal, and bronze applications, would involve some or combinations of the following:

$$3Fe_3O_4 + 8Al = 4Al_2O_3 + 9Fe + Heat$$
$$3NiO + 2Al = Al_2O_3 + 3Ni + Heat$$
$$3CuO + 2Al = Al_2O_3 + 3Cu + Heat$$
$$Cr_2O_3 + 2Al = Al_2O_3 + 2Cr + Heat$$

This type of reaction provides the two basic anti-shrinkage aids; production of superheated metal in the riser where heat is needed, and production of a refractory insulating cap in the riser to conserve this heat.

Certain previous experimental work at the Boston Naval Shipyard in making castings entirely of exothermic materials had shown that a small steel bushing could be produced without apparent shrinkage. The casting was made by filling the mold with exothermic powder mixture and igniting the charge in the mold. The surface of the charge was ignited and the reaction proceeded downward through the charge with clean separation of metal and slag.

Figure 1 shows this casting sectioned after removal of the slag layer. It appears that the hot refractory insulating slag on top plus favorable convection currents controlled the cooling rate so that all shrinkage passed out through the top surface without the usual piping effect.

Insulating Slag

The foregoing experiment indicated the desirability of superheated metal plus insulating slag in the riser portions of the mold such as might be obtained by exothermic reactions. This suggested that the mold be filled with ordinary furnace metal to the bottom of the riser, followed by a charge of exothermic mixture to fill the riser. Great heat would be produced at the critical point and the method would be applicable not only to steel but, if proper exothermic mixtures were available, to bronze, monel metal, and any other metal or alloy subject to injurious shrinkage.

However, it was anticipated that the beneficial heat effect accompanying the exothermic reaction might not be so all inclusive as to guarantee solidarity through long feeding paths to points at relatively great distances from the riser. To test this point some of the original experiments were performed on round billets of varying diameter and length.

All of these billets were 20 in. long but divided into groups having diameters of 7, 5 and 3 in., respectively. The slenderness ratio was thus varied, in the case of the last

Presented at a Brass and Bronze Session of the Fiftieth Annual Meeting, American Foundrymen's Association, at Cleveland, May 8, 1946. The opinions expressed in the paper are those of the authors and do not necessarily reflect the views of the Navy Department.

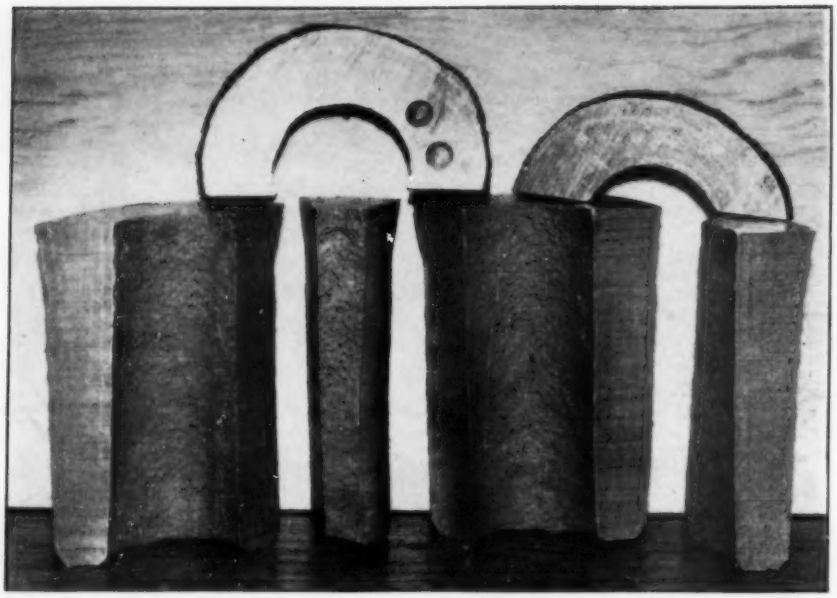


Fig. 1—Steel bushing melted by exothermic process. Slag left on top for insulation to promote feeding. Note that bushing is completely solid.

group the relation was almost 7 to 1, and this was to prove an exacting test. With regard to metal soundness the experiments were not entirely successful as the castings contained some amounts of secondary shrinkage.

However, when the proper amount of exothermic material was added to the risers, approximately 15 per cent by weight, all of the primary shrinkage disappeared. The limitation of feeding a casting from a riser was emphasized by this test. With such a condition directional solidification cannot be readily set up, even with exothermic treatment, and a design change is in order. Up to this point exothermic powders such as described herein serve a most useful purpose.

It is possible to stop pouring a mold when the metal just reaches into the riser and thereby obtain a high yield when the exothermic powder adds a layer of superheated metal at this point. An example of this higher yield is shown in the accompanying table. Both castings were found radiographically sound.

YIELD DATA FOR CAM CASTINGS

Casting Weight with Carbon-Type Liquifiers, lb.	
With gate and riser	117
With gate and riser removed	58
Yield, per cent	49.5
Casting Weight with Exothermic Material in Riser, lb.	
With gate and riser	67
With gate and riser removed	58
Yield, per cent	86.5

Powders of the exothermic type must be compounded chemically to match the metal to which they are to be added. A mixture of iron oxide with aluminum properly graded as to particle sizes and toned down to the proper heat level is satisfactory for steel, but would not, of course, be suitable for corrosion-resisting steel or non-ferrous metals.

For corrosion-resisting steel the metal generated from the reaction of the liquifier must, in most cases, contain substantial percentages of chromium and nickel. The oxides of chromium and nickel are accordingly incorporated in the original mixture in proper balance with the iron oxide.

In the case of monel metal the mixture is made up of nickel oxide, copper oxide, aluminum powder and other ingredients to control the heat level and the viscosity of the slag. A word of caution is desirable at this point as to the explosive character of copper oxide when inadvisedly used alone or in certain combinations with other oxides, particularly aluminum not properly graded.

This situation makes the produc-

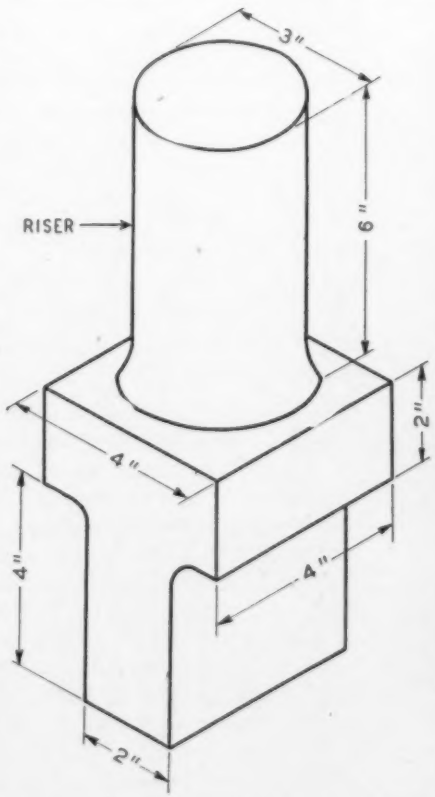


Fig. 2—Diagrammatic sketch of keel block test casting showing dimensions.

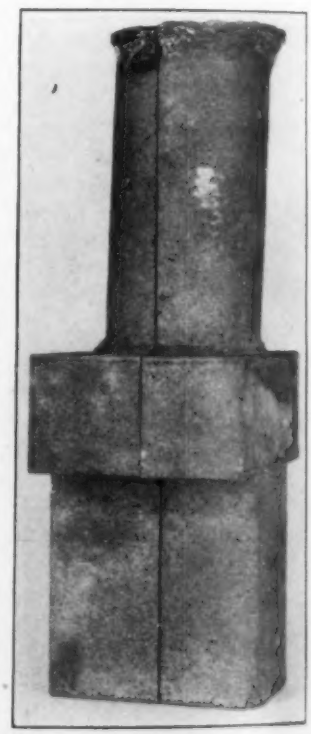


Fig. 3—Keel block test casting with riser. Note line showing where castings were later sectioned.

tion of exothermic powders for copper bronze rather more difficult than is the case with monel metal or steel. However, such exothermic mixtures have been made and can be safely used, with due precautions. No particular problem exists with ordinary steel, nickel-copper (monel) and chromium-nickel mixtures.

Experiments dealing with steel, 20-10 corrosion-resisting steel, and nickel-copper (monel) alloy, are described in the following paragraphs.

Figures 2 and 3 illustrate in half size the pattern and the casting employed in the test. This design of casting has been used recently in other investigations, particularly in the bronze field, to demonstrate the efficacy of gypsum insulating sleeves*.

Ten molds were made, three each for pouring medium steel and corrosion-resisting steel, and four for monel metal. In each case the gate was cut into the back at the base of the riser since in pouring directly down the riser it would have been difficult to stop pouring at the desired level in the riser.

Three castings were poured in medium steel at the same time and from one pot of metal (Figs. 4, 5, 6). One casting (Fig. 4) was poured to the top of the riser and was then covered with an ordinary carbon-type liquifier, following normal foundry practice.

Note that the shrinkage extends the complete length of the riser and into the casting. The second casting (Fig. 5) was poured almost to the top of the riser, the same as in the first casting, and 1½ lb. of exothermic powder added in the riser. This addition amounted to approximately 6 per cent of the weight of the casting and risers combined. The shrinkage then was concentrated at the top of the riser.

For the third casting, Fig. 6, the riser was poured short and 2½ lb. of exothermic material was added. Note that the casting itself is sound. The yield of the casting shown in Fig. 6 is 80 per cent, as compared with a yield of 50 per cent for the castings shown in Figs. 4 and 5. The percentage of exothermic addition in this case was 18 per cent.

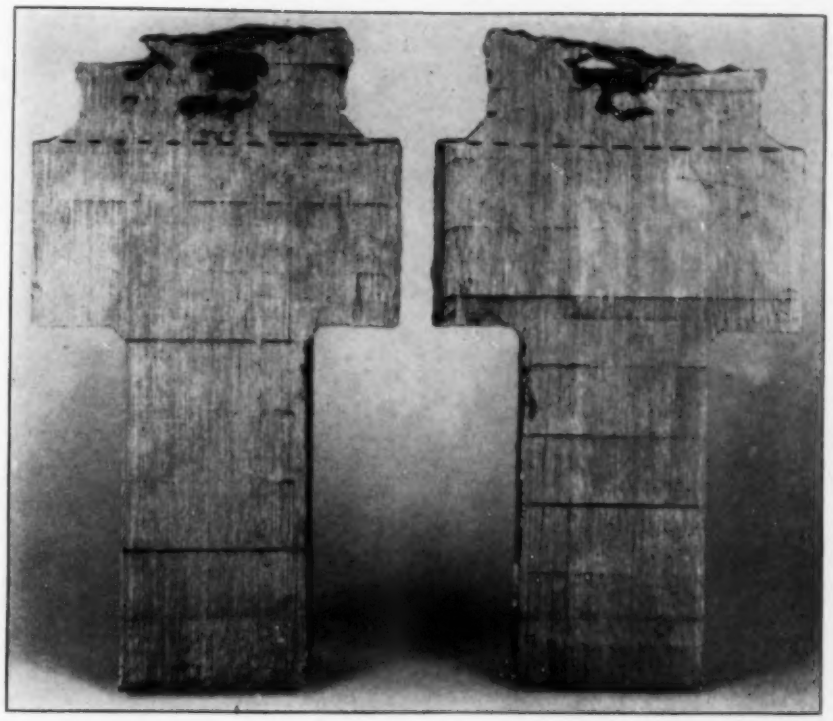


Fig. 6—Class B steel test casting; 2½ lb. steel exothermic powder used; 80 per cent yield of good casting. In this test pouring stopped at bottom of riser near dotted line shown.

*H. F. Taylor and W. C. Wick, "Insulating Pads and Riser Sleeves for Bronze Castings," P. 262.



Fig. 4—Test casting of Class B steel; carbon type pipe eliminator used. Shrinkage extends into casting. Weight of riser and casting, 28 lb.

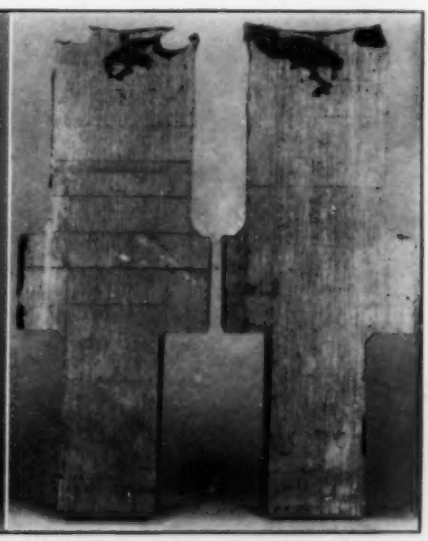


Fig. 5—Test casting of Class B steel; 1½ lb. of steel exothermic powder used in riser. Note lessened shrinkage in top of riser.

Corrosion-Resisting Steel

Similar castings were then poured with 20-10 corrosion-resisting steel. The results were the same. The first casting disclosed a deep, dangerous shrink, while the second and third castings were progressively better. Photographs for this test are not included, the appearance of the castings being almost identical with the photographs shown herein for monel metal castings.

Figures 7, 8, 9 and 10 show the nickel-copper (monel) alloy castings. The first casting, without the benefit of exothermic treatment, developed the deep shrink naturally associated with monel metal, in this case suffi-

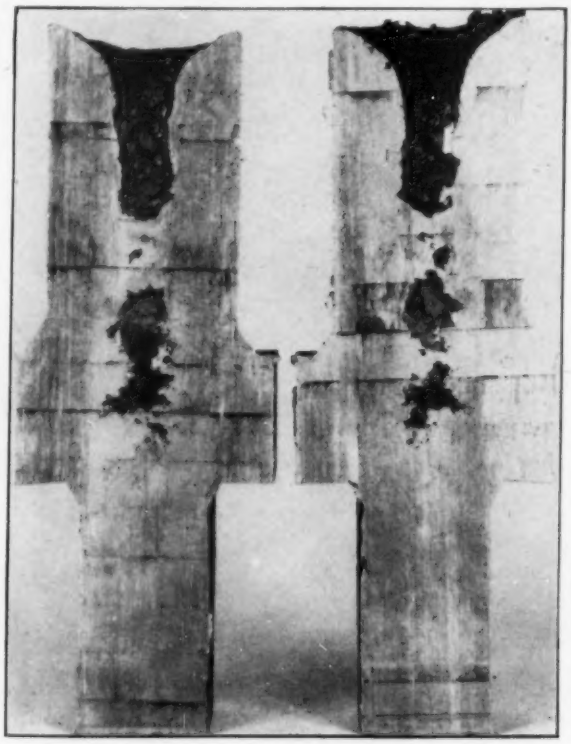


Fig. 7—Nickel-copper alloy (monel metal) test casting; conventional carbon-free pipe eliminator used. Note that shrinkage extends into casting.

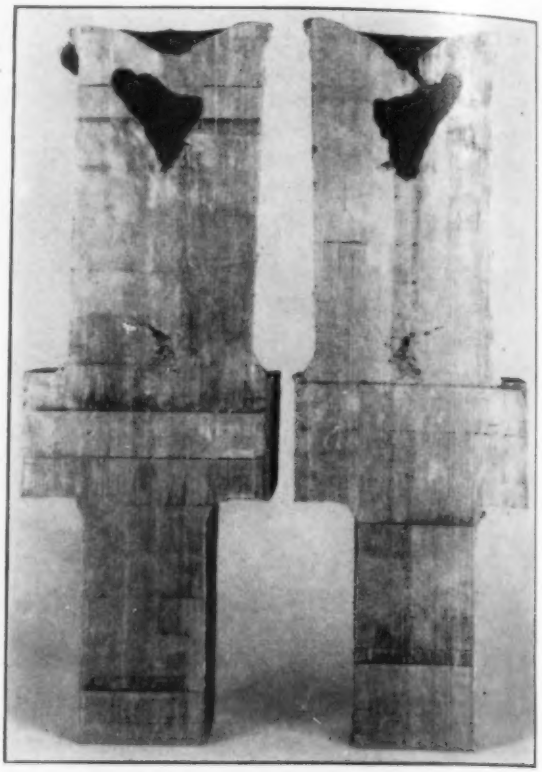


Fig. 8—Test casting of nickel-copper alloy (monel metal); 1½ lb. exothermic material in riser. Casting is sound, although secondary shrinkage is shown at base of riser.

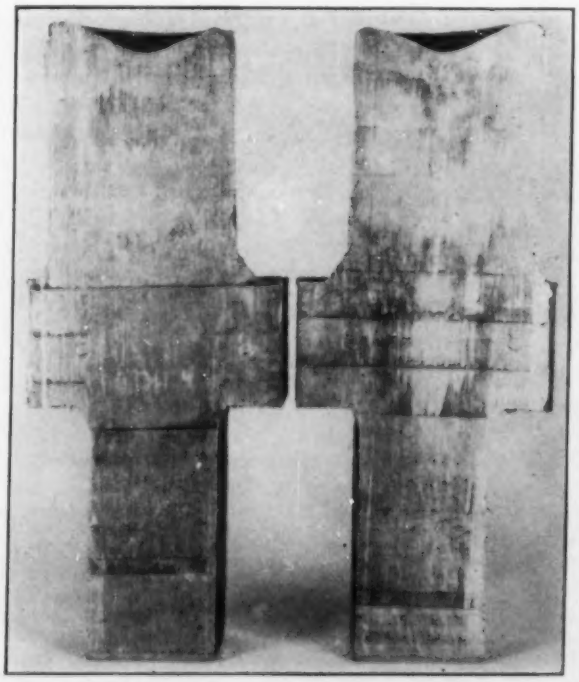


Fig. 9—Nickel-copper alloy (monel metal) test casting; 2½ lb. exothermic material in riser. Note that both casting and riser are solid.

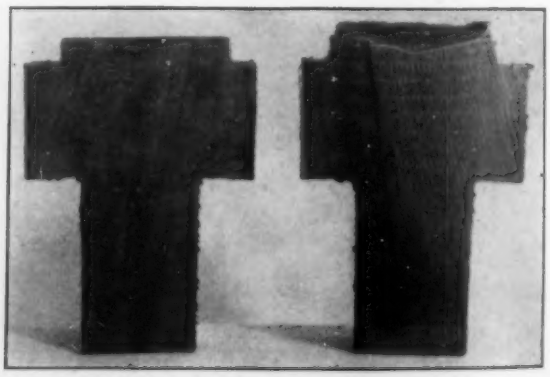


Fig. 10—Test casting of nickel-copper alloy (monel metal); 2½ lb. exothermic material in riser. Note that solid casting has been produced with extremely shallow riser.

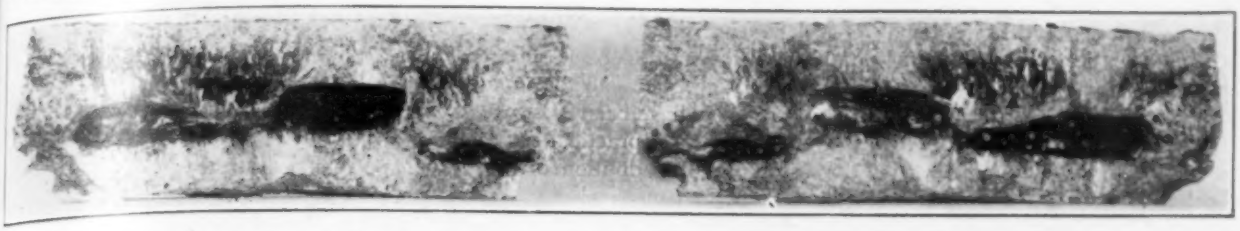


Fig. 11—Refractory slag "hot top" produced as a result of exothermic reaction.

cient to destroy the usefulness of the casting, as illustrated in Fig. 7.

After treatment with 1½ lb. of exothermic material the casting is much improved, but an area of slight secondary shrink persists deep in the riser. Figure 9 illustrates the perfection attained with the 2½-lb. addition. It was decided at this point to try a casting with a minimum of riser, pouring the mold only to the bottom of the riser and then adding exothermic powder.

Figure 10 shows the result—a perfect casting without shrinkage and practically without riser. In the case of these two latter tests particular attention should be paid to the appearance of the riser surface, which suggests that during cooling the level of the molten metal fell uniformly across the riser diameter with little if any inward side freezing. Directional solidification from drag to cope to riser is thus indicated.

Same Composition

A different type of exothermic powder was used for each of the three alloys, i.e., steel, corrosion-resisting steel, and monel metal, so that the hot metal produced exothermically matched the metal of the casting.

In these experiments the exothermic reaction was self-starting from the heat present in the riser metal, and the superheated metal was produced at a temperature of 1000 to 1500° F. above the usual pouring temperature for the metal.

Refractory insulating "hot top" material is practicaily pure Al_2O_3. The material, as it solidifies, usually contains a number of air layers of various sizes interspersed throughout the slag, or one large cap on the



Fig. 12—Nickel-copper alloy propeller casting with riser. Casting is gated in center of hub at bottom. Casting and riser weight, 450 lb.

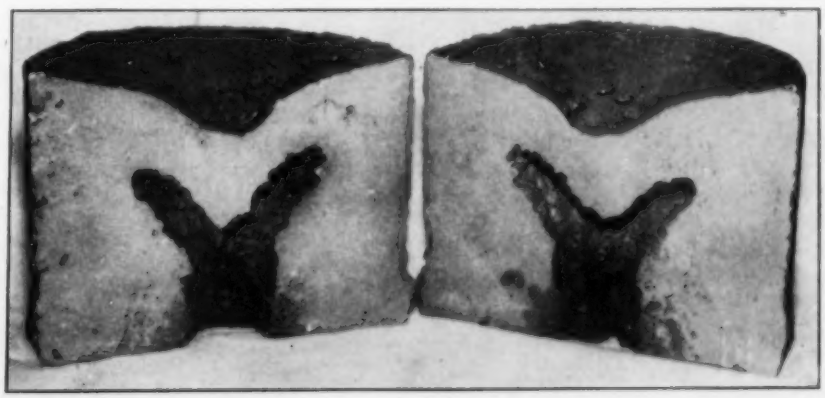


Fig. 13—Sectioned riser from propeller casting showing amount of shrinkage when conventional carbon-free liquidizer is used. Riser, 9½-in. diameter.

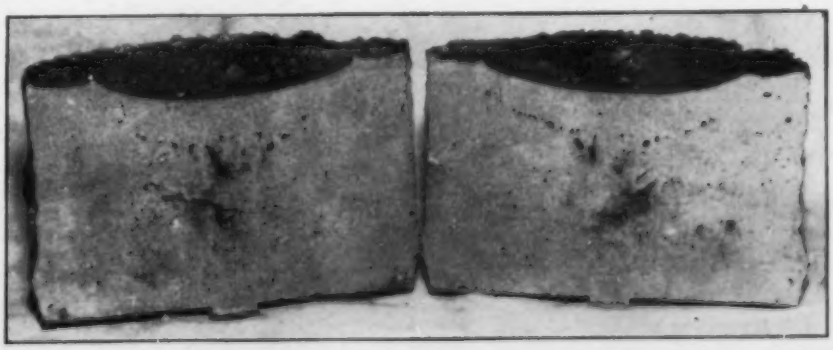


Fig. 14—Sectioned riser from propeller casting; 25 lb. nickel-copper exothermic material used on riser. Note slight amount of shrinkage.

riser. For best results a slag layer over 2 in. thick is desired, although in many instances 1½ in. may be quite satisfactory. Figure 11 shows a sectioned piece of refractory "hot top."

In applying exothermic powders, the success of the operation depends upon using a sufficient amount. In experimental work conducted to date, additions have ranged from 5 to 25 per cent of the weight of the casting and risers combined, the normal amount being about 15 per cent. It appears that the amount of exothermic material will be about the same for all risers of the same diameter.

It is thought in this connection that charts can be prepared showing riser size versus amount of exothermic material. From preliminary work, this method of calculating agrees closely with the percentage basis aforementioned.

The suggestion that exothermic material be used as an anti-piping compound is not new in the case of steel, but is believed to be new in the case of non-ferrous alloys. In the case of steel, recommendations in the literature of some years past as to additions to be made now appear to have been inadequate. For this reason, a practical development and general adoption of the use of exothermic material may have been delayed.

Shrinkage Reduced

Application of exothermic powder to a nickel-copper (monel) alloy production casting is presented as an additional example (Figs. 12 to 14). Figure 12 shows a propeller casting weighing about 270 lb., with a 9-in. diameter riser attached. This was poured by the two methods, first using good normal practice employing the usual protective liquifier, and then with nickel-copper exothermic powder in the riser. The difference in amount of shrinkage can be readily noted from the photographs of the sectioned risers, Figs. 13 and 14.

Figure 13 shows that the shrink defect was 2 in. in diameter and extended deep into the casting. In the case of the casting shown in Fig. 14 the shrink has almost disappeared, but not entirely. The slight shrink may be attributed to the somewhat "skimpy" use of only 8 per cent by weight of the exothermic powder. This test, therefore, again suggests the desirability of using adequate additions.

Conclusions

1. A new method has been demonstrated for making certain types of castings sound and free from shrinkage by applying large amounts of exothermic powders in risers.

2. Exothermic additions to risers beyond those formerly used have produced these results by guaranteeing casting directional solidification.

3. The key to exothermic treatment for castings lies in sufficiency and in the knowledge that there is a minimum percentage quantity necessary to do the job.

4. Cost features have not been considered in this article; these would vary on individual applications and would be influenced by increased yield, cost of the treatment, the necessity for quality, and many other factors.

METALLURGICAL ASPECTS of CAST IRON REPAIR WELDING

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J. A. Blastic
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General Motors Corp.

ALTHOUGH REPAIR WELDING of cast iron has been carried on for a number of years, considerable confusion still seems to exist as to what the necessary factors for correct welding procedure are and what actually takes place in the weld and adjacent cast iron during this repair operation. Some of the procedures which have been used in the past and have been developed by cut-and-try methods are not correct in all details; and, although the essential factors may be sound, some of the items involved might be not only useless but actually harmful. Recognizing this condition, the authors outlined a test and study program in an endeavor to clarify the limitations of cast iron welding and establish more functional procedures.

In reviewing the work to be done, the authors decided to limit the scope of this paper to the welding of gray cast iron by use of cast iron, nickel, or monel filler metal. While it is recognized that brazing, soldering, and the various carbon-arc techniques are also often used, the authors have tried to concentrate on the methods listed in that they are most widely used, and in the belief that the findings would be helpful even in the interpretation of those processes not included.

As a further limitation, it was decided that, in order to keep the paper within reasonable limits, only one composition of gray cast iron parent metal, two compositions of the gas welding cast iron filler metal, one composition of the cast iron arc filler metal, and one composition each of the bare and coated nickel and monel materials should be used.

A test program for the purpose of establishing the limitations of cast iron welding and developing welding procedures which will produce desired physical properties in welds and heat-affected zones is outlined by the authors. The scope of the study was limited, for reasons of practicability, to the welding of gray cast iron by use of cast iron, nickel, or monel filler metal, and further limited in respect to compositions of materials.

A program for the arc and gas welding with the cast iron filler metal was planned with the following preheat and postheat conditions:

1. Weld at 70° F., or room temperature (no preheat or postheat).
2. Preheat and postheat at 500° F.
3. Preheat and postheat at 1000° F., corresponding to usual practice.
4. Preheat and postheat at 1400° F., recognized as being higher than usual practice.

The monel and nickel electrodes were applied with generally accepted practice involving no preheat or postheat.

Test plates were cast under production conditions and were all from one heat of iron. (See Table 1 for chemical compositions of test plates.)

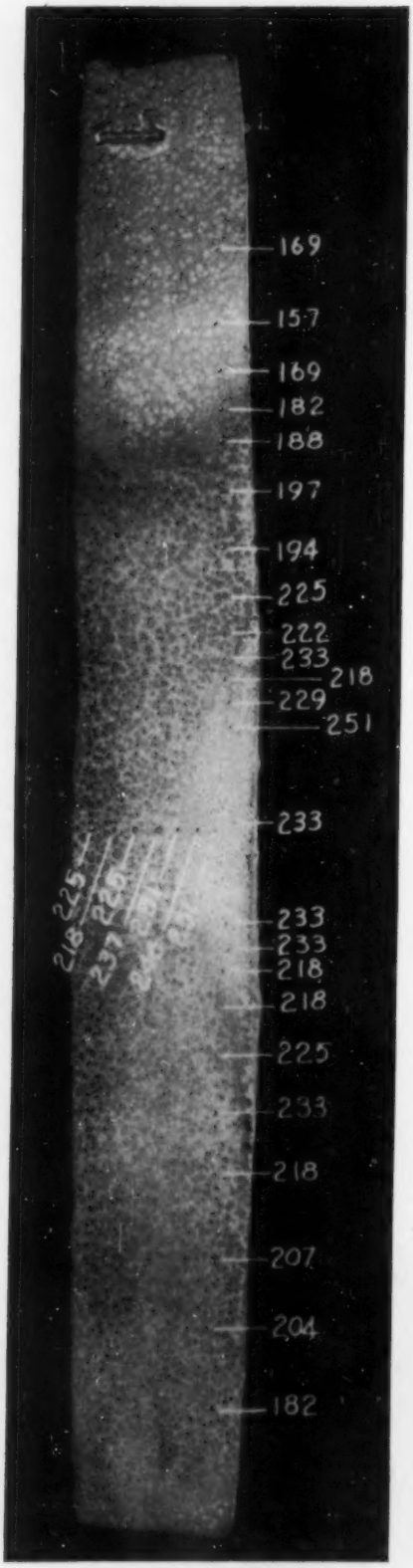
Table 1

CHEMICAL COMPOSITIONS OF MATERIALS UTILIZED

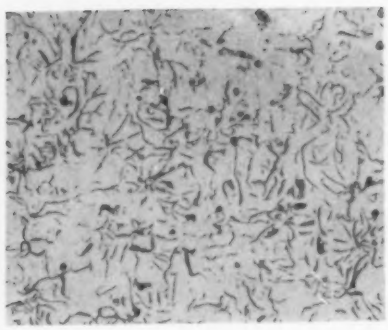
Element	Test Plate Analysis, per cent	Rod Analysis — Gas Rod No. 1, per cent	Rod Analysis — Commercial Gas Rod No. 2, per cent	Pad Analysis: Commercial Arc Rod, per cent
Total Carbon	3.30	3.63	3.42	3.49
Graphitic Carbon	2.9	2.78	2.91	3.46
Combined Carbon	0.39	0.85	0.51	0.03
Manganese	0.66	0.60	0.55	0.77
Phosphorus	0.149	0.60	0.61	0.90
Sulphur	0.199	0.09	0.08	0.022
Silicon	2.32	3.25	3.07	3.33
Chromium	0.16	0.14	—	0.05
Nickel	0.22	0.19	—	0.06
Molybdenum	0.02	0.02	—	0.11
Copper	0.10	0.10	—	—
Vanadium	0.025	0.05	—	—

NOTE: The nickel electrode material used was commercially pure quality of 98 per cent nickel, remainder impurities; the monel electrodes were essentially 30 per cent copper and 70 per cent nickel.

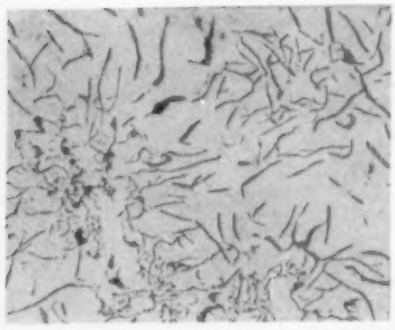
Presented at a Gray Iron Welding Session of the Fiftieth Annual Meeting, American Foundrymen's Association, at Cleveland, May 10, 1946. The authors are connected with the following divisions of General Motors Corp., Detroit: J. M. Diebold, GMC Truck & Coach Div.; J. A. Blastic, Detroit Diesel Div.; J. A. Griffin, Pontiac Motor Div.



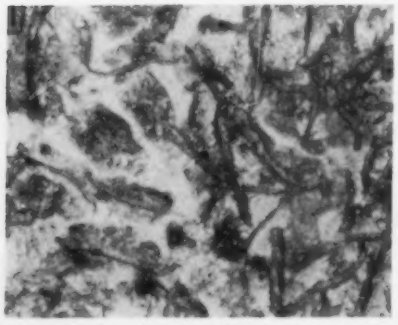
Macrosection. Vickers Brinell (10 kg.).



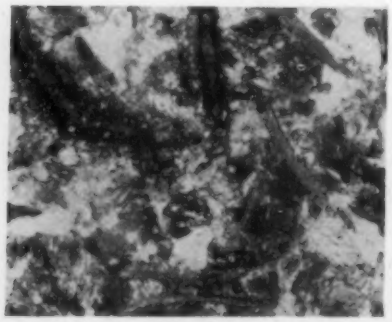
Weld. 100X



Heat-affected zone. 100X



Weld (nital etch). 500X



Heat-affected zone (nital etch). 500X

Fig. 1—Gas-weld specimen with rod No. 1, at room temperature showing hardness traverse and representative microstructure at weld and heat-affected zone, both showing pearlite matrix with steadite and carbide areas.

These plates were cast to the same size, ⅜x3x20 in., and were then cut into pieces 7 in. long for each test condition. Holes of ⅜-in. diameter were drilled half-way through the plates to serve as flaws. In the case of the coated monel and nickel specimens, it was found to be expedient to counterbore these holes in order to give better condition for fusion at the bottom. Preheating and postheating were carried on in a small laboratory-type furnace, with the welding time being held to an absolute minimum.

Postheating practice was carried on in the following manner: the 1400° F. specimens were furnace cooled to 800° F.; the 1000° F. pieces to 500° F.; and the 500° F. specimens to 200° F.; the remainder of the cooling was done in still air. The welded specimens were then cut across the filled flaw and macrosections with hardness traverses were taken of each weld, using Vickers machine with a 10-kg. load. Microstudies were made both for graphite distribution and microstructure.

Tensile test bars (0.505-in. dia.) were made, utilizing sections of mold feeders of the same material. Bars with arc and gas cast iron filler metals were made with double "V" joint preparation for the weld, which was placed crosswise of each specimen.

Since these tensile specimens were small, preheat and postheat for the gas welding was done with the torch, but furnace preheat and postheat of 1000° F. was considered necessary for the arc-welded bar. In all cases, the welding was done by the same operator under as nearly identical conditions as possible.

Gas Welding Specimens. Comparing the hardness traverses of specimens subjected to the first three heat treatments first, it will be noted that the specimens welded at room temperature (Figs. 1 and 2) show that the highest hardnesses are found not in the heat-affected zone but in the welding metal itself, with a peak hardness of 270 Vickers as compared with 250 Vickers for the heat-affected zone.

Studying the microstructure of these specimens, it will be noted that weld metal showed a medium to fine pearlite matrix with some small areas of steadite and carbides. The heat-affected zone was of a

rather heterogeneous consistency, as might be expected. Again the medium to fine pearlite formed a matrix, with some carbide and steadite areas and small amounts of ferrite making up the remainder. The parent metal was found to be of a medium coarse pearlitic structure.

Examining the hardness traverse of the 500° F. specimen in Fig. 3, it can be seen that the hardness ranges are about the same as those of the 70° F. specimens (Figs. 1 and 2). Furthermore, the microstructures closely approximate those of the specimens at room temperature.

At 1000° F. preheat and postheat, the hardness range still remains at about the same as shown in Fig. 4, although upon examination of the microstructure a slight reduction in the amount of carbides and a beginning of ferrite formation around the graphite flakes in the heat-affected zone will be noted.

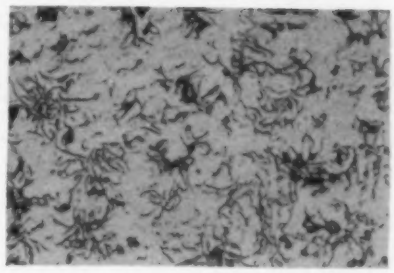
Summarizing the information on these first three ranges on gas welding, the structure and hardness has remained quite constant throughout.

Arc Welding. As might be expected, the 70° F. arc-welded specimen shows a considerably higher hardness range than the corresponding gas-welded specimen (Fig. 5). Again, note that the weld metal with a high hardness of 370-381 Vickers is considerably harder than the heat-affected zone (315 Vickers). Turning to the microstructure, it is readily apparent that the deposited metal and the heat-affected zone are much more sensitive to the absence of preheat and postheat than were the gas-welding specimens.

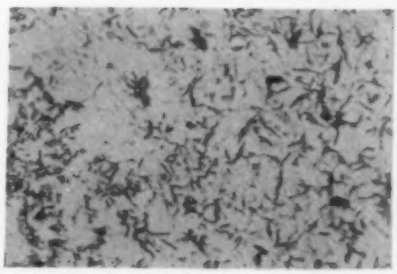
Incidentally, note the fine graphite pattern of the arc-weld metal (Fig. 5).

The weld metal shows a martensite matrix with steadite areas tending to form a network. The heat-affected zone shows a mixture of medium pearlite and platelike martensite and small carbide areas. The parent metal shows a pearlitic matrix, as in the gas-welded specimens.

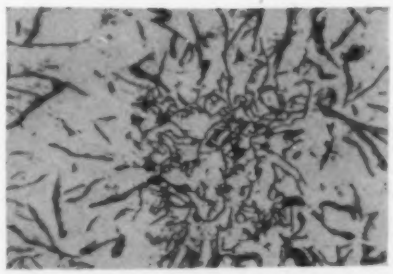
At a 500° F. preheat and postheat an appreciable reduction in the weld metal hardness is noted (Fig. 6), but not much change in the heat-affected zone. Again the weld metal shows a martensite matrix with a steadite network, and the



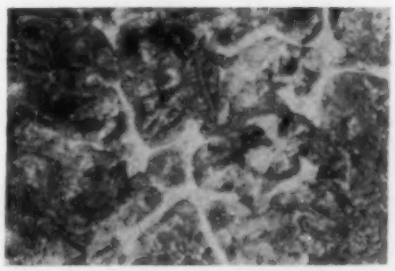
Parent metal. 100X



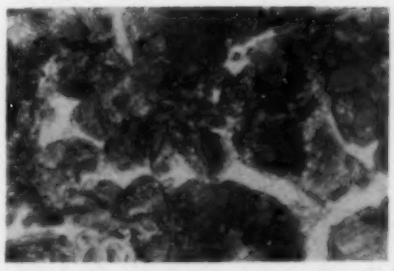
Weld. 100X



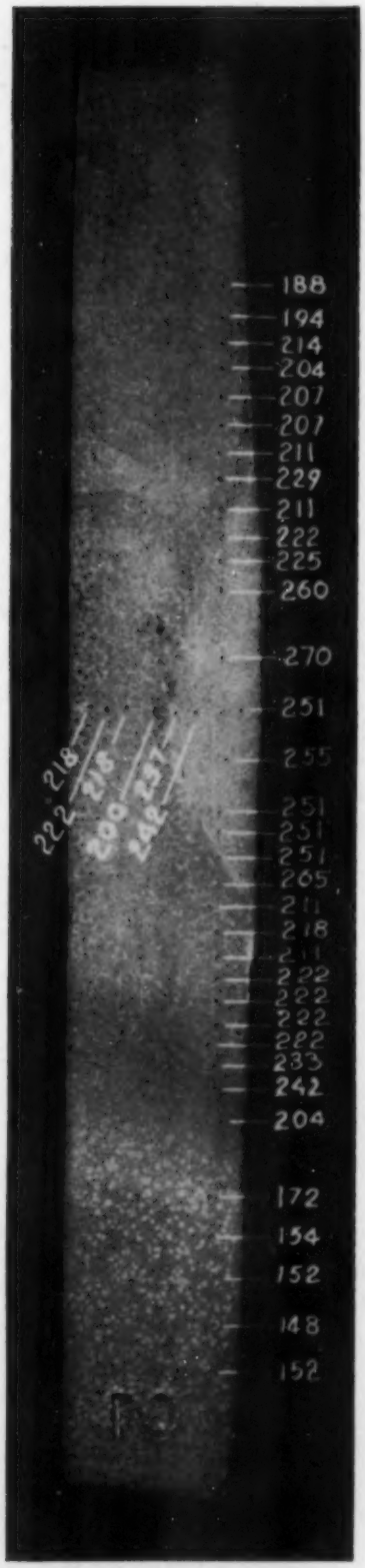
Heat-affected zone. 100X



Weld. 500X

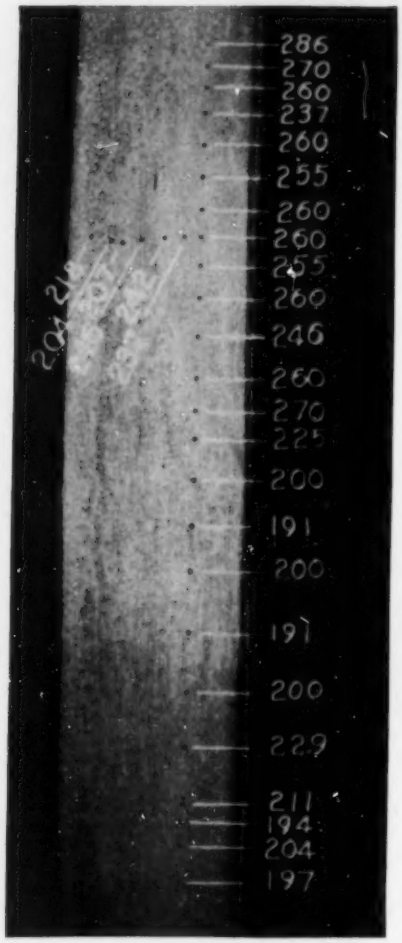


Heat-affected zone (nital etch). 500X

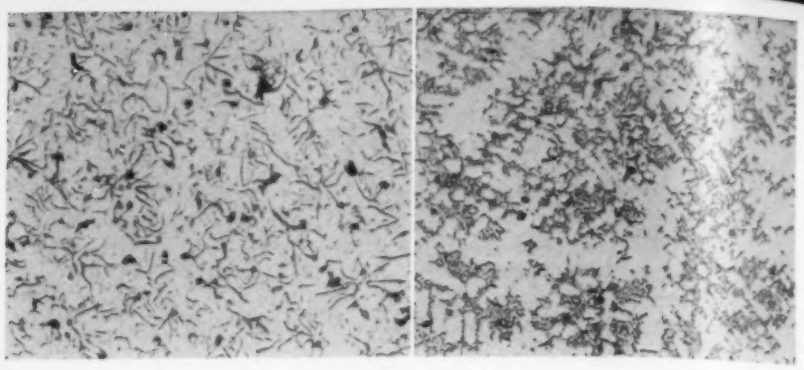


Macrosection. Vickers Brinell (10 kg.).

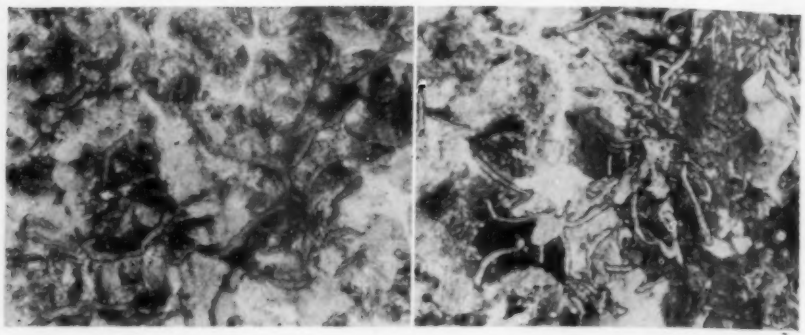
Fig. 2—Gas-weld specimen with rod No. 2 at room temperature showing similarity of structure and hardness with the same condition.



Macrosection. Vickers Brinell (10 kg.).



Weld. 100X Heat-affected zone. 100X



Weld (nital etch). 500X Heat-affected zone (nital etch). 500X

Fig. 3—This gas weld, made with 500° F. preheat and postheat, shows a similar hardness range and microstructure to those welded at room temperature.

Macrosection. Vickers Brinell (10 kg.).

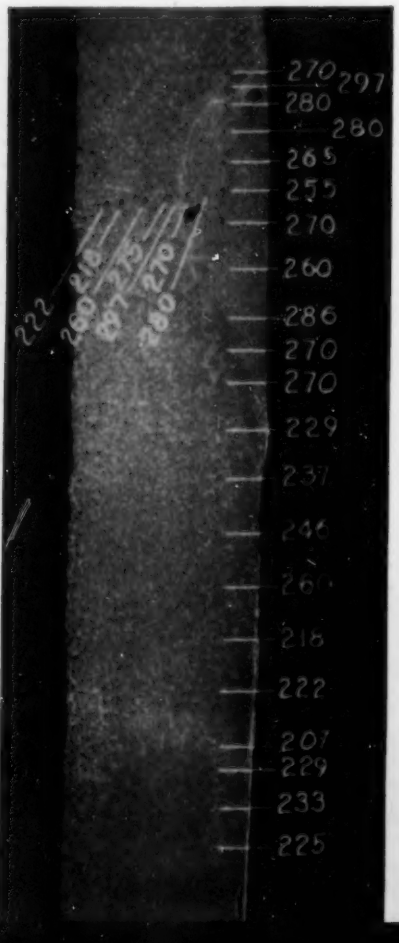
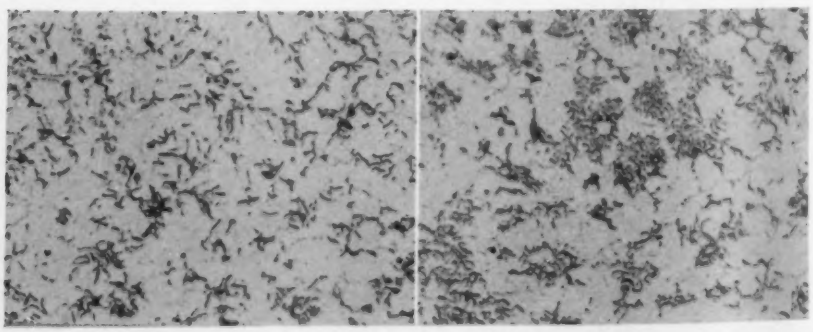
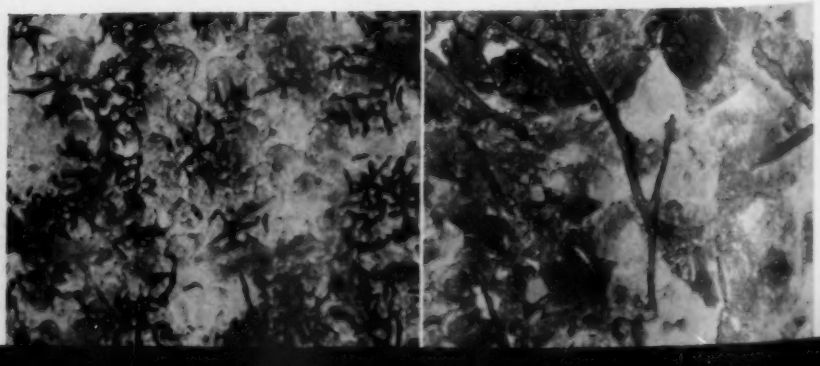


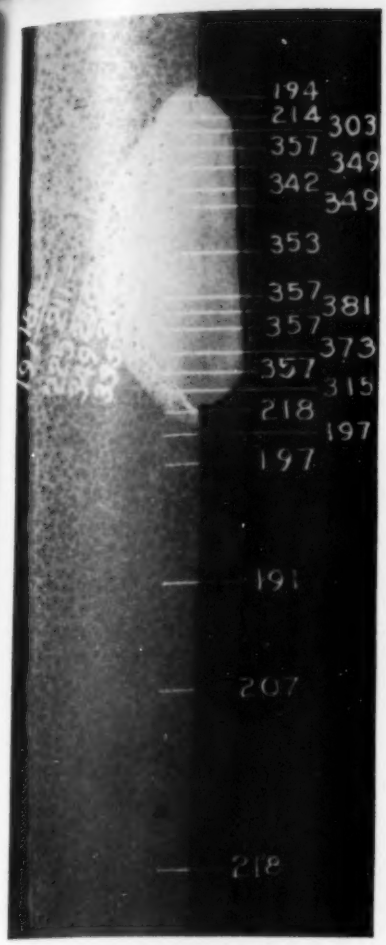
Fig. 4—The gas weld with 1000° F. preheat and postheat again shows a hardness and structure similar to Figs. 1, 2 and 3, but note a slight decrease in carbides together with an indication of ferrite formation around the graphite flakes.

Weld. 100X Heat-affected zone. 100X

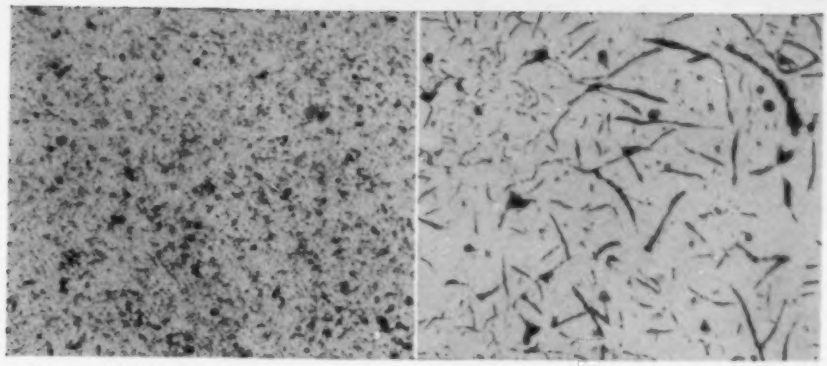


Weld (nital etch). 500X Heat-affected zone (nital etch). 500X

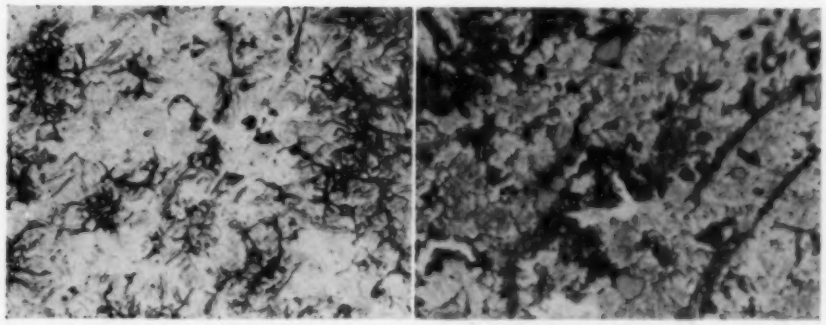




Macrosection. Vickers Brinell (10 kg.).



Weld. 100X Heat-affected zone. 100X



Weld (nital etch). 500X Heat-affected zone (nital etch). 500X

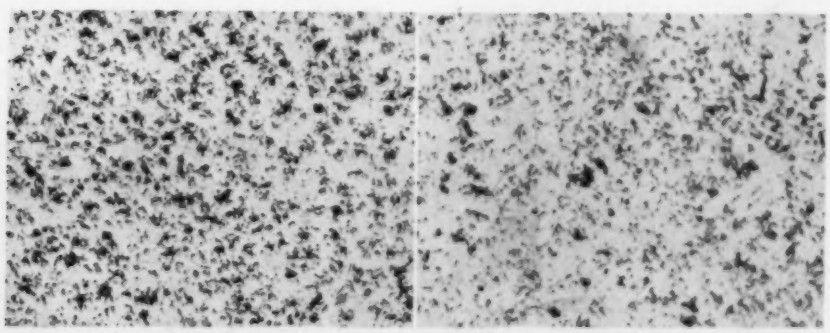
Fig. 5—Arc weld with cast iron electrode welded at room temperature. Note high hardness and martensitic structure in weld and heat-affected zone.

Macrosection. Vickers Brinell (10 kg.).

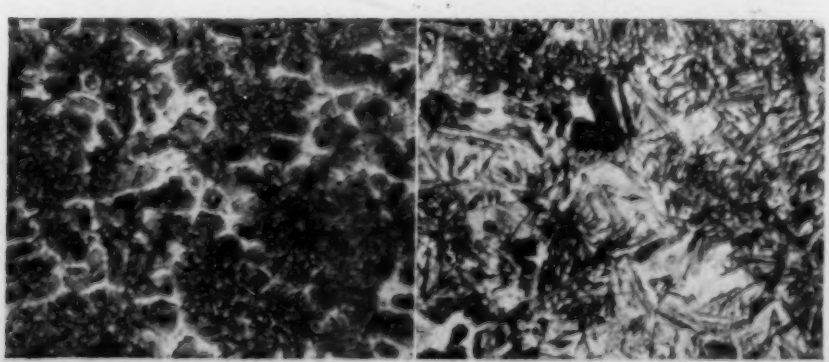


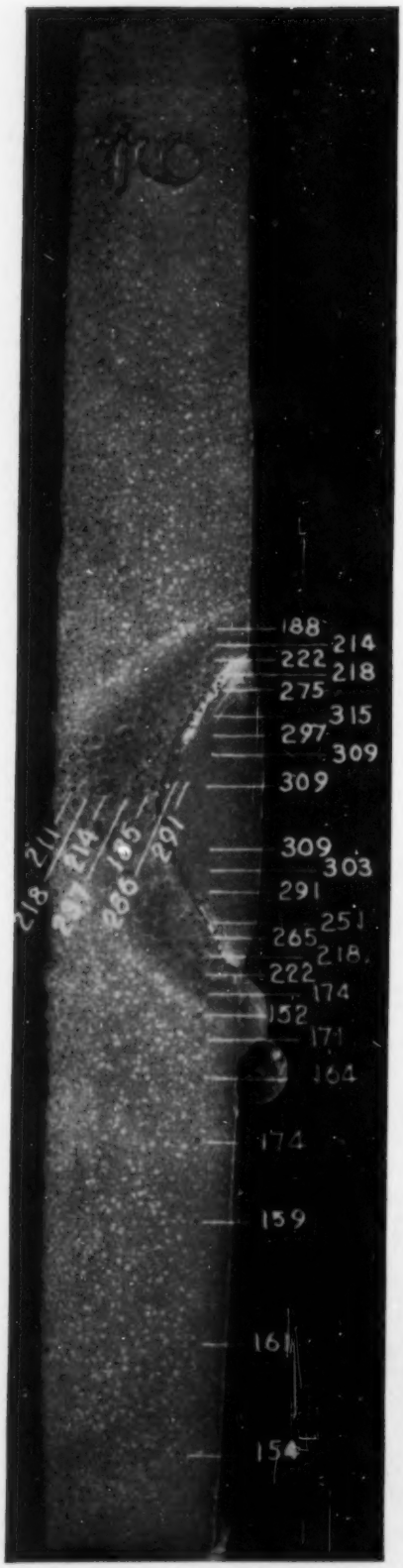
Fig. 6—The arc-welded specimen with 500° F. preheat and postheat, while showing some improvement in weld hardness still has a martensitic matrix in weld and coarse acicular structure in heat-affected zone.

Weld. 100X Heat-affected zone. 100X

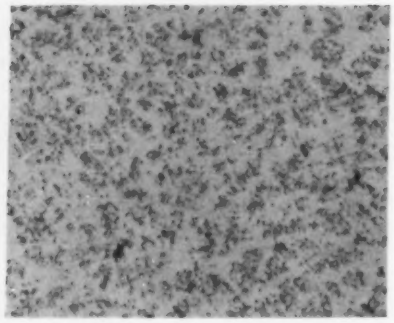


Weld (nital etch). 500X Heat-affected zone (nital etch). 500X





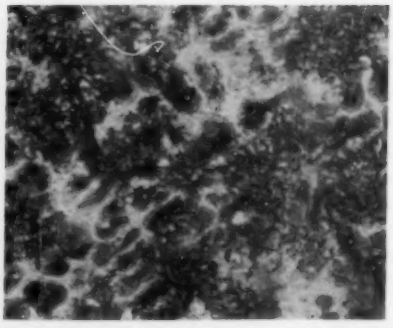
Macrosection. Vickers Brinell (10 kg.).



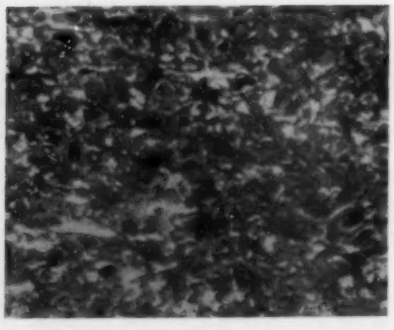
Weld. 100X



Heat-affected zone. 100X



Weld (nital etch). 500X



Heat-affected zone (nital etch). 500X

Fig. 7—When arc welded at 1000° F., hardness of the weld and heat-affected zone is definitely reduced and the structure shows a change from a martensitic to a sorbitic matrix.

heat-affected zone shows a coarse acicular structure with some steadite and carbides present.

The 1000° F. preheat and postheat arc-welded specimen shows a better condition in hardness traverse (Fig. 7) than the other previously discussed arc-welded specimens. Although averaging about 40 points Vickers higher than the comparable gas-welded specimen (Fig. 4), it now has an appearance more closely agreeing with the gas-welded condition.

Upon examination of the microstructure, it will be noted in both the weld metal and heat-affected zone the martensitic matrix has been replaced by a sorbitic structure, but with the steadite network still present in the weld metal and carbides in both weld metal and heat-affected zone. This steadite network in the arc-weld metal undoubtedly is due to the high phosphorus of this electrode. It will also be noted that in the heat-affected zone are the beginnings of ferrite formations around the graphite flakes.

Due to the small size of the specimens, the conditions noted for the arc-welded technique might be more nearly representative of conditions found on large castings than the gas-welded specimens since some preheat and postheat effect occurred during gas welding. The structures for the arc-welded specimens, both at 70° F. and 500° F. are too hard for good machinability and are not desirable from a strength standpoint.

While the hardnesses of the best of these specimens are still somewhat higher than those usually associated with good machining practice, no machining difficulties have been reported from the authors' production machine shops, although a large number of such repairs have been machined in the past.

Arc and Gas welding with 1400° F. Preheat. Studying the conditions of both arc- and gas-welded specimens which were subjected to 1400° F. preheat and postheat, it will be noted that the hardness range dropped severely not only in the weld and heat-affected zone but also in the parent metal, and that the heat-affected zone as such is practically eliminated (Fig. 8). This indicates that in spite of lower and more desirable hardnesses some structural change has taken place

throughout the test specimen.

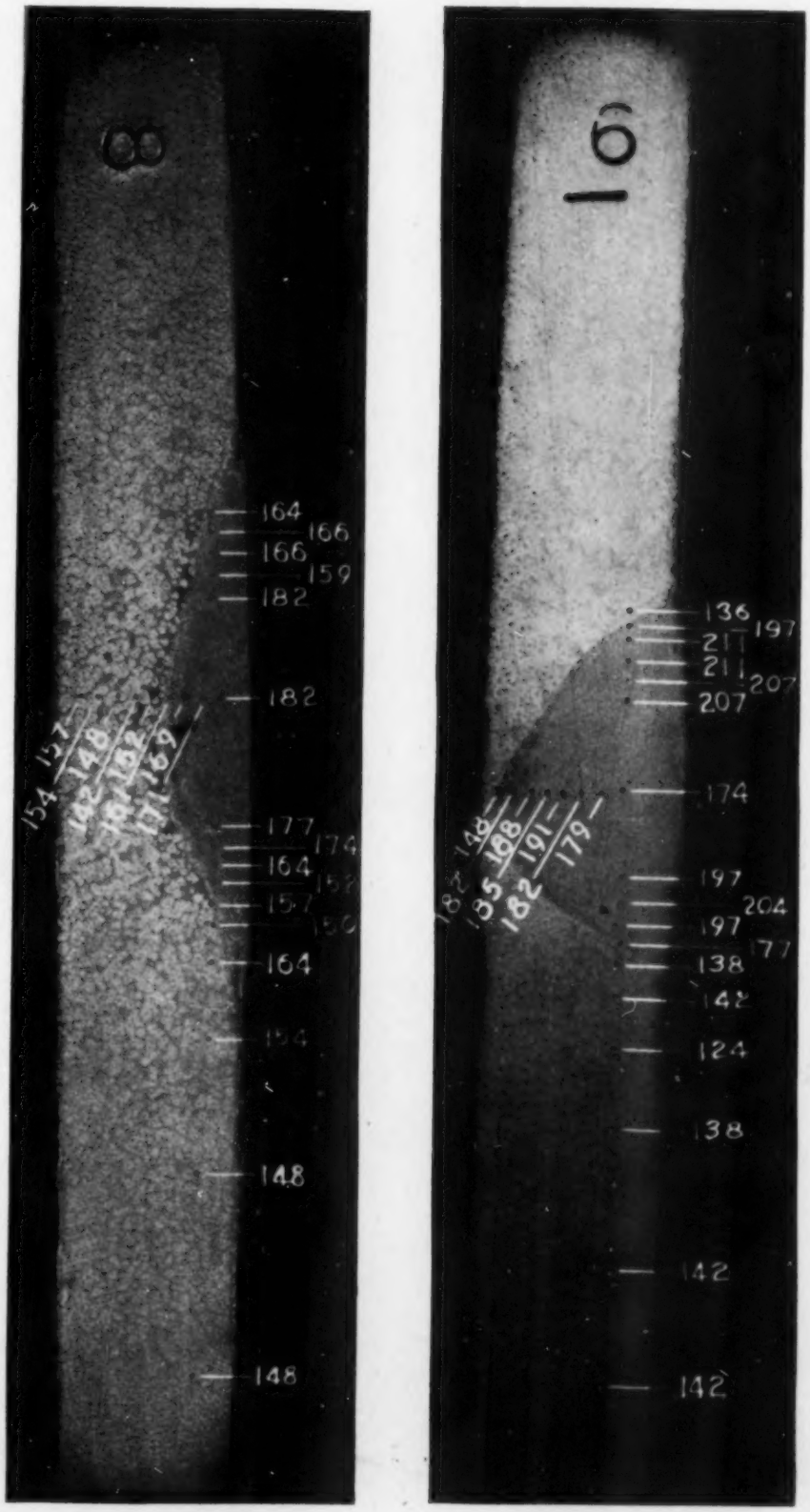
Microstructure of the gas specimen shows (Fig. 9) a ferrite matrix with some coarse pearlite near the graphite flakes and the continued presence of steadites and carbides. In the case of the arc-welded specimen, the ferrite matrix again is present, as well as the steadite network in the weld metal. The heat-affected zone showed a ferrite structure with small amounts of pearlite and some carbides. The parent metal in both cases showed a ferrite matrix with some areas of pearlite.

The dissimilarity between the two types of welding with this heat treatment is small, and in neither case is the structure obtained desirable. Obviously, the amount of combined carbon in these structures is low and the iron involved would have poor strength characteristics. Furthermore, these heat treatments which were held to minimum time length would be much shorter in duration than those which might be expected on a complete casting, and therefore this transformation on large castings would be even more complete.

Reviewing the trends in these studies, the gas welding conditions found at 70° F., 500° F. and 1000° F., all showed a reasonably desirable structure and hardness range without much apparent difference between the various heat treatments. However, at 1000° F., the beginnings of ferrite formation around the graphite flakes and the tendency toward reduction in the amount of carbides present apparently resulted from a slower rate of cooling from temperatures above 1200° F. brought about by the higher preheat and postheat.

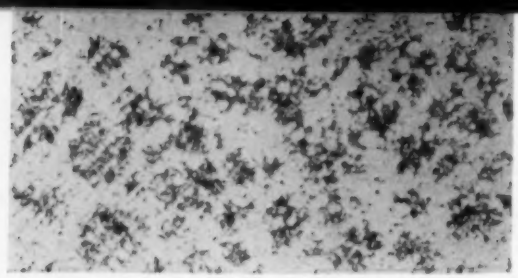
This also indicates that even at this temperature the pearlite is beginning to break down into ferrite, which tendency effected a complete change-over at 1400° F. While a more desirable machining hardness might be obtained by a higher preheat and postheat, reduction in combined carbon would have to be watched closely in this range between 1000° F. and 1400° F. so that the damage would not more than offset the advantage.

Of course, the change in the rate of cooling was much more pronounced in the case of the arc welding where preheats and postheats in excess of 500° F. are indicated in

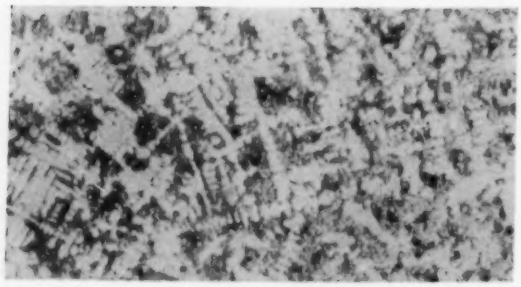


Gas-welded specimen. Arc-welded specimen.

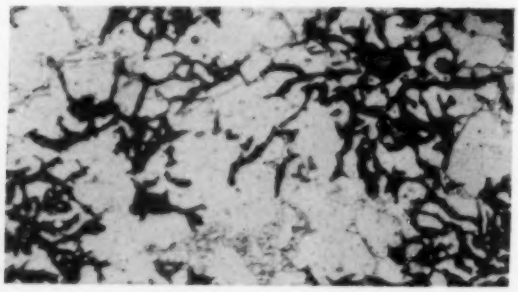
Fig. 8—Note that low hardness is found on both arc- and gas-welded specimens with 1400° F. preheat and postheat, not only in weld and heat-affected zone but also in parent metal.



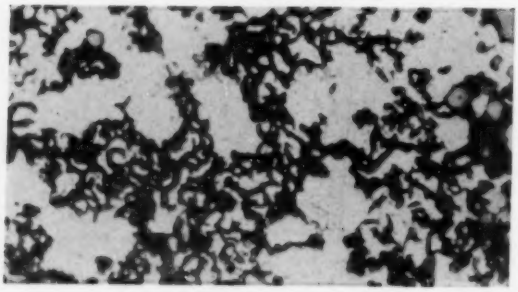
Weld. 100X



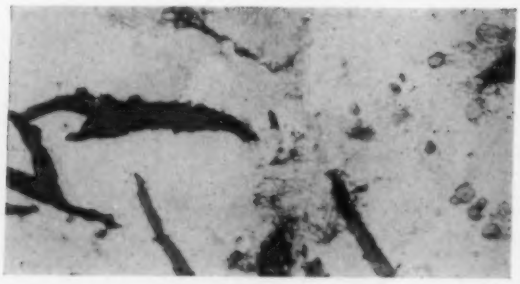
Heat-affected zone. 100X



Weld. 500X

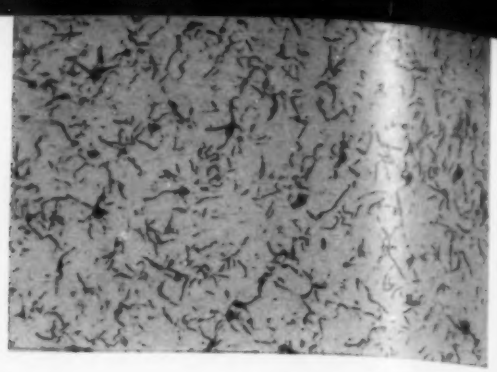


Heat-affected zone (nital etch). 500X

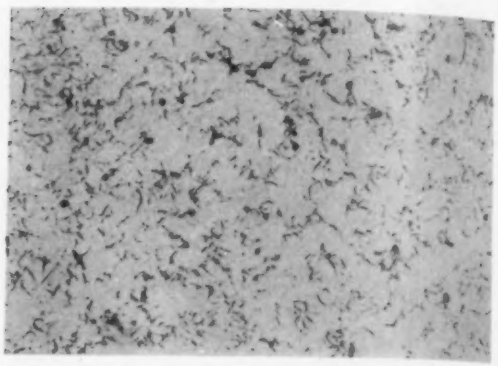


Parent metal (nital etch). 500X

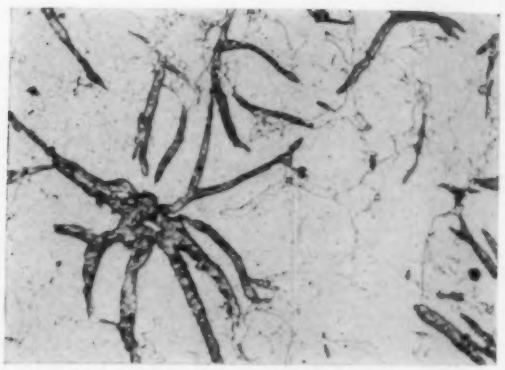
Arc-welded specimens.



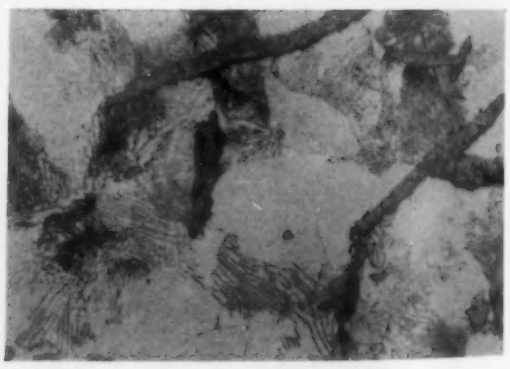
Weld. 100X



Heat-affected zone. 100X



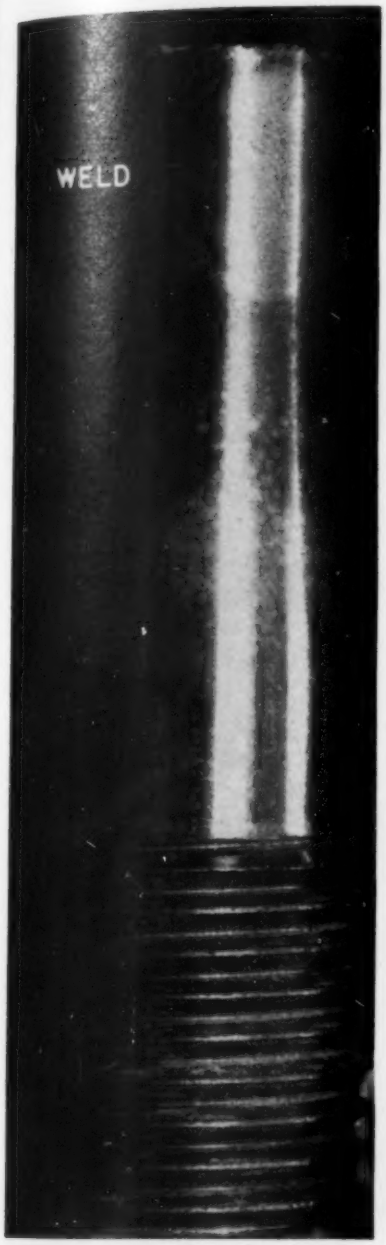
Weld (nital etch). 500X



Heat-affected zone (nital etch). 500X

Gas-welded specimens.

Fig. 9—The structure of these welded specimens with 1400° F. preheat and postheat shows a ferrite matrix in heat-affected zone and parent metal, although steadite and carbides are still present. Low combined carbon.



Specimen (1)
Gas Weld
27,500 psi.
GM 13M Iron

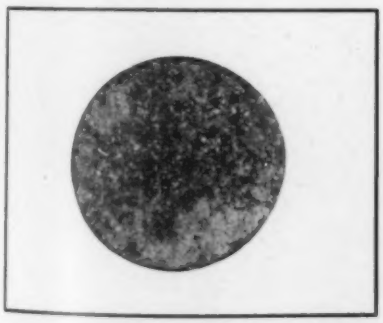
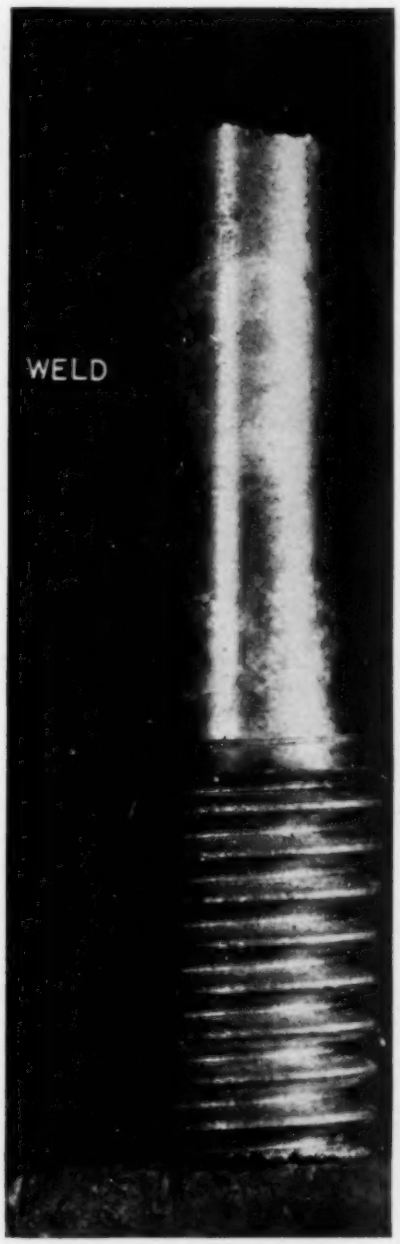


Fig. 10—Gas-weld specimen (type No. 1 rod) highly etched to show weld location.



Specimen (2)
Gas Weld
29,000 psi.
GM 13M Iron



Fig. 11—Gas-weld specimen (type No. 2 rod) showing failure outside weld.



Specimen (4)
Arc Weld
26,000 psi.
GM 13M Iron



Fig. 12—Arc-welded specimen with 1000° F. preheat and postheat showing failure well outside of weld.

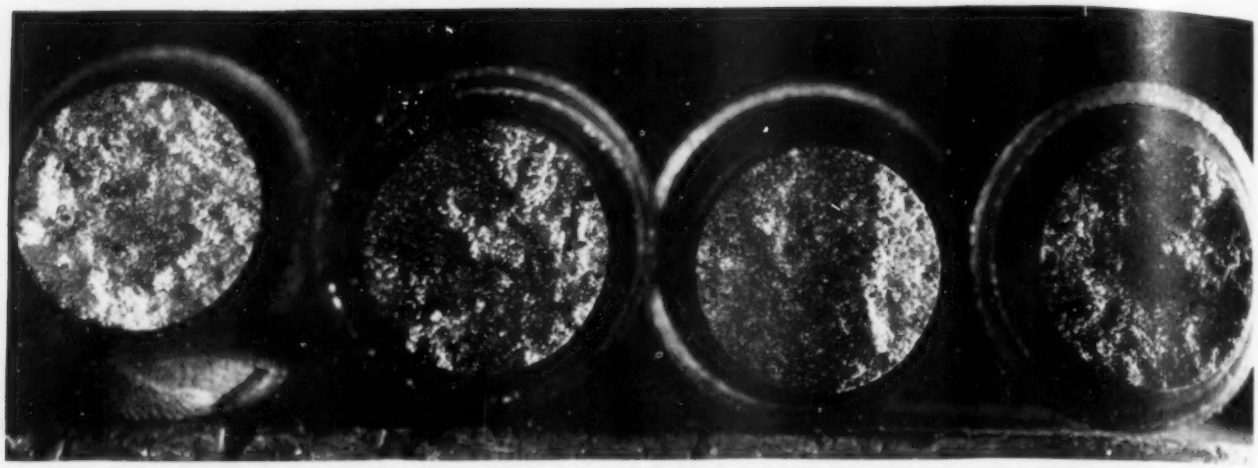


Fig. 13—Comparison of the type of fractures of the four 0.505-in. bars. No. 3 is the parent metal specimen (38,500 psi.).

order to obtain a desirable structure for either machining or strength characteristics. While the arc welding is indicated as a more limited process, its advantage of much greater production speed should not be overlooked.

Physical Properties. The gas-welded 0.505 in. tensile bars made represented the first condition outlined in the foregoing, that is, with only local preheat and postheat. However, the arc-welded specimen was welded with a 1000° F. preheat and postheat, as arc welding at room temperature was felt to be too severe to offer practical information. The tensile strengths of the gas-welded specimens at 27,500 psi. (Fig. 10) and 29,000 psi. (Fig. 11) were somewhat higher than that of the arc-welded specimen at 26,000 psi. (Fig. 12).

However, it will be noted that all specimens broke outside the weld and in the heat-affected zone, indicating possibly that the general preheat and postheat might have weakened the heat-affected structure somewhat. Comparison with the base metal strength of 38,500 psi. (Fig. 13) indicates that the heat-affected zones have been reduced in strength about one-third. Since these welds were made in as nearly a production manner as possible and then x-rayed for soundness, they are believed to be fairly representative of what might be expected under production conditions.

Therefore, in the use of welding, the strength of the heat-affected zone and not the weld should be considered. All of this indicates that a strength factor of two-thirds of casting strength should be used in considering these welding processes.

Nickel Welding. During the study of this type of welding it was found that the different methods of application were equally different in the results obtained. Welding with

Fig. 14—Typical microstructure of cold (bare nickel) welding showing penetration of the nickel into the iron. 100X (left) and 500X (right).

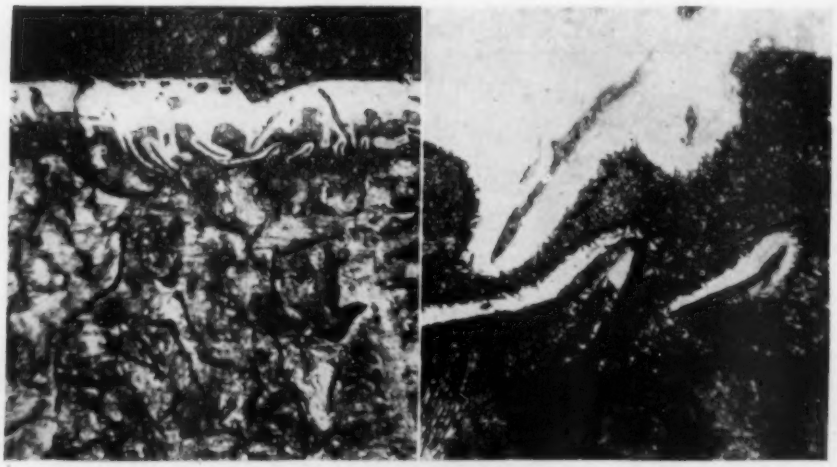
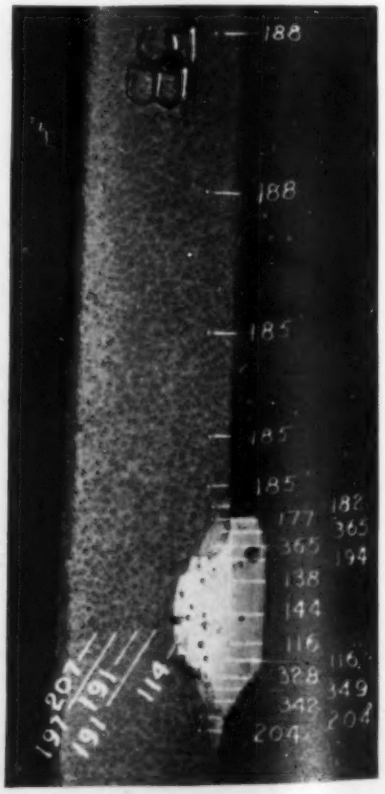


Fig. 15—The weld made with the copper-sheathed nickel without coating shows a narrow heat-affected zone, but also hard. This can be controlled to some extent by varying procedure.



the bare nickel electrode with the low-voltage high-current technique (commonly called bore and cold welding) gave almost no perceptible heat-affected zone.

In fact, while the technique would indicate that only a mechanical bond is obtained, this is not true as there does seem to be a penetration of the nickel right into the cast iron base metal (Fig. 14). Of course, this technique is practical only for small flaws, but when applied with reasonable skill should give highly desirable results.

Welding with the copper-sheathed nickel electrode gives a more apparent heat-affected zone, although narrow, and is done in a more conventional manner. Again, good fusion is in evidence, but localized high hardnesses are observed (Fig. 15).

The coated electrode welding has the penetration and fusion characteristics of standard mild-steel electrodes. This is true of both the nickel-core electrode as well as the monel-core electrode. (Fig. 16). It will be noted that both of these deposits have much the same appearance and the same heat-affected zone characteristics (hardness, range, etc.). The microsection of the heat-affected zone (Fig. 17)

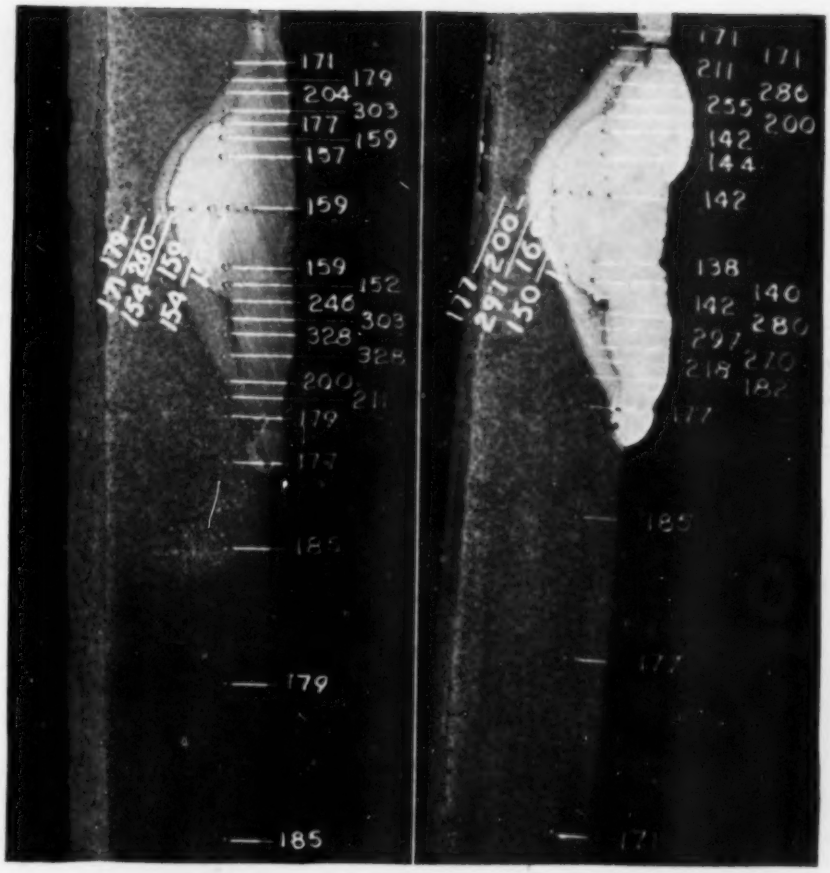


Fig. 16—Comparison of the coated nickel (right) and monel (left) electrode deposits show a close similarity between the extent and hardness of the heat-affected zones, although the weld metals etch differently.

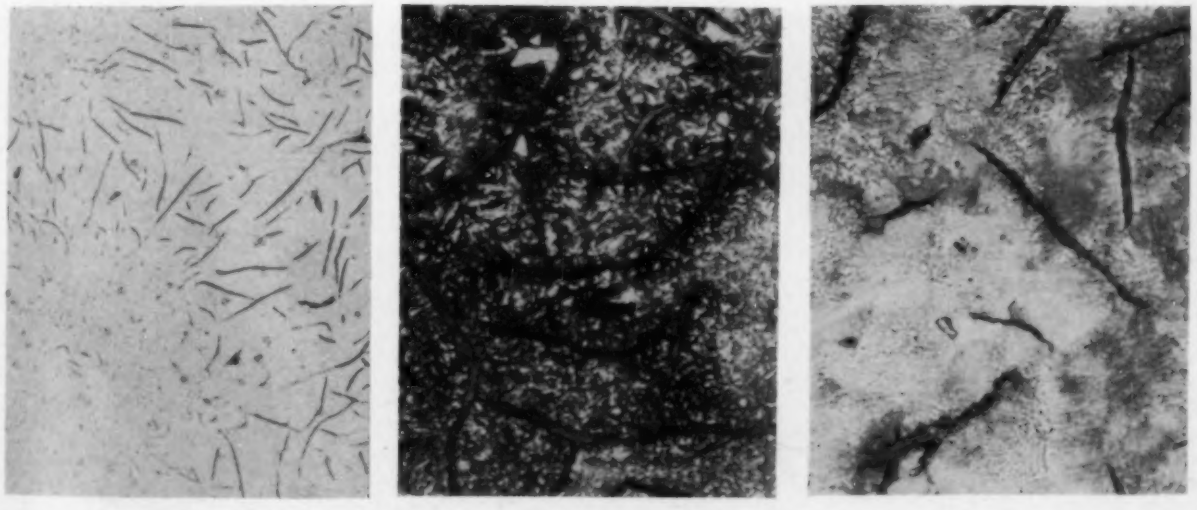


Fig. 17—The transition from martensite to sorbite matrix in the heat-affected zone shows the sudden change in structure that can be expected with monel or nickel coated electrode.

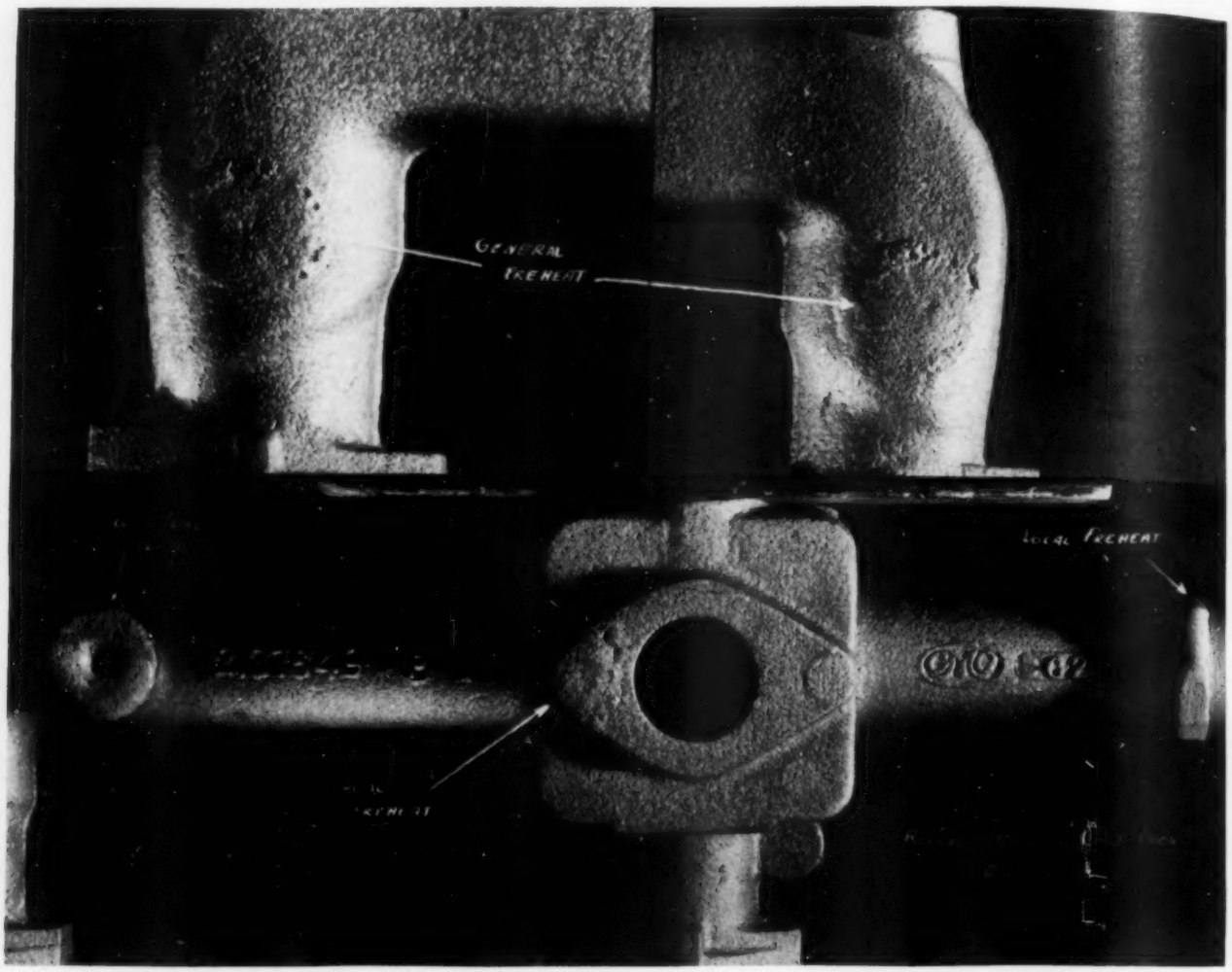


Fig. 18—Sectional views of one of the intake manifolds with arrows and notes pointing out the gas-weld repairs and the type of preheat and postheat.

shows a martensitic structure not unlike that found in the cast iron arc-welded deposit.

However, both the bare and coated monel, as well as the nickel, have lower maximum hardness ranges in the heat-affected zone than the cast iron arc-welded material, even though the parent metal was at the same temperature and condition at the start of welding. This probably is due to the higher amperage required for the cast iron material (170 amp.) as compared to the rather low amperage used with the nickel and monel materials (80 amp.).

Softness of the deposited metal, plus the relatively low heat-affected zone hardness, indicates that this process should be satisfactory for areas to be subsequently machined. Although ordinarily used only for small flaws, the strength of the joint again was found to be a function of the heat-affected zone, and therefore should approximate the tensile properties shown for the cast iron filler metal.

Table 2
COMPOSITION OF GAS-WELDED HIGH-SULPHUR CAST IRON

Element	*Bar, per cent*	*Weld, per cent*
Total Carbon	3.63	
Graphitic Carbon	2.78	
Combined Carbon	0.85	
Manganese	0.54	
Phosphorus	0.11	0.32 *
Sulphur	0.27	0.091*
Silicon	2.60	2.70 *
Nickel	0.19	
Chromium	0.14	
Molybdenum	0.02	
Copper	0.10	
Vanadium	0.05	

*The elements only checked on weld.

High Sulphur Cast Iron. During the war, the subject of the effect of sulphur content in the cast iron was brought up repeatedly. One of the criticisms of high sulphur was that it impaired the weldability of the cast iron. A study was made of some gray iron castings in which excess sulphur was added in the ladle.

Two manifold castings were made from a production pattern, and gas-weld repairs were made under normal conditions (Fig. 18). A 0.505-in. bar was made from a welded piece of runner stock from the same iron (Fig. 19). Chemical composition of this material is shown in Table 2.

Macrosection of one of the repaired flanges is shown in Fig. 20 with hardness range. Although this weld was done with local preheat and postheat, it will be noted that the hardnesses are moderate except in the deposited weld metal itself. No indication of porosity can be seen, or of abnormal structure of greater magnitude than that usually found in this type of repair. Aside

from a pronounced sulphur odor during welding, no difference was noted in welding on this material.

While the 0.505-in. tensile bar showed a somewhat lower ultimate strength than those previously listed, the type of failure was the same and the results are considered within possible variation of the process and parent metal. Even though the sulphur content was considerably above permissible Navy specification limits, both castings and welds were found sound.

Conclusions

Gas Welding. Preheat and postheat treatments serve the primary function of minimizing thermal stresses in the casting and not necessarily simplification of the welding technique. Therefore, welding done on areas without potential locked-up stresses needs only local preheat and postheat with the torch in order to minimize effects of local thermal gradients. Even with the local heat treatment, the maximum hardnesses found were in the weld metal, which remained quite constant throughout test, until the preheat and postheat temperatures of 1400° F. were used.

This process gives physical properties with tensile strength 70 per cent of that of the unaffected parent metal. The heat-affected zone is the critical area, and does not seem to be materially improved by the preheat and postheat treatment. Therefore, weld failures on repaired castings indicate stresses in the welded areas in excess of this 70 per cent factor. Sulphur content up to 0.27 per cent gave no noticeable effect on the casting or its weldability.

Arc Welding. In order to minimize the much sharper thermal gradients found with this process, 1000° F. preheat and postheat treatment is necessary if the hardness range of the weld and heat-affected zone is to be kept within reasonable limits. However, the soundness of the weld and its physical properties appear equally as good as those obtained by the gas-welding process. As the speed of welding is much faster with the arc process (and the electrode has good operation characteristics), this process should come into increasing favor.

Cast Iron Filler Metal Processes. Preheat and postheat treatment in excess of 1000° F. is potentially harmful to the casting even with

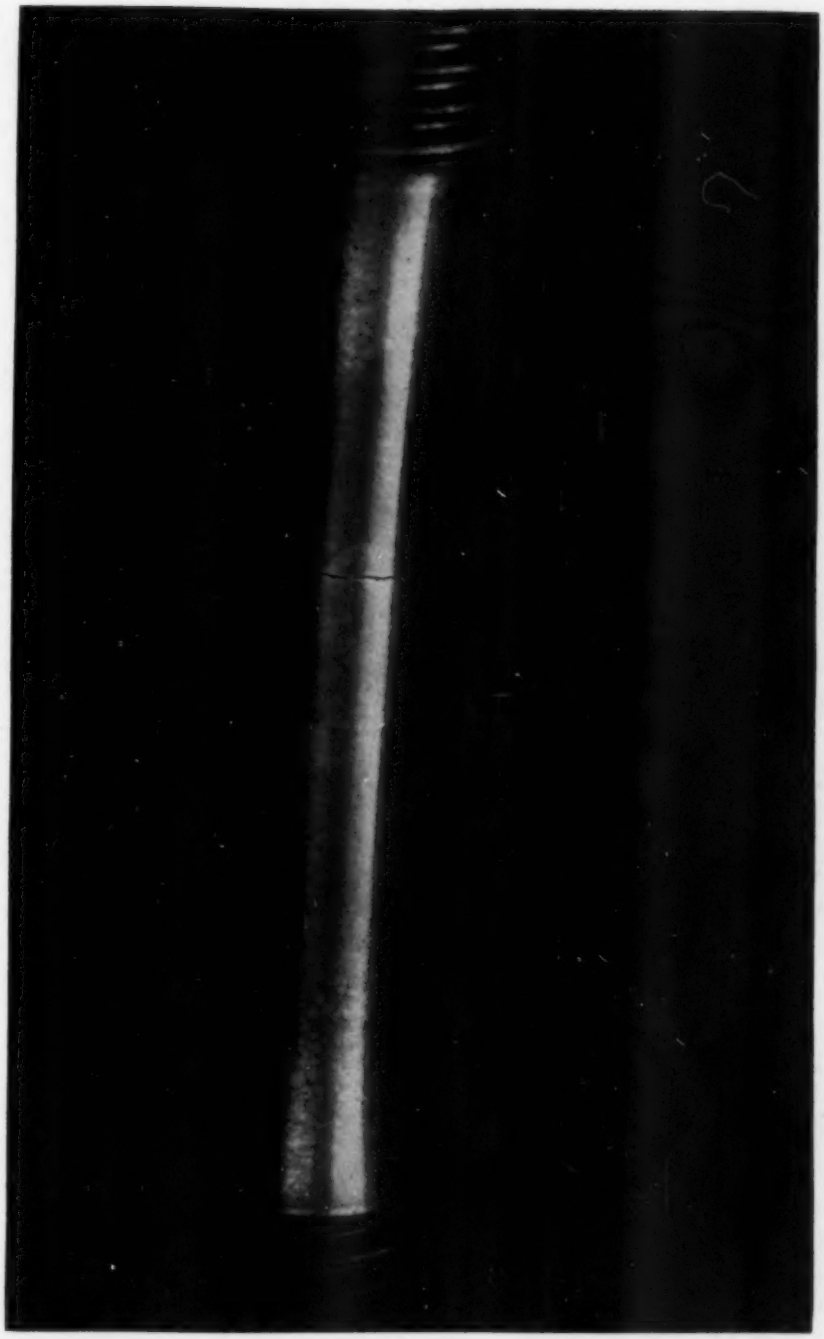
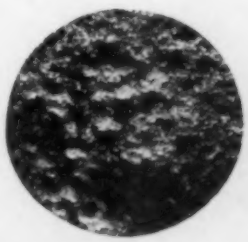


Fig. 19—High sulphur gray cast iron 0.505 in. diameter tensile bar. Tensile strength, 20,000 psi.



Failure in heat-affected zone and weld.



No perceptible flaws in failure.

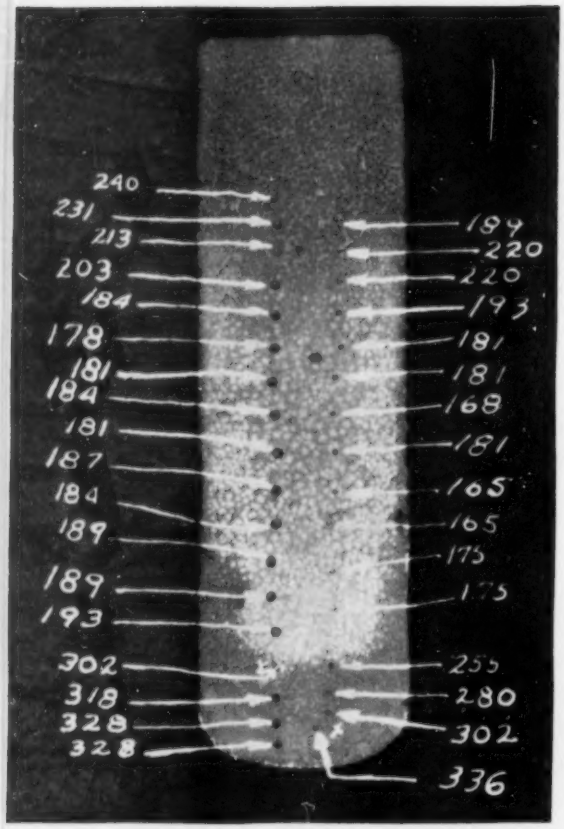


Fig. 20—Although only locally preheated and postheated (see Fig. 18), the hardnesses and structure of this gas weld on high-sulphur iron seem normal.

short heat treatment cycles. Even at 1000° F. some evidence of ferrite formation is apparent around the graphite flakes, which at 1400° F. has grown into a ferrite matrix. The nature of the structure at 1400° F. indicates a low amount of combined carbon, and the resultant structure is weak and not nearly as desirable for the end use of most gray iron castings, even though the heat-affected zone was almost entirely eliminated at this heat.

Nickel Welding. Both nickel and monel materials indicate good results for small flaws, with varying amounts of heat-affected zone with the different techniques. Cold welding showed no appreciable heat-affected zone, while the coated electrodes gave zones similar to the cast iron filler metal in the arc process.

Using Coated Electrodes

However, in the case of the coated electrodes, the heat-affected zone was found to be much lower in hardness than with the cast iron process, and this is attributed to the lower electrical energy input during welding. This factor indicates that the practice of making short weld segments and cooling before continuing welding is sound, both from the heat input standpoint and that of minimizing thermal stresses.

The bare nickel shows a peculiar affinity for cast iron even though arc welding in the true sense is not accomplished. All of the deposited metals in this class are soft and suitable for machining. Intermittent welding results checked seem to be equally good, but have a greater tendency toward porosity.

1. Both gas and arc cast iron filler metals produce structures or assemblies having two-thirds the ultimate strength of the parent material.
2. These physical characteristics can be expanded in the future to make practical fabrication of cast iron weldments.
3. Cast iron arc welding is equally as good as gas welding under proper conditions, and is advantageous because of its higher production speed.
4. Excessive preheat and postheat treatments often are more harmful than no treatment at all.
5. The peculiar affinity of nickel for iron makes possible the elimination of the heat-affected zone in repairing small flaws. Both monel and nickel offer satisfactory material for repairing without the use of preheat and postheat.
6. The sulphur content of the iron when less than 0.27 per cent is not considered a factor in weldability.
7. All of the foregoing methods investigated have a useful place, and if studied and properly controlled will do much to minimize foundry scrap and expedite production.

Acknowledgment

The authors wish to gratefully acknowledge the help of R. C. Nelson, Walter Zeller, and E. O. Falberg of GMC Truck & Coach Div., and of S. M. Lenhoff and J. P. Thomas of Detroit Diesel Div., of General Motors, in the preparation of this paper.

LOW TEMPERATURE PROPERTIES of CAST STEELS

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Columbus, Ohio

A SERIES OF COMMERCIAL and experimental steels was studied to determine what correlations exist between the notched-bar properties and other factors such as composition, deoxidation practice, heat treatment, and microstructure in an attempt to learn more about the component factors of notched-bar properties of cast steels.

Melting and Foundry Practice. The steels tested in this investigation comprised two groups whose compositions are given in Tables 1 and 2. The 11 manganese-molybdenum steels were made commercially in open hearth or electric furnaces and received different deoxidation additions but were otherwise similar in composition. The test samples were cast in double-keel test blocks having two coupons 1⅛ in. square and 14 in. long. The coupons were burned from the raw castings with an oxyacetylene torch.

The other 18 steels were made in a laboratory induction furnace and cast in plate molds using a blind riser and atmospheric pressure for feeding. Figure 1 shows the test casting used for these steels.

Sixteen of the laboratory heats were melted in a magnesia crucible from a charge of ingot iron and ferrosilicon sufficient to give 0.10 per cent at melt down. After a temperature of 3000° F. was reached, the ferro-alloy additions were made, and the heat was held 2 min.

The final deoxidizer was then added, and the metal was poured into a small ladle. The time between deoxidation and casting was about one min., and the pouring temperature averaged 2965° F. as indicated by an optical pyrometer. The steel was cast in oil-sand molds which had been skin dried with a gas flame.

Two steels (Nos. 1 and 12, Table 2) exhibited central gas porosity when made by the procedure described in the foregoing. The defects were round, shiny holes and bore no similarity to shrinkage cavities except for their location. The typical appearance and the unsoundness in the silicon-killed steels with 0.63 per cent manganese is shown by the photograph in Fig. 2 at ¾X.

This type of porosity was found in castings made in skin-dried oil-sand molds and in molds dried for 6 hr. at 450° F. Five of the six test castings made from these compositions in oil-sand molds exhibited the center-line defects. The good casting was one of three made with a carbon boil during melting and oil-sand molds baked at 450° F. Six castings made later in fire clay-bonded molds dried at 450° F. for 4 hr. were sound. The composition of the mold mixtures is indicated in Table 3.

Porosity obtained in these castings was first attributed to hydrogen in the charge, but this theory was discarded when the carbon boil effected no improvement. The fact that the change to fire clay-bonded molds eliminated the trouble indicated that the core oil was the source of the trouble.

It seems likely that the core-oil residue, remaining after drying, reacted with the steel to form carbon monoxide, carbon dioxide, and hydrogen. Presumably some of the gas dissolved in the steel and later formed bubbles near the center where it was concentrated by dendritic segregation.

Similar steels deoxidized with aluminum (Nos. 2 and 13) and casting No. 3 with 1.55 per cent manganese were sound when cast under conditions giving porosity in the steels discussed in the foregoing. Apparently the presence of strong deoxidizers or the extra manganese was sufficient to prevent gassing in these castings.

Heat Treatment. The manganese-molybdenum steels were heat treated

Twenty-eight cast steels were tested to determine their notched-bar behavior at temperatures between 80 and −100° F. The data indicated that the notched-bar properties characteristic of a heat of steel were not reflected consistently by any other property measured. Therefore, it appears necessary to test each lot of steel when low-temperature properties are considered important, because little reliance can be placed on generalities correlating them with other characteristics of materials.

Presented at a Steel Session of the Fiftieth Annual Meeting, American Foundrymen's Association, at Cleveland, May 10, 1946.

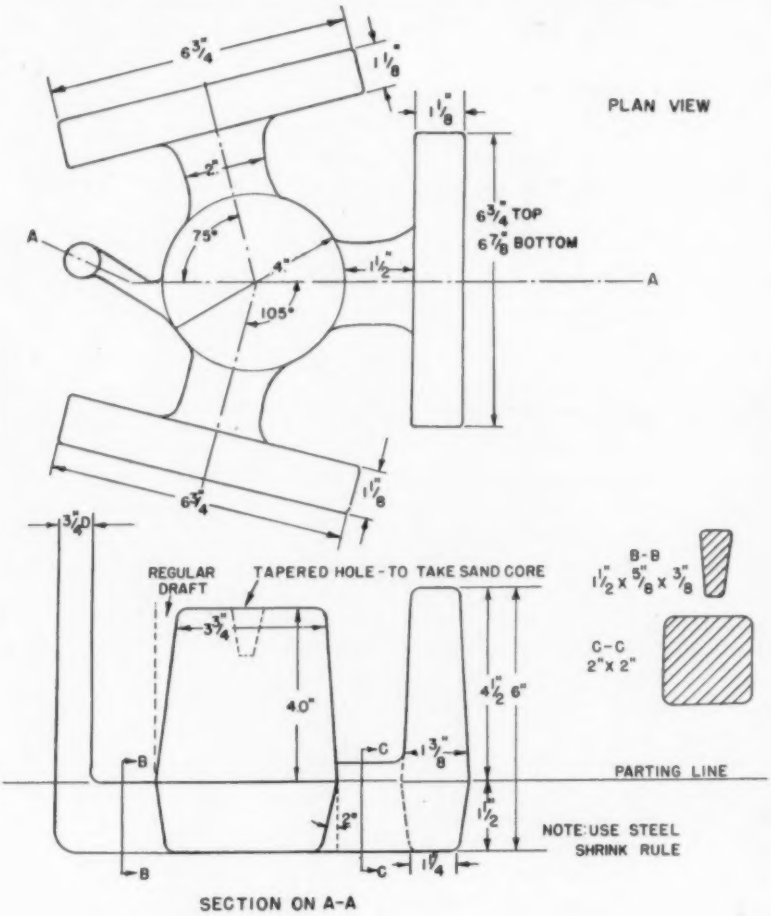


Fig. 1—Sketch of pattern for cast steel test plates.

in sections 1⅛ in. square and 14 in. long. The steels made in the laboratory were heat treated as plates 1⅛x6x6½ in. All raw castings were heated 2 hr. at 1800° F. and air cooled before receiving the final heat treatments. The manganese-molybdenum steels were water quenched from 1600° F., tempered 5 hr. at 1240° F., and air cooled.

Previous experiments had shown that this grade of steel was not temper brittle. The laboratory steels were tested in both normalized and drawn and water-quenched and tempered conditions. The heat-treating temperatures varied somewhat because of the differences in composition.

Grain Size and Hardenability. The austenite grain size developed by the actual heat-treating schedule was determined for all steels by means of an interrupted quench. The austenite grain-growth characteristics of the steels were evaluated in a qualitative way by carburizing samples for 8 hr. at 1750° F.

This temperature was about 150° F. higher than the actual heat-treating temperature and caused grain growth in some steels which were fine grained at the lower temperature. These tests showed that the experimental steels varied in their grain size and grain-growth characteristics with the deoxidation practice.

Standard end-quench hardenability tests were run on the laboratory steels using the same temperature and schedule employed in the actual heat treatment. The end-quench data indicated that steels Nos. 3 and 11 had the highest hardenability and were fully hardenable under the conditions employed. These two materials were silicon-killed medium-manganese steels.

Since no stock was available for end-quench tests on the manganese-molybdenum steels, their approximate hardenabilities were calculated according to Grossmann's method.[1] The hardenability data obtained by experiment and calculation are given in Table 4.

Metallography. Metallographic examinations were made on unetched samples of the various steels to determine the effect of differences in deoxidation practice on the inclusions. The distributions of the sulphides, which were the most numerous and probably the most important inclusions, were classified according to the system proposed by Sims and Dahle[2]. These classifications are illustrated by the photomicrographs in Fig. 3.

Globular Type I sulphide distribution was characteristic of all silicon-killed steels including those receiving calcium or vanadium additions. The sulphide inclusions in the commercial steels with less than 0.01 per cent residual titanium were also globular. This inclusion distribution

Table 1

COMPOSITION OF COMMERCIAL STEELS TESTED

Heat No.*	Composition, per cent C	Mn	Si	S	P	Mo	Residual Deoxidizer	Deoxidation Deoxidant	Treatment—lb. per ton
158	0.29	1.00	0.30	0.034	0.042	0.42	0.008 Ti	Ti	2.0
9982	0.28	1.01	0.27	0.040	0.035	0.40	0.008 Ti	Ti	2.0
131	0.29	0.99	0.31	0.037	0.032	0.38	0.008 Ti	Ti	2.0
608	0.29	0.95	0.31	0.033	0.036	0.38	0.033 Al	Al	2.8
1141	0.31	1.05	0.32	0.034	0.034	0.33	0.050 Al	CaMnSi Al	2.5 2.8
536	0.29	0.96	0.33	0.033	0.036	0.16	0.089 Al	Al, Boron added	3.0
252	0.27	1.00	0.47	0.031	0.030	0.41	0.002 Al	Al	2.5
1261	0.30	1.01	0.30	0.039	0.035	0.34	0.009 Ti	Ti	2.0
1008	0.28	1.00	0.35	0.035	0.025	0.40	0.007 Ti	Ti	2.0
1092	0.32	0.96	0.30	0.038	0.029	0.34	—	CaMnSi	2.5
1739	0.29	1.00	0.34	0.032	0.030	0.32	—	CaMnSi	2.0

*Steels with heat numbers below 700 were made in an electric furnace; the others were made in an acid open hearth furnace.

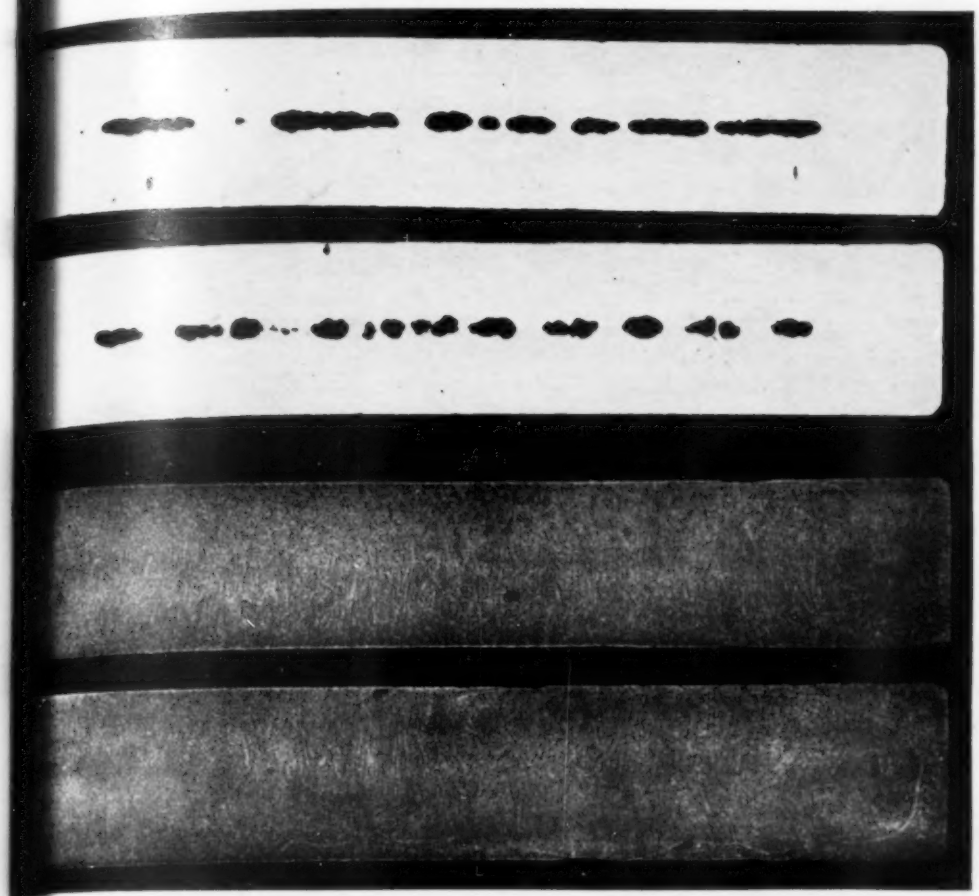


Fig. 2—The two sections at the top were taken from plates cast in an oil-sand mold; the two at the bottom from plates cast in a fire clay-bonded mold. Both steels were made by the same melting practice. ¾X.

is generally considered to have the least harmful effect on ductility.

Manganese sulphide inclusions were comparatively massive and angular in the steels containing at least 0.03 per cent aluminum or 0.05 per cent zirconium. This distribution was classified as Type III, and steels with inclusions of this type generally had good ductility.

Eutectic or chain Type II sulphides were found in steels containing small amounts of aluminum, zirconium, or titanium, and when present, caused embrittlement in both tensile and notched-bar tests. These three elements act similarly in that sufficiently small additions which are largely oxidized will give Type I sulphides, while slightly larger additions form the Type II. Zirconium, like aluminum, forms the more massive Type III sulphides when added in excess, as in steel No. 9.

The five commercial steels, which contained less than 0.01 per cent titanium, had globular Type I sulphides similar to silicon-killed steels. The three laboratory steels (6, 7, 18), on the other hand, containing 0.02, 0.07, and 0.16 per cent titanium, respectively, all had Type II chain or film sulphides. It was concluded, therefore, that the change from film sulphides to massive Type III sulphides is not produced by excess of titanium.

It appeared that the inclusions

Table 2

COMPOSITION OF EXPERIMENTAL CAST STEELS TESTED

Casting No.	Composition, per cent					Deoxidation Treatments	
	C	Mn	Si	S	Special Element	Deoxidant	lb. per ton
1	0.27	0.62	0.33	0.026	—	Si only	
2	0.31	0.66	0.38	0.033	0.053 Al	Al	1.5
3	0.31	1.55	0.42	0.036	—	Si only	
4	0.31	1.56	0.42	0.039	0.011 Al	Al	0.4
5	0.29	1.49	0.42	0.037	0.038 Al	Al	1.5
19	0.32	1.30	0.39	0.030	0.120 Al	Al	4.0
6	0.30	1.41	0.35	0.033	0.020 Ti	Ti	1.0
7	0.31	1.51	0.44	0.032	0.070 Ti	Ti	2.0
18	0.30	1.59	0.60	0.018	0.16 Ti	Ti	5.0
8	0.31	1.42	0.33	0.036	0.053 Zr	Zr	1.5
9	0.27	1.37	0.42	0.036	0.080 Zr	Zr	3.0
10	0.28	1.29	0.39	0.040	0.110 V	V	2.0
11	0.32	1.45	0.40	0.036	—	Ca	2.0
12	0.28	0.65	0.38	0.030	1.54 Ni	Si only	
13	0.30	0.68	0.36	0.041	1.55 Ni, 0.04 Al	Al	1.5
15	0.32	0.65	0.33	0.044	1.55 Cu, 0.02 Al	Al	1.5
16	0.16	1.00	0.97	0.033	1.67 Cu	Si only	
17	0.19	0.89	0.81	0.035	1.67 Cu, 0.03 Al	Al	1.5

Table 3

PROPORTIONS OF INGREDIENTS USED IN MOLD MIXTURES FOR EXPERIMENTAL CASTINGS

Components	Mixture	
	Oil Sand	Dry Sand
Glass rock sand, lb.	80.0	88.0
Western bentonite, lb.	0.6	—
Silica flour, lb.	7.0	—
Corn flour, lb.	1.3	—
Water, qt.	2	3
Core oil, qt.	1	—
Fire clay, lb.	—	11.0



Globular sulphides, Type I.



Angular sulphides, Type III.



Eutectic or film sulphides, Type II.



Eutectic or film sulphides, Type II.

Fig. 3—Characteristic occurrence of sulphide inclusions in cast steels, classified in three types. Unetched. 1000X.

became increasingly finer and more continuous as the titanium content increased. The photomicrograph in Fig. 4 shows the sulphide distribution in the steel with 0.16 per cent titanium. These inclusions were extremely fine and were light tan or pink in color instead of the usual dove-gray sulphides high in manganese.

The belief that the change in shape and color indicated that they were titanium sulphides was strengthened when evolution analyses gave sulphur values about half of those for other steels made with the same melting practice. As would be expected from the inclusion distribution, this steel was brittle in mechanical tests. The low-carbon ferrotitanium used in making the laboratory steels contained about 8.6 per cent aluminum, so this steel (No. 18) probably had an appreciable residual aluminum content.

In addition to sulphides, the steels contained other inclusions distinctive of their finishing treatments. All of the silicon-killed steels, including those treated with calcium compounds, contained impure silicates. All of the calcium-treated steels had "eyes" or dark spots in the globular sulphides, which seems to be a characteristic effect of this element. The appearance of such sulphides is illustrated by the photomicrograph in Fig. 4.

Inclusion Types

Aluminum-killed steels contained some small alumina inclusions. The heats deoxidized with zirconium contained random clusters of zirconium silicates, as shown in Fig. 4. The zirconium steels were the dirtiest of the steels studied.

Induction furnace steels contained more inclusions than the commercial materials, probably because of the higher pouring temperatures. Steels 1261 and 1739 had more inclusions than the other manganese-molybdenum commercial steels. One of these open hearth steels was deoxidized with titanium and the other with a calcium-manganese-silicon alloy.

All of the steels were examined carefully in the etched condition at various magnifications. The microstructures of representative steels are shown in Figs. 5 and 6. Little difference existed in the amount of free ferrite or the degree of spheroidization of similar steels receiving the same quench and draw treatment.

Some of the steels had a more pronounced pattern or network caused by segregation of manganese, but this had no discernible effect on the properties. The photomicrographs for the normalized and drawn condition show that the two best medium-manganese steels had a finer actual grain size than the others. Both steels contained appreciable amounts of residual aluminum.

Titanium steels were equally fine grained but had inferior ductility and toughness because of the Type II inclusion distribution. No consistent differences were found between the microstructures of the steels which were embrittled at low temperatures and those which were relatively unaffected.

Tensile Tests. Tensile tests were made on the induction furnace steels in the normalized and drawn, and the quenched and tempered conditions. The same heat treatments were also used for the notched-bar samples. Identical heat treatments were used on steels of the same grade in order to conserve stock and time.

It is realized that the heat treatments were not ideal for all steels, and the comparisons may be unfair to certain steels which received unfavorable heat treatments. The data given in Table 5, obtained on specimens taken from the top of the 1⅛-in. plates, show that the steels had normal properties for their grade and heat treatment.

Grade "B" steels and those with 1.55 per cent nickel or copper had lower strengths than the medium-manganese steels in the conditions tested. The copper-manganese-silicon steels were tested at somewhat higher strengths than the other steels, but gave good elongation and reduction in area values.

Ordinarily the steels gave better

Table 4

HARDENABILITY DATA FOR EXPERIMENTAL CAST STEELS

Sample No.	Carbon, per cent	ASTM Grain Size*	Special Element or Deoxidizer	Ideal Diameter, in.	End-Quench Test Data — Rockwell "C" Hardness at Indicated Distances, in. — Surface	1/32	1/16	1/8	3/16	1/4
GRADE "B" STEELS										
1	0.27	6	Si	0.80	45	43	37	15	11	—
2	0.31	7	+Al	0.83	45	43	24	12	9	—
MEDIUM-MANGANESE STEELS										
3	0.31	5	Si	1.74	49	48	47	44	38	22
4	0.31	7	Al	1.54	49	48	46	38	29	18
5	0.29	9	Al	1.36	48	47	45	29	20	13
19	0.32	9	Al	1.63	49	48	48	41	33	21
6	0.30	9	Ti	1.13	49	48	47	34	24	15
7	0.31	8	Ti	1.32	48	46	44	35	27	18
18	0.30	9	Ti	1.64	49	48	44	32	23	18
8	0.31	6	Zr	1.45	48	47	46	40	29	18
9	0.27	6	Zr	1.35	48	48	47	42	29	17
10	0.32	7	V	1.53	48	48	46	34	25	20
11	0.33	5	(Ca)	1.65	49	49	48	43	37	24
1.55 PER CENT NICKEL OR COPPER STEELS										
12	0.28	6	Ni+Si	1.40	48	46	44	32	20	12
13	0.30	8	Ni+Al	1.47	49	48	45	22	17	10
15	0.32	8	Cu+Al	1.35	49	48	38	17	17	14
COPPER-MANGANESE-SILICON STEELS										
16	0.16	7	Si	1.61	41	40	38	22	22	15
17	0.19	8	Al	1.65	41	37	28	14	14	13
MANGANESE-MOLYBDENUM STEELS										
158	0.29	5	Ti	2.82						
9982	0.28	5	Ti	2.60						
131	0.29	5	Ti	2.58						
608	0.29	8	Al	2.23						
1141	0.31	9	Al	2.22						
536	0.29	8	Al+B	2.58						
252	0.27	8	Al	2.28						
1261	0.30	5	Ti	2.52						
1008	0.28	5	Ti	2.74						
1739	0.29	6	Si	2.28						
1092	0.32	6	Si	2.25						

Hardenabilities calculated according to Grossmann's method[3]

*ASTM grain size was determined for actual heat-treating schedule.





Fig. 4—Photomicrographs showing unusual inclusions found in three experimental steels. Unetched.

Top—Fine eutectic sulphides in high-titanium (0.16 per cent Ti) steel. 400X.

Center—"Bird's eye" sulphides in calcium-treated steel. 500X.

Bottom—Clusters of zirconium oxides and silicates (0.05 per cent Zr). 1000X.



ductility values in the quenched and tempered condition than they did after the other heat treatment. There were no consistent differences in the tensile properties of the steels deoxidized with aluminum and those which had been silicon killed.

Titanium-bearing steels (Nos. 6, 7, 18) and the steel with 0.05 per cent zirconium had the poorest ductility of the experimental steels. The brittleness of the titanium steels, which increased with increasing amounts of this element, is attributed to the presence of intergranular sulphides. Zirconium steel (No. 8) had an inordinate oxide inclusion content which probably had a deleterious effect. The aluminum-killed steel which contained film sulphides (No. 4) had the poorest ductility of the other steels.

Data for the quenched and drawn samples show that the titanium and vanadium steels (Nos. 6, 7, 18, 10) had considerably higher yield strengths and slightly higher ultimate strengths than the other medium-manganese steels. The fact that a high yield strength-ultimate strength ratio was obtained with only 0.02 per cent titanium is somewhat surprising. The high-yield ratio is attributed to a fine dispersion of titanium or vanadium carbides, because aluminum-killed steels of similar grain size had normal yield-strength ratios.

The fact that the higher ratios were not obtained in the normalized and drawn specimens shows that the effect varies with heat treatment. Experience on wrought titanium steels[3] confirms this belief, and it appears that the effect is more pronounced with manganese contents of over one per cent.

Notched-Bar Tests. Charpy notched-bar specimens were broken at temperatures between 75° F. and −100° F. to determine the susceptibility of the different steels to low-temperature embrittlement. The samples were cooled 3° F. below the testing temperature in a bath of dry ice and acetone and held 20 min. before testing.

Previous work indicated that this amount of supercooling compensated for the warming-up which occurred when the samples were transferred to the testing machine. Three or more bars were broken at each test temperature, and the reproducibility of the values was generally good. Test values are given in Tables 6, 7, and 8 and, as would be expected, the tougher samples gave a little more scatter.

Manganese-Molybdenum Steels

Table 9 gives a comparison of the notched-bar values and other characteristics of the manganese-molybdenum steels. The materials are listed in increasing order of embrittlement at low temperatures. The minimum values at various temperatures were used, because they are the safest for comparison.

The data show that the six steels which were made without aluminum were similar in most respects, except for their low-temperature impact behavior. Three of them were the best, and three of them were the worst of the series when tested at −80° F. The steels were alike in being coarse grained and in the Type I distribution of the sulphide inclusions.

All of the steels gave Charpy values in excess of 30 ft.-lb. in the room temperature tests. The data show that these steels varied somewhat in hardenability, hardness, and grain-coarsening tendencies, but that none of the variations were consistently related to their behavior in notched-bar tests.

The four aluminum-killed steels gave lower notched-bar values at room temperature than the other

Table 5

MECHANICAL PROPERTIES OF HEAT TREATED EXPERIMENTAL CAST STEELS

Casting No.	Heat* Treatment	Deoxidizer, per cent	Hardness, Rockwell "B"	Average Vee-Notch Charpy Value, ft. lb.	Elongation in 2 in., per cent	Reduction in Area, per cent	Yield Strength, (0.2% offset) psi.	Ultimate Tensile Strength, psi.
GRADE "B" STEELS								
1	a	Si only	71	30.0	33.0	47.8	40,250	70,500
2	a	0.05 Al	78	34.0	25.0	44.1	47,500	75,750
MEDIUM-MANGANESE STEELS								
3	b	Si only	88	21.0	30.0	55.1	65,750	92,500
4	b	0.01 Al	88	25.0	26.5	46.0	65,500	93,250
5	b	0.04 Al	88	50.0	31.0	57.3	65,500	91,750
19	b	0.15 Al	84	44.5	28.5	52.8	59,000	87,750
6	b	0.02 Ti	89	24.5	26.5	44.9	62,500	94,750
7	b	0.07 Ti	92	24.0	25.0	40.2	66,000	96,000
18	b	0.16 Ti	84	12.5	18.8	22.4	64,750	87,500
8	b	0.05 Zr	88	29.5	25.5	40.4	60,750	93,000
9	b	0.08 Zr	80	23.0	25.0	51.2	51,000	80,000
10	b	0.11 V	84	32.0	31.0	56.8	55,000	84,500
11	b	Ca treated	84	22.0	32.0	56.5	49,000	87,000
1.55 PER CENT NICKEL OR COPPER STEELS								
12	b	Si only	78	31.0	31.0	49.3	48,500	78,000
13	b	0.04 Al	80	34.5	32.0	56.2	50,000	77,000
15	b	0.02 Al	86	27.5	28.0	44.9	62,000	82,250
COPPER-MANGANESE-SILICON STEELS								
16	c	Si only	92	15.5	27.0	60.3	70,750	93,250
17	c	0.03 Al	93	29.0	25.0	50.6	78,250	96,500
GRADE "B" STEELS								
1	d	Si only	77	24.0	25.5	46.6	44,250	71,750
2	d	0.05 Al	82	41.0	31.5	54.4	57,750	79,000
MEDIUM-MANGANESE STEELS								
3	e	Si only	89	64.0	29.0	64.0	67,500	90,500
4	e	0.01 Al	89	27.0	23.0	46.6	66,750	88,750
5	e	0.04 Al	90	55.0	29.5	60.6	68,750	88,750
19	e	0.15 Al	90	42.0	25.0	55.0	65,750	87,250
6	e	0.02 Ti	91	26.0	23.5	45.8	76,750	92,250
7	e	0.07 Ti	91	22.5	21.5	37.3	77,500	93,750
18	e	0.16 Ti	90	9.5	20.5	28.5	78,750	92,000
8	e	0.05 Zr	88	47.5	22.5	37.3	67,500	90,750
9	e	0.08 Zr	89	42.0	27.0	53.8	62,750	87,500
10	e	0.11 V	92	42.0	25.0	58.3	74,500	94,250
11	e	Ca treated	91	54.0	24.0	60.1	69,000	92,500
1.55 PER CENT NICKEL OR COPPER STEELS								
12	e	Si only	82	31.0	25.5	43.8	51,250	78,000
13	e	0.04 Al	85	36.0	31.5	60.3	56,750	82,500
15	e	0.02 Al	88	34.0	30.0	56.0	63,250	85,000
COPPER-MANGANESE-SILICON STEELS								
16	f	Si only	98	23.0	23.0	56.0	91,000	118,250
17	f	0.03 Al	96	38.0	23.0	50.6	85,750	103,500

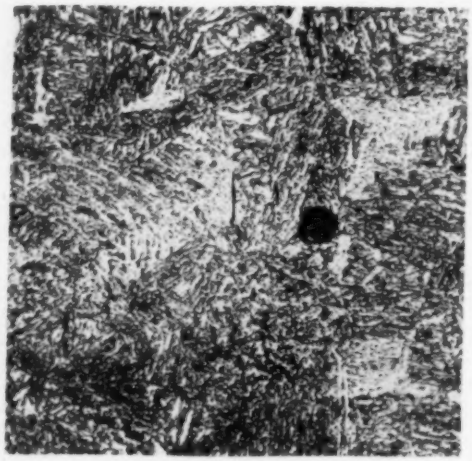
*Heat Treatment. All raw castings were heated 2 hr. at 1800° F. and air cooled before receiving treatments indicated as follows:
a. 1 hr. at 1625° F., air cooled, tempered 2 hr. 1200° F., air cooled.
b. 1 hr. at 1575° F., air cooled, tempered 2 hr. 1200° F., air cooled.
c. 1 hr. at 1600° F., air cooled, tempered 2 hr. 1000° F., air cooled.
d. 1 hr. at 1625° F., water quench, tempered 2 hr. 1250° F., air cooled.
e. 1 hr. at 1575° F., water quench, tempered 2 hr. 1250° F., air cooled.
f. 1 hr. at 1600° F., water quench, tempered 2 hr. 1000° F., air cooled.

Table 6

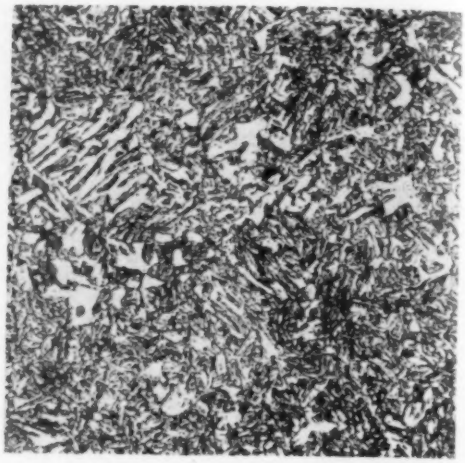
KEYHOLE NOTCH CHARPY IMPACT VALUES OBTAINED ON TEN MANGANESE-MOLYBDENUM STEELS IN THE QUENCHED AND TEMPERED CONDITION, AT TEMPERATURES BETWEEN 75° AND −80° F.

Testing Temperature, °F.	Steel Heat No. 158	9982	131	608	1141	536	252	1261	1008	1739
	Charpy Impact Value, ft.-lb.									
75	38, 38	41, 39	33, 32	29, 28	28, 28, 26	26, 25, 24, 23	16, 15	36, 33	41, 41, 39, 37	42, 40
32	40, 38	40, 38	34, 32	26, 25	29, 27, 25	27, 26	16, 15	36, 33	40, 40	41, 40
0	37, 35	35, 33	32, 32	—	29, 28, 27	24, 24	15, 15	—	37, 32	—
−20	30, 30	34, 34	—	25, 24	25, 25, 22	26, 26, 26	—	34, 20	31, 25	34, 33
−30	33, 32	33, 28	34, 31	23, 23	—	26, 24, 22, 19	15, 13	24, 24	34, 33	38, 33, 29
−40	34, 33	29, 27	—	—	24, 22, 21	24, 24, 20	—	30, 29	31, 31	37, 7, 5
−50	34, 28	33, 30	30, 29	21, 20	—	23, 22	14, 14	—	31, 28, 25, 6	—
−60	30, 27	26, 23	—	21, 19	25, 23, 21	22, 20, 17	—	30, 22	27, 26	32, 31, 28, 18
−70	28, 28	24, 24	27, 24	19, 16	—	20, 13, 12	—	25, 22	30, 25, 5, 3	34, 30
−80	27, 26	27, 24	26, 24	19, 18	20, 19, 19	19, 19, 18	14, 12	23, 5, 4	30, 22, 16, 11	16, 6, 6

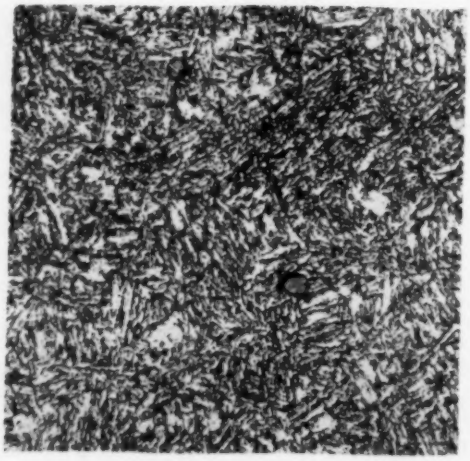
NOTE: Heat treatments were carried out on keel-block coupon 1¼x1¼x14 in.
Heated 2 hr. at 1800° F., air cooled.
Heated 2 hr. at 1600° F., water quenched.
Heated 5 hr. at 1240° F., air cooled.



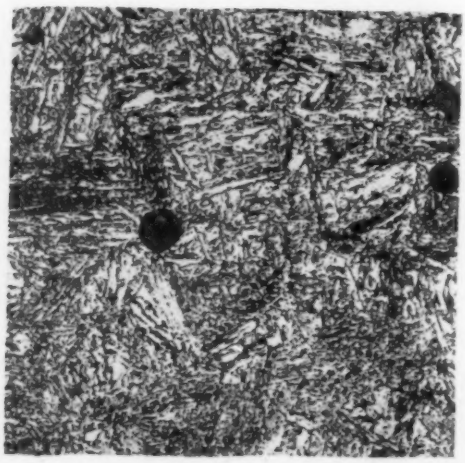
Steel 9982 0.008 per cent Ti



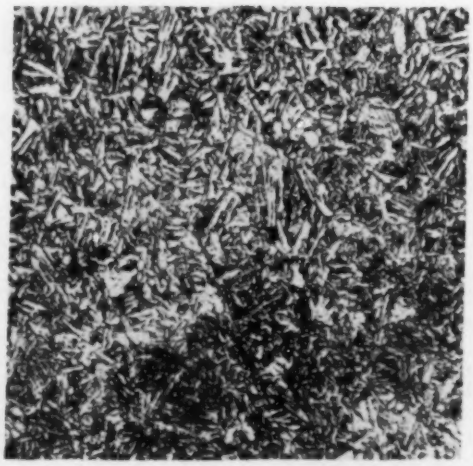
Steel 252 0.002 per cent Al



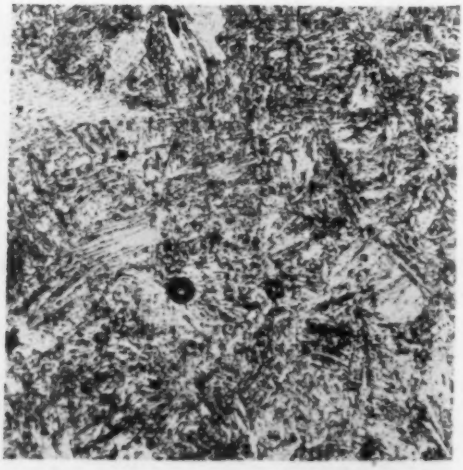
Steel 131 0.008 per cent Ti



Steel 1008 0.007 per cent Ti

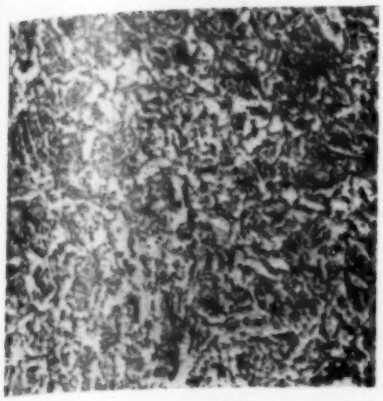


Steel 1141 0.050 per cent Al

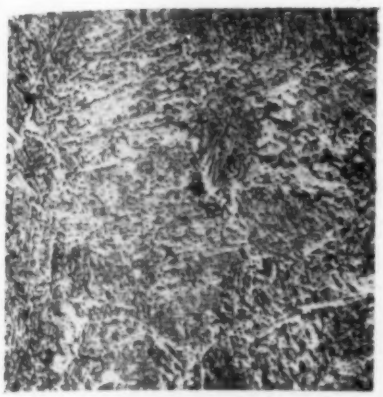


Steel 1739 0.340 per cent Si

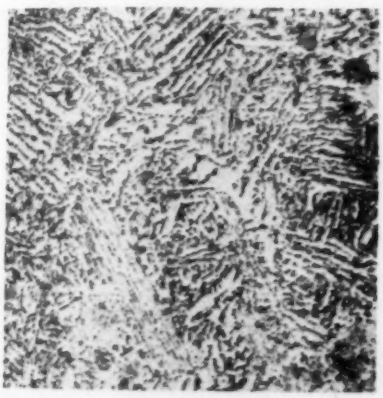
Fig. 5—Microstructures of manganese-molybdenum steels (water quenched from 1600° F. and tempered at 1240° F.). Steels shown in column at left were not so badly embrittled at low temperatures as those shown in column at right. Nital etch. 500X.



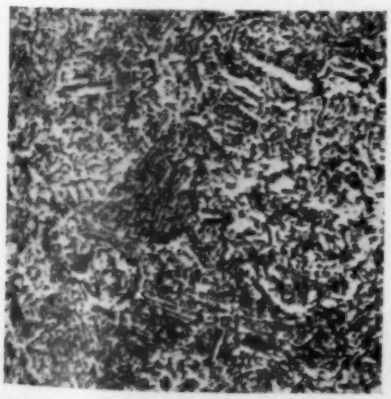
Steel 5 0.04 per cent Al



Steel 11 Silicon killed



Steel 9 0.08 per cent Zr



Steel 6 0.02 per cent Ti

steels, but were not embrittled by temperatures down to −80° F. The fact that these fine-grained steels were inferior in notched-bar properties to the others at temperatures above −40° F. is contrary to results obtained by some other investigators[4,5,6].

Aluminum additions were responsible for the finer grain size and resistance to grain growth exhibited by these steels. Except for the boron steel (No. 536), these materials had lower hardenability than the other group. Steel 252 was so embrittled by film sulphides that decreasing temperatures had practically no effect on the Charpy values.

The data showed that the low-temperature properties of a heat were not reflected consistently by any other property measured. The results also indicated that different

Left — Water quenched 1575° F., tempered 1250° F.

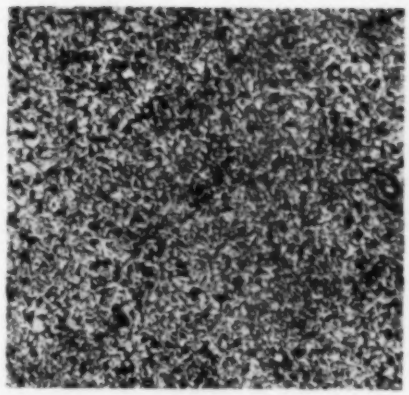
Fig. 6—Microstructures of medium-manganese steels. The two steels at top in each group were least susceptible, and those at the bottom were the most susceptible to low-temperature embrittlement of 11 similar steels tested. Nital etch. Left—500X. Right—100X.

Right — Normalized 1575° F., tempered 1200° F.

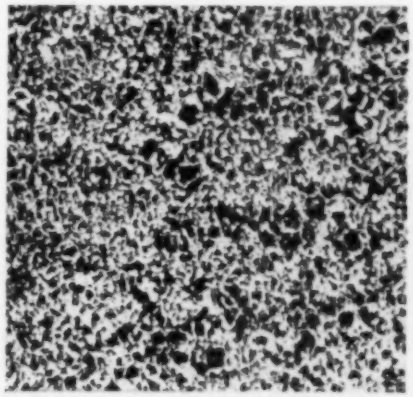
melting methods or slight differences in microstructures did not confer any specific effects measured by notched-bar tests.

All of the aluminum-killed steels had poorer room-temperature Charpy properties than the other steels but were quite resistant to embrittlement by low temperatures. Three of these steels were made in an electric furnace and the fourth was an open hearth steel.

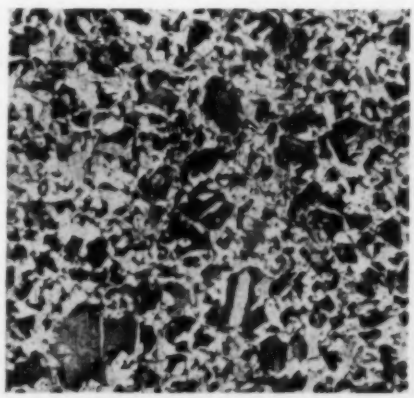
Standard keyhole-notch specimens were used in testing the commercial steels for comparison with results available on other steels of this grade. During this phase of the investigation, some of the materials were also tested with other types of specimens. Figure 7 shows that tests with standard and double-width specimens led to the same conclu-



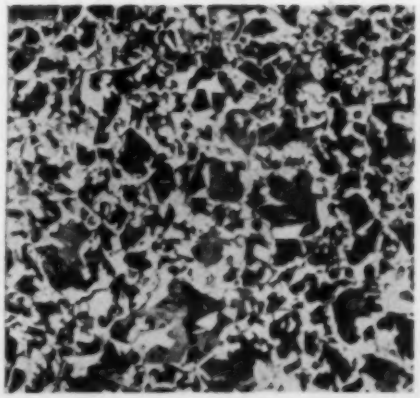
Steel 5 0.04 per cent Al



Steel 19 0.12 per cent Al



Steel 9 0.08 per cent Zr



Steel 3 Silicon killed

Table 7

VEE-NOTCH CHARPY IMPACT VALUES OBTAINED ON 18 CAST STEELS TESTED IN THE NORMALIZED AND DRAWN CONDITION AT TEMPERATURES BETWEEN 75° AND −100° F.

Steel No.	Heat Treatment*	Testing Temperature, ° F.							
		75	32	0	−20	−40	−60	−80	−100
		Charpy Impact Values, ft.-lb.							
GRADE "B" STEELS									
1	a	35, 31, 29, 28	20, 17, 17	8, 8, 6, 5	6, 6, 5, 4	4, 3, 3, 2	—	—	—
2	a	36, 34, 33, 32	19, 19, 16	12, 12, 9	11, 11, 8	7, 7, 7, 6	6, 5, 5, 5	—	—
MEDIUM-MANGANESE STEELS									
3	b	23, 20, 19	13, 12, 12	6, 5, 4, 3	3, 2, 2	6, 2, 2	3, 3, 1, 1	1, 1, 1	—
4	b	25, 25, 25	16, 15, 14	11, 10, 9	7, 7, 6	6, 5, 5, 5	4, 4, 3, 3	4, 3, 3	—
5	b	51, 49, 50	50, 48, 48	35, 34, 31	24, 23, 23	22, 22, 21, 17	17, 16, 16	14, 14, 13, 13	11, 8
19	b	50, 47, 36	31, 27, 22	23, 22, 19	22, 18, 18	13, 12, 11	14, 12, 11, 6	10, 9, 9, 7	7, 6, 6
6	b	25, 25, 24, 24	21, 18, 17	15, 14, 13	10, 9, 9	9, 8, 7, 7	7, 5, 5, 4	3, 3, 2	—
7	b	26, 25, 24, 22	22, 22, 21	16, 15, 15	12, 11, 11	8, 8, 7, 6	6, 6, 5, 4	—	—
18	b	14, 13, 12, 11	11, 10, 10	10, 9, 8, 7	7, 7, 7, 5	5, 5, 5, 4	5, 4, 4, 4	—	—
8	b	32, 29, 29, 29	20, 19, 16	14, 13, 10, 10	11, 8, 8	11, 10, 10, 7	5, 4, 4, 3	—	—
9	b	24, 22, 22, 22	13, 13, 12	7, 7, 7	6, 5, 5	4, 3, 3	3, 2, 2	—	—
10	b	38, 38, 34, 30	23, 22, 22	16, 12, 12	14, 11, 8	8, 8, 7	8, 8, 5	4, 4, 4	—
11	b	23, 23, 23, 21	16, 15, 14	12, 9, 8	9, 8, 6	8, 7, 5	6, 5, 5	5, 4, 4	—
1.55 PER CENT NICKEL OR COPPER STEELS									
12	b	32, 32, 29, 29	22, 20, 19	18, 16, 15, 13	14, 14, 10, 8	11, 7, 6, 5	5, 5, 4, 4	—	—
13	b	37, 37, 32, 28	25, 25, 24	16, 17, 16	16, 12, 12	11, 10, 10	12, 12, 10	6, 5, 4	—
15	b	30, 29, 25, 24	17, 17, 16	12, 12, 10	10, 9, 8	7, 7, 6	11, 11, 9	4, 3, 3	—
COPPER-MANGANESE-SILICON STEELS									
16	c	17, 14, 14, 13	12, 10, 9	12, 7, 6	8, 8, 6	5, 3, 3	3, 2, 2	—	—
17	c	30, 30, 27	22, 18, 17	14, 14, 12	13, 11, 11	11, 10, 9	8, 8, 7	7, 6, 4	4, 4, 3

*a. 1 hr. 1625° F., air cooled, tempered 2 hr. 1200° F., air cooled.
*b. 1 hr. 1575° F., air cooled, tempered 2 hr. 1200° F., air cooled.
*c. 1 hr. 1600° F., air cooled, tempered 2 hr. 1000° F., air cooled.

Table 8

VEE-NOTCH CHARPY IMPACT VALUES OBTAINED ON 18 CAST STEELS TESTED IN THE QUENCHED AND DRAWN CONDITION AT TEMPERATURES BETWEEN 75° AND −100° F.

Steel No.	Heat Treatment	Testing Temperature, ° F.							
		75	32	0	−20	−40	−60	−80	−100
		Charpy Impact Values, ft.-lb.							
GRADE "B" STEELS									
1	d	26, 26, 22, 20	16, 15, 10	11, 10, 9, 5, 4	12, 10, 8, 8	3, 3, 3, 2	—	—	—
2	d	47, 41, 39, 39	24, 23, 23	16, 15, 14	14, 13, 11	11, 11, 11	9, 9, 8, 7	6, 5, 5, 4	4, 4
MEDIUM-MANGANESE STEELS									
3	e	75, 60, 58	59, 49, 48	36, 33, 32	32, 32, 28	21, 19, 18, 17	23, 23, 20, 16	20, 16, 12, 11	14, 9
4	e	28, 27, 24	27, 26, 25	25, 24, 20, 19	17, 17, 17	14, 12, 11	10, 10, 9, 7	9, 8, 7, 7	6, 4
5	e	56, 55, 53	55, 50, 50	51, 48, 43, 39	46, 36, 35	33, 27, 24	22, 21, 20, 20	18, 18, 16, 15	13, 13, 12
19	e	51, 46, 29	44, 29, 25	41, 33, 30	25, 23, 19	23, 22, 22	21, 19, 14, 13, 13	18, 16, 12, 11	7, 7, 7
6	e	29, 28, 22	26, 23, 22	27, 25, 24, 17	15, 13, 13	16, 12, 10, 10	8, 7, 7, 6	7, 6, 6, 4	5, 4, 4
7	e	29, 22, 20, 19	27, 21, 18	17, 17, 17	24, 23, 13, 13	11, 10, 10	7, 7, 7, 6	6, 5, 5, 4	5, 4, 4
18	e	10, 10, 9, 9, 9	9, 9, 8	7, 7, 6, 6	7, 6, 6, 6	5, 5, 4, 4	5, 4, 4	—	—
8	e	49, 49, 47, 44	39, 34, 33	25, 23, 21	20, 20, 17	21, 16, 14, 13	17, 16, 16, 14	11, 11, 11, 6	11, 7, 6
9	e	46, 38, 38	48, 31, 20	29, 25, 15	21, 9, 8	6, 6, 6	18, 13, 4	12, 4, 3	7, 5, 2
10	e	44, 40, 35, 27	25, 25, 20	20, 20, 16	17, 16, 14	21, 14, 14	11, 9, 6	6, 5, 4	—
11	e	56, 53, 51	43, 41, 41	31, 30, 29	35, 37, 26	23, 21, 21	23, 20, 15	18, 16, 12	11, 10, 8
1.55 PER CENT NICKEL OR COPPER STEELS									
12	e	32, 32, 30, 30	22, 22, 21	20, 20, 18, 16	19, 18, 13, 12	15, 13, 12, 8	13, 10, 10, 7	6, 4	—
13	e	36, 35, 33	31, 30, 26	26, 26, 20	20, 16, 15	14, 13, 13	12, 10, 8	9, 8, 7	9, 6, 6
15	e	38, 34, 30	36, 34, 34	21, 20, 20	17, 17, 15	16, 13, 12	5, 4, 4	11, 9, 8	6, 5, 4
COPPER-MANGANESE-SILICON STEELS									
16	f	25, 21, 21	16, 15, 15	12, 12, 11	13, 13, 12	11, 10, 7	10, 8	9, 7, 7	7, 5, 3
17	f	38, 38	30, 26, 23	20, 20, 19	16, 14, 12	14, 14, 11	14, 12, 11	11, 10, 9	10, 8, 8

HEAT TREATMENT

d. 1 hr. 1625° F., water quench, tempered 2 hr. 1250° F., air cooled.
e. 1 hr. 1575° F., water quench, tempered 2 hr. 1250° F., air cooled.
f. 1 hr. 1600° F., water quench, tempered 2 hr. 1000° F., air cooled.

Table 9

COMPARISON OF THE LOW-TEMPERATURE NOTCHED-BAR PROPERTIES OF MANGANESE-MOLYBDENUM CAST STEELS WITH OTHER CHARACTERISTICS*

Heat No.	Hardness Rockwell "B"	Hardness Brinell	Residual Deoxidizer, per cent	Sulphide Inclusion Type	Austenite Grain Size 1600° F.	Austenite Grain Size 1750° F.	Austenite Grain-Growth Characteristics	Calculated Hardenability Ideal Diameter, in.	Minimum Charpy Values,° ft.-lb. 75° F.	0° F.	−40° F.	−80° F.	Other Values	Break in Impact vs. Temperature Curve
158	94	202	0.008 Ti	I	5	1 & 5	Coarse grain readily coarsened further.	2.82	38	35	33	26	—	None
9982	96	212	0.008 Ti	I	5	5	Coarse grain, not readily coarsened further.	2.60	39	33	27	23	—	None
131	98	225	0.008 Ti	I	5	5		2.58	33	32	29	23	—	None
1092[b]	96	212	Si	I	6	5	Fine grain, not readily coarsened.	2.25	32	28	24	22	—	None
608	91	189	0.033 Al	III	8	7		2.23	26	24	21	16	—	None
1141	97	219	0.050 Al	III	9	9		2.22	25	27	21	19	—	None
536[c]	96	212	0.089 Al	III	8	8		2.58	23	23	20	18	12 at −70° F.	−70° F.
252	98	225	0.002 Al	II	8	7		2.28	15	15	14	11	—	None
1261	92	195	0.009 Ti	I	5	1 & 5	Coarse grain readily coarsened further.	2.52	33	32	29	4	—	−70° F.
1008	96	212	0.007 Ti	I	5	3		2.74	37	32	31	3	—	−50° F.
1739	92	195	Si	I	6	1	Fine grain, readily coarsened further.	2.28	37	33	5	4	—	−40° F.

*All steels were of closely similar composition and were water quenched from 1600° F. and tempered at 1240° F.
a. Tests made on standard keyhole-notch Charpy specimens.
b. The keel blocks from this heat exhibited pinhole porosity.
c. Steel 536 also contained boron.

sions about the properties of steel No. 1008.

Both types of samples showed a fairly uniform drop in test values between 32° F. and −20° F., and considerable scatter at lower temperatures. Although duplicate standard specimens gave good checks at −30° F. and −40° F., a variation would have been encountered if more specimens had been tested.

Plots in Fig. 8 give the data for keyhole and Vee-notched bars of steels Nos. 1092 and 1141. The values for the keyhole specimens decreased uniformly from 30 ft.-lb. at room temperature to about 18 ft.-lb. at −80° F. The drop was about one-third of the room-temperature value for both steels. Values for the Vee-notch specimens started at a higher level for room temperature and decreased about one-half to the same level as the keyhole specimens at −80° F. These data also show that neither steel was temper brittle when heat treated as a keel block 1¼ in. square.

Tests were also made on two wrought steels to determine the effect of specimen dimensions on notched-bar properties. These steels were similar to cast steels in composition but were forged and rolled to ½-in. plate for convenience in obtaining a large number of specimens. The compositions, heat treatment, and tensile properties of the two materials are given in Table 10. The data in Figs. 9 and 10 show that the keyhole-notched specimens gave reasonable agreement in all cases, and the Vee-notched samples of the carbon steel also gave good checks. Data for all types of specimens show that the toughness of both steels decreased gradually with decreasing temperature and that the carbon steel was inferior to the other material. The great variation in values for Vee-notched bars of the manganese-molybdenum steel may indicate that it is more sensitive to unintentional variations in specimen or test conditions.

Specimen Shapes

Work on specimens of different shapes indicated that tests with either of the standard-width Charpy specimens would give the same information concerning low-temperature embrittlement. In most of the tests, the Vee-notched bars gave higher test values at room temperature than keyhole specimens, and equally small values for brittle fracture.

It was concluded, therefore, that the Vee-notched specimens were preferable, because they gave a wider range in values. It seemed desirable to test a larger number of specimens when studying the laboratory steels because of the likelihood of encountering greater scatter with the Vee-notched specimens.

Table 11 gives a comparison of the minimum Charpy values of the laboratory steels and lists the other characteristics of the materials. The steels are grouped on the basis of chemical composition instead of resistance to embrittlement because several grades are represented.

None of the laboratory steels was as tough at low temperatures as the commercial steels containing molybdenum. However, the room-temperature values were normal for the

Table 10

MECHANICAL PROPERTIES AND COMPOSITION OF TWO WROUGHT STEELS USED FOR STUDY OF EFFECT OF SPECIMEN SHAPE ON CHARPY NOTCHED-BAR VALUES

Steel No.	Mechanical Properties* Ultimate Strength, psi.	Yield Strength, psi. (0.2% offset)	Elongation in 2 in., per cent	Brinell Hardness	Composition, per cent C	Mn	Mo	Si	S
10490	93,000	74,250	27.5	185	0.28	0.90	0.30	0.35	0.035
10491	78,000	47,750	36.0	158	0.27	0.67	—	0.40	0.037

*Values are average of two tests on 2.00x0.500x0.500-in. gage-length specimens.
Steel 10490 was heated 2 hr. at 1600° F., water quenched, tempered 5 hr. at 1240° F., and air cooled.
Steel 10491 was heated 2 hr. at 1650° F., and air cooled.

Table 11

COMPARISON OF THE LOW-TEMPERATURE NOTCHED-BAR PROPERTIES OF VARIOUS EXPERIMENTAL STEELS MADE WITH DIFFERENT DEOXIDATION PRACTICE AND TESTED IN TWO CONDITIONS OF HEAT TREATMENT

Casting No.	Hardness, Rockwell "B" N.,D.	Q.,D.*	Residual Deoxidizer, per cent	Sulphide Inclusion Type	Austenite Grain Size 1600° F.	1750° F.	Austenite Grain-Growth Characteristics	Hardenability Ideal Diameter, in.	Minimum Charpy Values, ft.-lb. Quenched and Drawn, °F. 75	0	−40	−80	Normalized and Drawn, °F. 75	0	−40	−80
GRADE "B" STEELS																
1	71	77	Si	I	7	2	Fine, readily coarsened.	0.80	20	4	2	—	28	5	2	—
2	78	82	0.053 Al	III	7	5	Fine, less readily coarsened.	0.83	39	14	10	3	31	9	5	—
MEDIUM-MANGANESE STEELS																
3	88	89	Si	I	5	4	Coarse, readily coarsened.	1.74	58	32	17	11	19	3	2	1
4	88	89	0.011 Al	II	7	6	Fine, resistant to coarsening.	1.54	24	19	11	7	25	9	5	3
5	88	90	0.038 Al	III	9	8		1.36	53	39	24	15	45	31	17	13
19	84	90	0.150 Al	III	9	8-5		1.63	29	30	22	11	36	19	11	7
6	89	91	0.020 Ti	II	9	8		1.13	22	17	10	4	24	13	7	2
7	92	91	0.070 Ti	II	8	8		1.32	19	17	10	4	22	15	6	—
18	84	90	0.160 Ti	II	9	9		1.64	9	6	4	—	11	7	4	—
8	88	88	0.053 Zr	III	6	4	Fine, readily coarsened.	1.45	44	21	13	6	29	10	7	—
9	80	89	0.080 Zr	III	6	4		1.35	38	15	6	3	22	7	3	—
10	84	92	0.11 V	I	7	3-4	Fine, readily coarsened.	1.53	27	16	14	4	30	12	7	4
11	84	91	Ca added	I	5	3	Coarse, readily coarsened.	1.65	51	29	21	12	21	8	5	4
1.55 PER CENT NICKEL OR COPPER STEELS																
12	78	82	Si	I	6	3	Fine, readily coarsened.	1.40	30	16	8	4	28	13	5	—
13	80	85	0.04 Al	III	8	7	Fine, resistant to coarsening.	1.47	33	20	13	7	24	16	10	4
15	86	88	0.02 Al	III	8	7		1.35	30	20	12	8	24	10	6	3
PRECIPITATION HARDENED, COPPER-MANGANESE-SILICON STEELS																
16	92	98	Si	I	7	3-4	Fine, readily coarsened.	1.61	21	11	7	7	13	6	3	—
17	93	86	0.02 Al	III	8	8	Fine, resistant to coarsening.	1.65	38	19	11	9	27	12	9	4

NOTE: Tests made on standard vee-notch Charpy specimens.
*N., D.—Normalized and drawn.
*Q., D.—Quenched and drawn.

different grades and comparable to results obtained on keel-block samples.

Quenching usually produced better notched-bar properties at room temperature than the normalizing treatment. The superiority of the quenched and drawn samples was more pronounced at lower testing temperatures. The aluminum-treated steels usually were tougher at low temperatures than their silicon-killed comparison steels, although they were not always superior at room temperature.

Aluminum-killed steels Nos. 5 and 19 were almost as good in the normalized and drawn condition as they were as quenched and drawn. These two medium-manganese steels had the best low-temperature properties of all the laboratory steels. Steel No. 4 with a smaller residual aluminum content was poorer, because it was embrittled by film sulphides. The silicon-killed steels Nos. 3 and 11 were tough at low temperatures in the quenched and drawn condition but brittle as normalized and drawn.

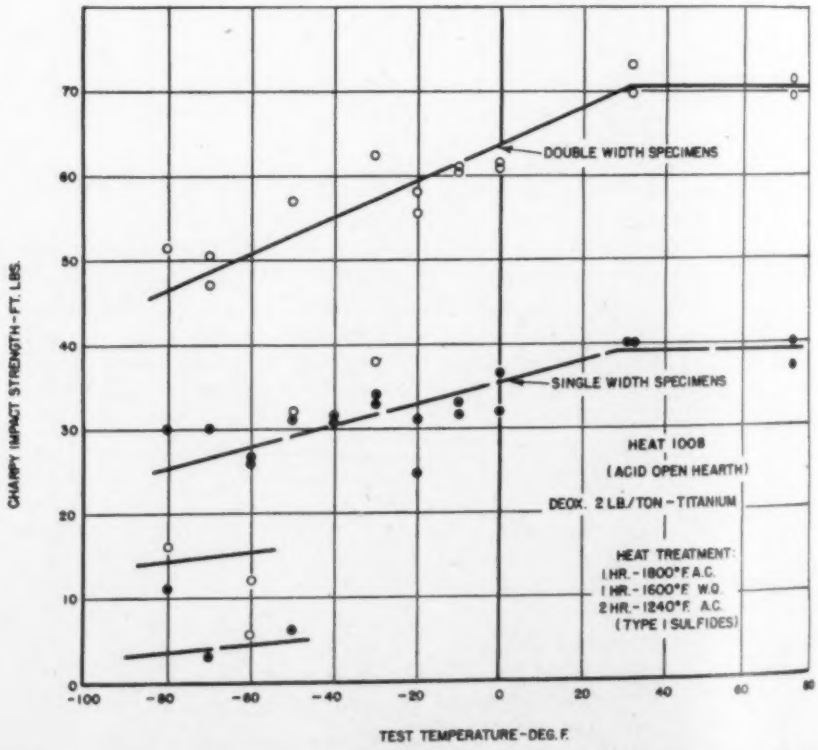


Fig. 7—Low-temperature notched-bar impact strength (Charpy keyhole single and double width specimens.)

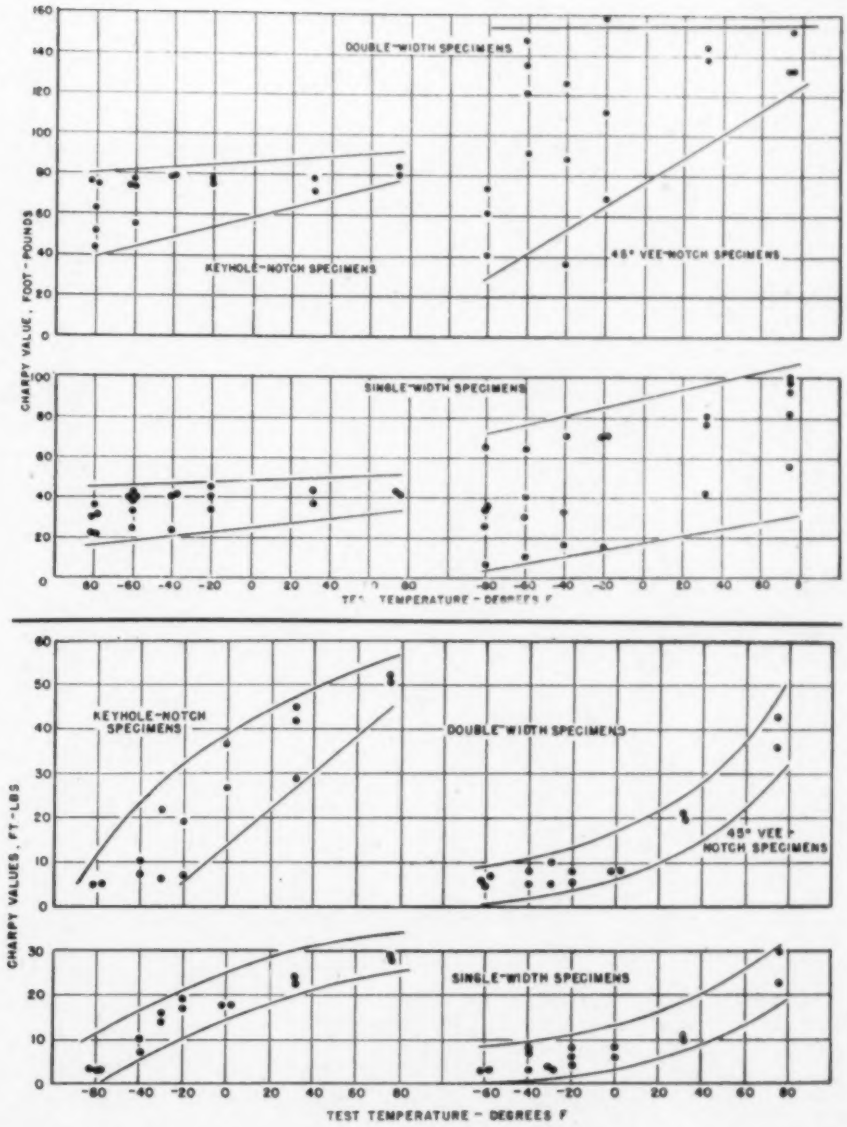


Fig. 9 (above)—Low-temperature notched-bar properties of a normalized, medium-carbon rolled steel.

Fig. 10 (right)—Low-temperature notched bar properties of a quenched and tempered manganese-molybdenum rolled steel.

This behavior may have been related to their greater hardenability, but none of the other data indicated any relationship between this factor and low-temperature notched-bar properties. The effect of heat treatment on the toughness of zirconium steel No. 8 was similar to that on the silicon-killed steels, although it did not have a high hardenability.

Notched-bar properties of the steels with 1.55 per cent nickel or copper were not so good as those of the better medium-manganese steels. In the quenched and drawn condition, the copper steel was as resistant to low-temperature embrittlement as the nickel steel and nearly as good when normalized and drawn.

Precipitation-hardened steel deoxidized with aluminum (No. 17) had Charpy properties equivalent to the nickel steels, although it had a much higher tensile strength. The Grade "B" steels were the poorest of the materials tested.

Fractures of Notched-Bar Specimens. When notched bar impact specimens break with a completely brittle fracture, they show no deformation, and the fracture surfaces often have a crystalline texture. Tough specimens undergo considerable plastic deformation before rupture and often show a fibrous fracture.

Fractures with a crystalline area in the center surrounded by material which deformed before rupture are often observed in the so-called transition zone between ductile and brittle fracture[7]. These facts have led some investigators to conclude that the appearance of the fracture is as important as the energy values in interpreting the results of notched-bar impact tests.

In this investigation, the appearance of fractures of the test bars was noted, and photographs were taken

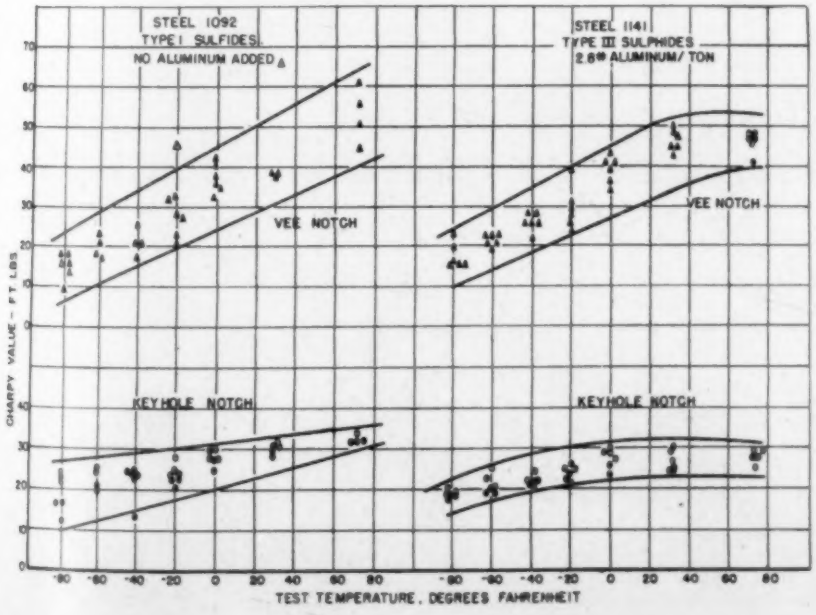


Fig. 8—Low-temperature notched-bar properties of two manganese-molybdenum cast steels in the quenched and tempered condition. Solid points represent bars furnace cooled from 1240° F. draw; open points represent air-cooled samples.

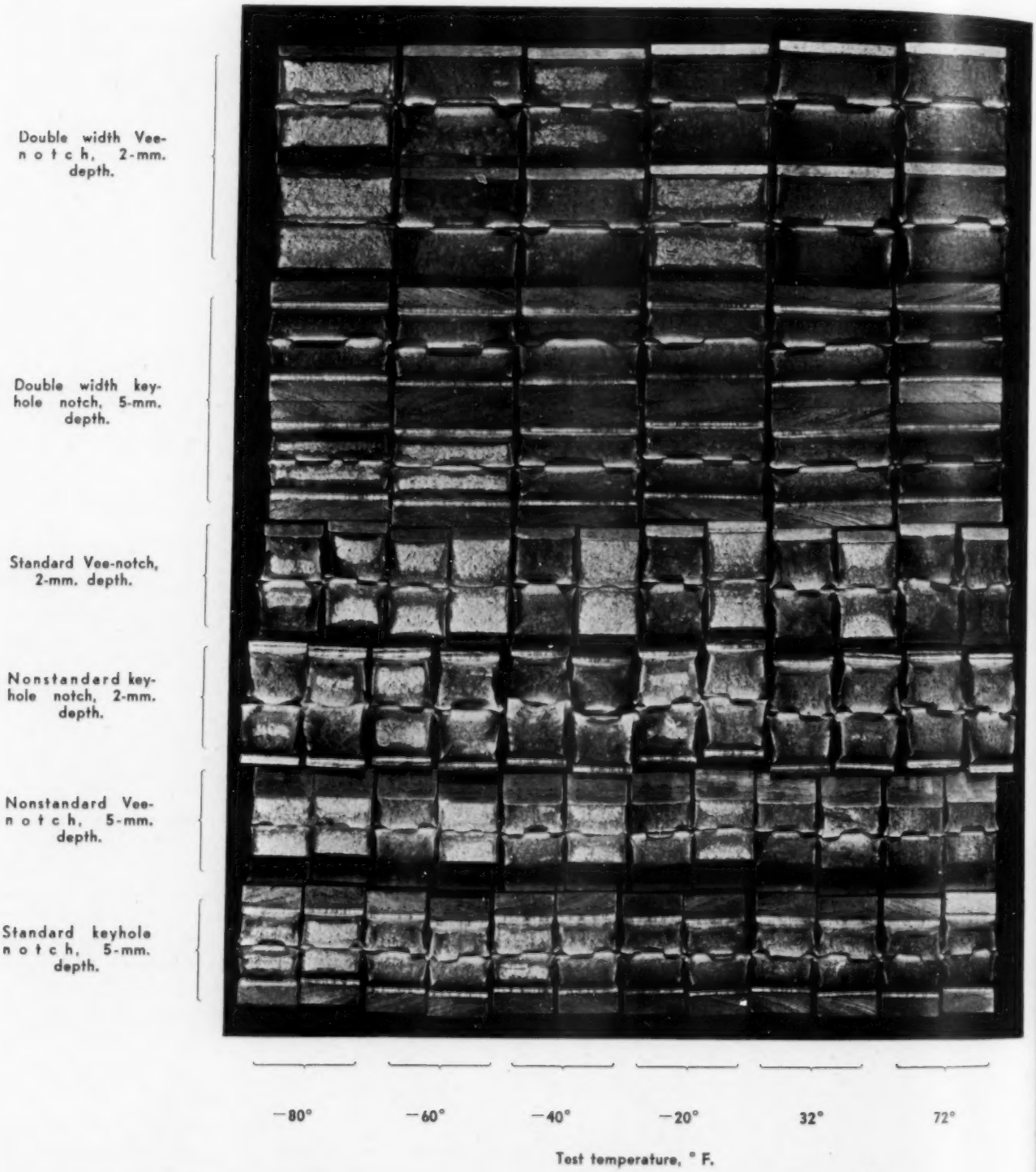


Fig. 11—Appearance of notched-bar fractures for different kinds of specimens from a rolled manganese-molybdenum steel (10490) in the quenched and tempered condition. Actual size.

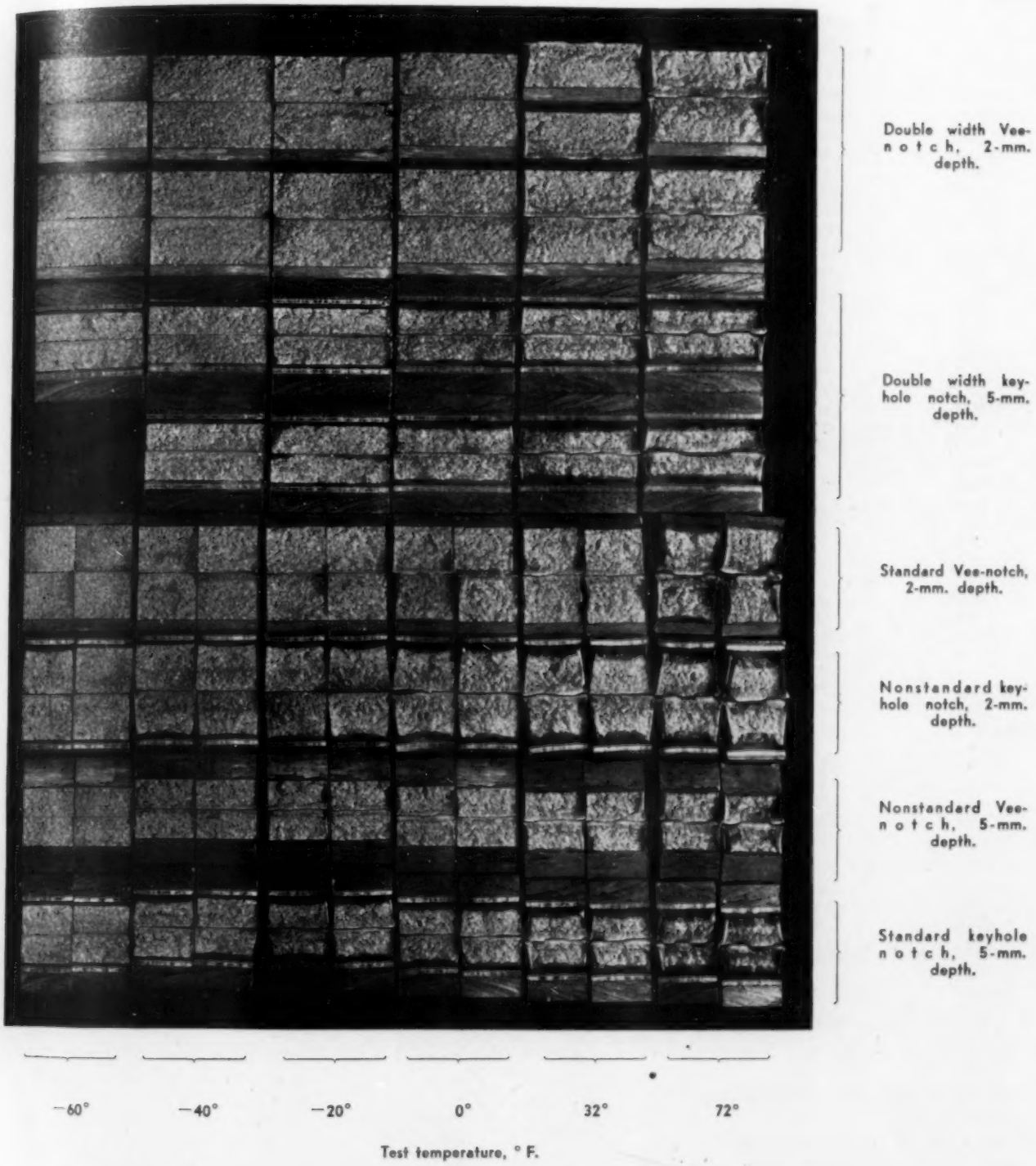


Fig. 12—Appearance of notched-bar fractures for different kinds of specimens from a rolled medium-carbon steel (10491) in the normalized condition. Actual size.

Table 12

HIGHEST TEMPERATURES AT WHICH NOTCHED-BAR FRACTURES SHOWED CRYSTALLINE TEXTURES OR NO APPRECIABLE DEFORMATION AND THE CORRESPONDING CHARPY VALUES

Steel and Treatment*	Start of Crystalline Fracture		Deformation Characteristics	
	Temp., ° F.	ft.-lb.	Temp., ° F.	ft.-lb.
STANDARD CHARPY KEYHOLE SPECIMENS				
536Q	None	—		17.5
131Q	—50	29		23
1261Q	—30	24	Deformed at —80	20
10490Q	—40	23		20
252Q	—80	12		12
1008Q	—10	32	No deformation at —70	3
10491N	+75	28	No deformation at —60	3
STANDARD CHARPY VEE-NOTCH SPECIMENS				
5Q	—20	35	Deformed at —100	12
5A	+32	48	No deformation at —100	8
11Q	—20	26		12
8Q	—20	17	No deformation at —80	6
10490Q	+32	42		6
3Q	—20	28		11
13Q	+75	33		8
10Q	+75	27		6
6Q	—20	13	No deformation at —60	6
7Q	—40	10		6
4Q	0	19		7
17Q	+75	38		11
16Q	+75	21	No deformation at —40	7
13A	—20	12		10
9Q	+75	38		6
7A	—30	8		11
15A	+75	24		8
2Q	—20	11		11
2A	—20	8		8
10491N	+75	23		4
17A	—20	11		11
6A	—20	9	No deformation at —20	9
8A	+75	29		8
10A	+75	30		8
11A	+75	21		6
4A	0	9		6
9A	+32	12		5
3A	+75	19		2
16A	+75	13		11
1A	+75	28	No deformation at 0	5
1W	+75	20		4
DOUBLE-WIDTH KEYHOLE SPECIMENS				
10490Q	—60	55	Deformed at —80	44
1008Q	0	62	No deformation at —60	5.5
10491N	+75	51	No deformation at —20	7
DOUBLE-WIDTH VEE-NOTCH SPECIMENS				
10490Q	—20	60	Deformed at —80	41
10491N	+75	36	No deformation at 0	8

*Heat Treatment. Q—Quenched and tempered; A—Normalized and tempered; N—Normalized.

for record purposes. The photographs in Figs. 11 and 12 show the appearance of the fractures of two steels which were tested at various temperatures with six kinds of specimens.

Figure 11 shows that the crystalline texture appeared at 32°, −20°, −40°, and −60° F., depending upon the shape of the specimen, although all samples deformed considerably at −80° F., the lowest test temperature employed. The standard Vee-notch specimens, which showed a crystalline area at 32° F., deformed considerably more at −80° F. than the double-width Vee-notched and the 5 mm. Vee-notched bars which did not show the crystalline fractures at temperatures above −20° F.

Fracture appearance of another steel is shown in Fig. 12. All six types of specimens gave crystalline fractures at room temperature, although the temperature at which rupture occurred with no deformation varied wtih the specimen shape. The temperature of completely brittle failure varied from 0° to −60° F. for this steel, depending upon the shape of the specimen.

Figure 13 shows that the crystalline fracture appeared in steel No. 1008 at 0° F. in the double-width specimens and at −10° F. in the standard keyhole samples. However, all of the specimens showed some deformation in tests at −80° F.

Data in Table 12 summarize the observations on the temperatures at which crystalline patches, and completely brittle fractures first appeared in notched-bar tests on various steels. It is apparent that crystalline textures were found in samples which gave widely different test values.

Temperatures at which the crystalline patches appeared in a steel bore no relationship to the temperature at which the steel failed without appreciable deformation. The data also show that all of the samples which broke without appreciable plastic deformation, gave test values of 12 ft.-lb. or less. It is believed that much of the energy required for rupture of the specimens which broke with a brittle fracture was absorbed by elastic deformation and in knocking the pieces from the anvil.

Conclusions

1. Susceptibility to low-temperature embrittlement of steels of the same grade was not consistently related to differences in melting practice, deoxidation treatment, inclusion distribution, grain size, or grain-growth characteristics. The slight variations found in hardenability and microstructure of similar steels were without apparent effect on the notched-bar properties.

2. Relative resistance of various steels to low-temperature embrittlement bore no relationship to their room-temperature tensile or notched-bar properties.

3. Quenching usually developed better notched-bar properties than normalizing before drawing to the same strength level. The aluminum-killed steels, which were tough as normalized, did not evidence this response to heat treatment.

4. Fine-grained steels often gave poorer Charpy values at room temperature than coarse-grained steels.

5. In general, the aluminum-killed steels showed less embrittle-

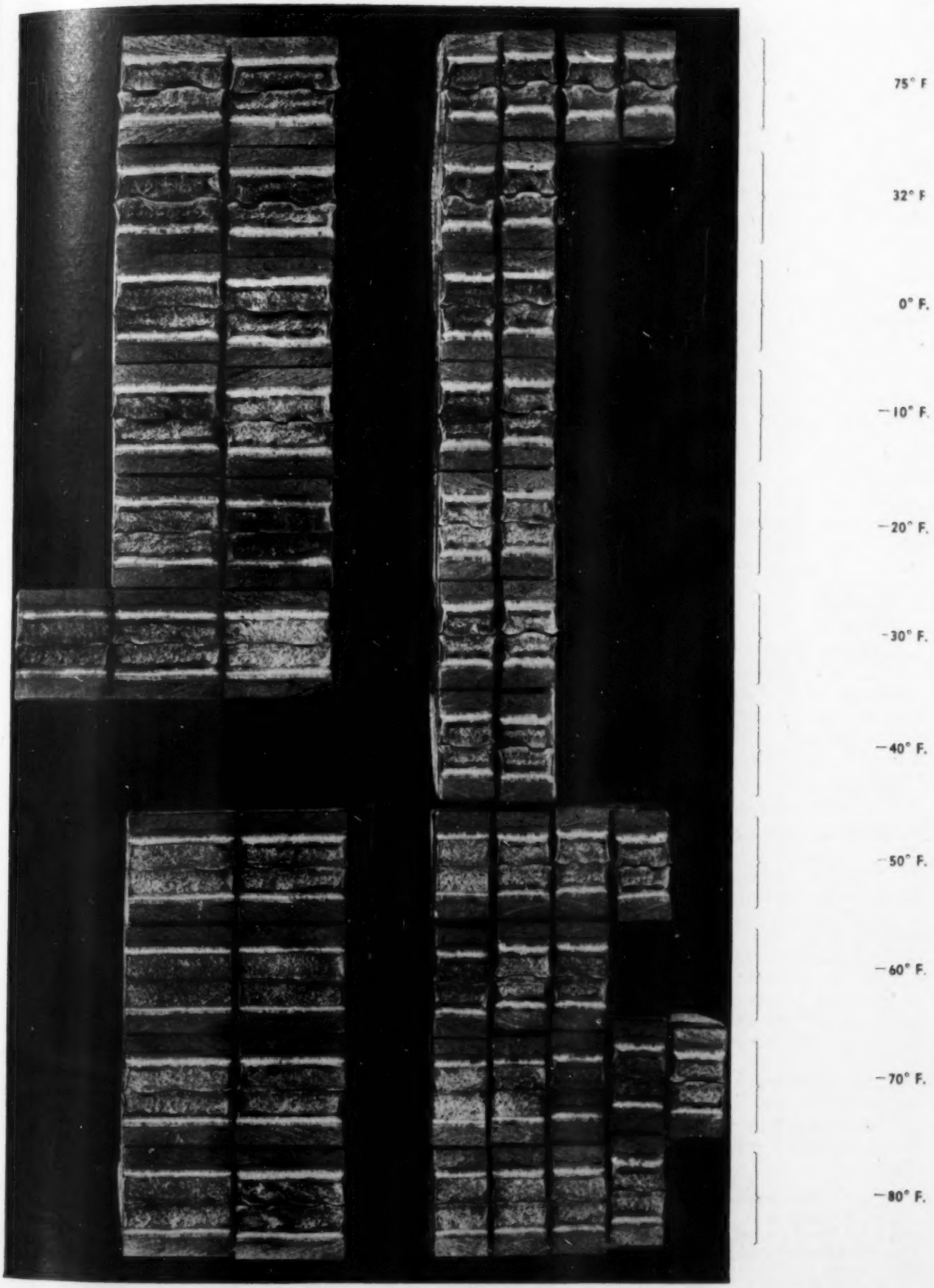


Fig. 13—Appearance of fractures of notched-bar impact specimens of Steel 1008 tested at various temperatures. Actual size.

ment by low temperatures than similar materials deoxidized with other elements.

6. Film sulphide, Type II, inclusions produced poor ductility and Charpy properties at room temperature, but the embrittlement was less noticeable at low temperatures.

7. Steels containing less than 0.01 per cent titanium had globular Type I sulphides; those containing 0.02, 0.07, and 0.16 per cent titanium contained chain or film Type II sulphide inclusions. The undesirable eutectic sulphides persisted in steels containing much higher titanium contents than those usually employed. It appears, therefore, that the action of excess titanium on sulphide distribution is different than the action of aluminum of zirconium.

8. In the steels studied, copper seemed to be as beneficial as nickel in conferring resistance to low-temperature embrittlement. The results on medium-manganese and Grade "B" steels suggested that manganese also improved low-temperature toughness.

9. Porosity encountered in some castings was attributed to gases picked up from core-sand molds. Higher manganese contents, or additions of strong deoxidizers were sufficient to prevent the center-line gas porosity.

Acknowledgment

This investigation was made possible by an appropriation from a fund for research and education established at Battelle Memorial Institute by the Ohio Steel Foundry Co. This fund is to be used for work at the Institute on the technology of the steel foundry, the general dissemination of knowledge in this field, and the education of men skilled in the foundry art. The cooperation of H. R. McCoy and Frank Kiper of the Ohio Steel Foundry Co. in furnishing information and materials, and the assistance of J. A. Finley of the Battelle staff are gratefully acknowledged.

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DISCUSSION

Chairman: J. W. JUPPENLATZ, Lebanon Steel Foundry, Lebanon, Pa.

Co-Chairman: J. B. CAINE, Sawbrook Steel Castings Co., Lockland, Ohio.

CHAIRMAN JUPPENLATZ: What size of specimen did you use prior to heat treatment?

MR. BOULGER: One and one-half inch square bars were cut from keel blocks in the case of the commercial steels. The laboratory steels were treated as cast plates, 1x6x6 in., examined for soundness, and heat-treated without being cut into smaller sections.

CHAIRMAN JUPPENLATZ: Would it be possible to obtain more martensite in one than in the other?

MR. BOULGER: Yes.

C. W. BRIGGS[1]: What should be the proper residual aluminum content in order to obtain maximum properties?

MR. BOULGER: We have been investigating how the type of sulfide varies with manganese content and other elements, instead of depending exactly on the residual aluminum. For satisfactory distribution of the sulfides, the aluminum should be in excess of about 0.012 per cent. If you are using zirconium, you should use enough to get 0.03 per cent of zirconium (total).

MR. BRIGGS: Is there an upper limit?

MR. BOULGER: As far as we know, not as far as inclusions are concerned.

C. E. SIMS[2]: There probably is an upper limit, but for round numbers, about 0.02 per cent residual aluminum in Grade B steel is the minimum. Up to 0.15 per cent residual aluminum does not seem to affect the properties to any marked extent. Probably no one would add enough to give more than 0.15 per cent.

J. A. BOWERS[3]: We occasionally run into trouble, having the residual aluminum over 0.12 per cent.

CHAIRMAN JUPPENLATZ: Does that apply to low temperature or room temperature impact tests?

MR. BOWERS: It applies to room temperature tests.

DR. C. H. LORIG[4]: What was the carbon content? Was it Grade B steel?

MR. BOWERS: Not all of it. Most of our trouble was with high manganese steel.

MEMBER: Were the impact specimens given the same heat treatment in all cases for all steels?

MR. BOULGER: No. This depended upon the grade of steel. When we tested manganese-molybdenum steels, they were all tested under the same conditions of heat treatment.

MEMBER: We found some steels specified have a minimum of 15 ft.-lb. Charpy at 100° C., and by changing the heat treatment a little bit it might be raised from two or three ft.-lb. to well over 30 ft.-lb., and still be within the limits of commercial heat-treated impacts and be entirely satisfactory. I am just wondering whether you have tried anything like that. I did not notice any work here that indicated a transition from the high to the low impact. Were all of them within relatively straight lines?

MR. BOULGER: We did not find transitions on some steel we had, and on others we did. Between room temperature and —100° C., very often we did not encounter any sharp transition temperature.

MEMBER: You are limiting yourself to the 0.30 carbon steel?

MR. BOULGER: Yes.

CHAIRMAN JUPPENLATZ: It was mentioned that a slight change in heat treatment caused improvement. That should be explained in more detail.

MEMBER: We had a specification specifying a normalizing from 1650° F. and a draw at 1200° F. By cutting the normalizing temperature by 50° F., and dropping the draw by the same amount, we changed it from the very low impact up to the high level, which is probably ten times as high.

CHAIRMAN JUPPENLATZ: Did you have higher initial room temperature properties or higher low temperature properties?

MEMBER: We had higher low temperature properties. It definitely brings it from the zone of the low impact up to the zone of high impact.

CHAIRMAN JUPPENLATZ: Enough to get you over the 15 ft.-lb. requirements?

MEMBER: Oh, definitely. I do not think any of them ran less than about 26 ft.-lb.

CHAIRMAN JUPPENLATZ: Was that a wrought or cast?

MEMBER: Wrought.

CHAIRMAN JUPPENLATZ: Probably what you were doing there was changing the structural characteristics of that material, approaching a better structure, somewhat analogous to a mild quench.

MEMBER: The structure does not seem to determine it. There must be some change, otherwise we would not get the enormous change at that low temperature.

CHAIRMAN JUPPENLATZ: In practically every case, if specimens are drastically quenched, (there hardenability enters the picture), the temperature decreases for the transition point from a ductile to a brittle failure.

DR. LORIG: In steels of higher hardenability, it is possible to quench and temper and thus avoid a serious drop in notched-bar impact resistance at lower

temperature provided, of course, the structure is tempered martensite. It was not covered in the paper, but it is a significant fact that the structural characteristics of the steel have an important effect on low-temperature notched-bar properties.

A. P. GAGNEBIN[5]: If I understood correctly, it was indicated that there was no correlation between the impact level and the amount of deformed metal in the fracture, is that correct?

MR. BOULGER: That is not exactly correct. We found that when there was no deformation, all the values were below 12.

MR. GAGNEBIN: I thought you had a chart, showing the relation between the area of brittle zone in the fracture and the impact level, which indicated there was no correlation between these two factors. Did I misunderstand?

MR. BOULGER: The chart showed the lack of relationship between the temperature at which we first got this crystalline texture and the temperature at which the sample ceased to give any deformation.

In sample "5A," for example, the brittle texture—that is the crystalline instead of fibrous fracture texture—first appeared at 32° F., but the sample still showed deformation at —80° F. There was no relationship between the texture and the amount of deformation.

MR. SIMS: The amount of deformation and the ft-lb. value did correlate. The amount of deformation and the appearance of the fracture did not correlate; that is, we sometimes had a crystalline type of fracture when we obtained a considerable amount of deformation. We came to the conclusion that, except for the inertia of the specimen, all of the energy absorbed is in deformation of the specimen, and unless you get deformation, there is no energy absorbed.

MEMBER: When we want any apparatus to operate at lower than —100° C. (about —148° F.), we adhere to the 18-8 austenitic chrome-nickel stainless steel.

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[3] Battelle Memorial Institute, Columbus, Ohio.
[5] American Cast Iron Pipe Co., Birmingham, Ala.
[4] Battelle Memorial Institute, Columbus, Ohio.
[6] International Nickel Co., Bayonne, N. J.

UPGRADING LEARNERS

One of the most satisfactory ways to obtain skilled finishers, core-setters and journeymen molders is for the foreman to carry out the responsibilities he has in upgrading and promoting his personnel. The foreman must be able to appraise personnel under his supervision and select, train and upgrade those in whom he sees possibilities.

M. J. Gregory
Formerly Works Manager,
Foundry Div.
Caterpillar Tractor Co.
Peoria, Ill.

ONE OF THE most satisfactory ways to obtain skilled finishers, core-setters and journeymen molders is for the foreman to carry out the responsibilities he has in upgrading and promoting his personnel. Often times his first task is to educate the Employment Department at the plant because too often this department neglects to select carefully personnel to work in the foundry.

The idea often prevails that anyone is capable of working in a foundry. A good man is placed in the Machine Shop or elsewhere. The foundry industry requires personnel with just as high an intellect, perhaps even more, than the Machine Shop.

The foundry must have careful selection of personnel in the beginning if the foreman is to have good material to work with. After careful selection, the foreman must be able to appraise personnel under his supervision and select those in whom he visions possibilities and potentialities.

Daily Instruction

Then through a system of daily contact and instruction, the foreman can call to the attention of his personnel the methods, technical operations and processes which are inherent in the job immediately ahead or the next job to which it may be possible to promote a specific man.

For example, let us consider a man who comes into the foundry and is given the job of unloading core ovens, i.e., taking the racks from the oven, taking the cores out of the driers, placing the cores in position for the cleaners and dippers and taking the empty driers back to the coremaker for re-use. It takes but little instruction from the foreman to point out to a man how cores are used, how they are made, why different cores consist of different sand mixtures and why they are made in different ways.

There are numerous opportunities for the foreman to call to the man's attention various operations necessary for cleaning, dipping, drying, and final assembling of cores. With constant daily contact by the foreman, the man becomes familiar with these various operations. Should an opportunity present itself, this man would be qualified for promotion to a job as a beginning core assembler.

Advanced Assignments

It then is a small step for this man to advance to more complicated core assembly assignments under similar instruction and supervision he had before. Through all these processes he becomes cognizant of all the operations that are required in the making of cores. He can easily comprehend why some cores require more cleaning before assembly than others because of work done imperfectly in core making. Let us suppose a need develops for a coremaker, for example, to make simple cores from a dump box on a bench or the more simple cores on a core blower.

This man is ready to assume such responsibility because he already has the background acquired in upgrading processes which give him the necessary knowledge to start as a coremaker.

He has become conscious of the processes and the reasons for them. It now becomes only a matter of time and constant practice to enable this man to develop the necessary skill which will make him a first class coremaker.

Developing Ability

It may be true that if a man were kept on one particular job of coremaking where it was necessary for him only to make a few cores of the same nature over a long period of time that he would never develop that necessary skill which would make him an all-around coremaker.

But in the normal course of events, different core boxes which require different techniques to make are given to him. In this way he soon acquires experience necessary for developing all-around ability.

As another example, let us start in the Molding Department with a man who comes in as a manual laborer requiring no skills. He slowly becomes familiar with foundry operations by observation.

In a short time he may be put on a job as a molding supply man. Here he will be required to bring supplies such as chaplets, gaggers, sand, etc., to the molders. In this job he becomes increasingly familiar with the operations of molding because he is closer to it.

Presented at an Apprentice Training Session of the Fiftieth Annual Meeting, American Foundrymen's Association, at Cleveland, May 9, 1946.

He becomes familiar with the materials used and how and why they are used so that he is soon ready to be advanced as helper to a machine molder. Here he shovels sand for the molder, sets gaggers, helps set flasks on the machine, helps operate the jolt levers or the squeezer and becomes familiar with all the techniques and all the operations involved in machine molding.

He becomes familiar with the number of jolts it takes to ram the various molds, and why it is necessary to ram them as hard as he is instructed. He becomes familiar with the properties of molding sand and learns the difference between good and bad sand. He becomes familiar with the core prints in the molds themselves and the reason for their presence and use.

He observes cores set by someone else, so that after he has become thoroughly familiar with all the operations of machine molding he is advanced from helper to that of machine molder.

His experience is further increased by the variety of patterns from which he is able to pass on to the next grade of skill as finisher and core-setter. Having become familiar with many of the operations at the machine molding stations, he soon develops into a skilled finisher and core-setter.

We have only mentioned two specific examples of how it is possible to train rapidly new people from unskilled through semi-skilled to skilled workmen. It is dependent on the constant instruction and ability of the foreman in charge.

This method has the advantage of producing a skilled worker in a phase of operations much sooner than through the regular apprentice course. We recognize the pronounced ability of some people and are able to advance them through the various stations at a much higher rate than others. In the regular apprentice training program a definite time must elapse before the man has completed his training. It is, however, true that in this case we are interested in a particular development of a person for a certain skilled job; whereas in the regular Apprentice Training, which requires four years in our shop, we are interested in making an all-around foundryman from the trainee.

In any training program the foreman is the key man. If the foreman is interested only in having a man fill a particular job at the moment, you can not expect that man to advance. However, if the foreman is interested in improving his personnel and thus his department, he will do a good job in both training systems and neither will result in a better finished product than the foreman who has been instrumental in instructing his personnel.

SUGGESTED CUPOLA REFRACTORY STANDARDIZATION

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Chicago Retort & Fire Brick Co.
Chicago, Ill.

A GENERAL CONFERENCE of producers and users of cupola refractories was called in May, 1932, in conjunction with the U. S. Department of Commerce. At this conference a simplified practice recommendation covering sizes of cupola block, tap-hole and slag-hole blocks, was drafted. The original acceptance of this recommendation in 1934 was later modified in 1938. This recommendation is identified as R-154-38. The majority of those concerned, users and producers of cupola refractories, subscribed to these recommendations.

Standard cupola block sizes for 6, 9 and 4½-in. thicknesses were tabulated. Five sizes of two-hole tap-out blocks, 9 x 4½ x 2½ in. with holes ranging from ½ to 1½-in. diameter in ¼ in. increments, were accepted as standard. Similar five sizes of one-hole tap-out blocks, also 9 x 4½ x 2½ in. with single holes of same size and increment, were also accepted. Two sizes only of slag-hole blocks, 9 x 9 x 4½-in. thick similar to the "F" calliau block were adopted having a 1½ or 2-in. hole centered.

It is the author's belief that with improved methods and control of charging, the foundryman will soon realize that control of the melting rate in terms of pig, scrap and coke charged per unit time is preferable to variable stop-offs or "bottings" of the cupola breast. Front slagging, continuous pouring cupola operation follows in this direction.

Presented at a Refractories Session of the Fiftieth Annual Meeting, American Foundrymen's Association, at Cleveland, May 7, 1946.

Use of odd shapes and sizes of cupola refractory blocks is uneconomical from the procurement standpoint. Also, difficulties in cupola relining can be expected if improper sizes of cupola block are used. Standardization of sizes and shapes of cupola refractories is a step in the right direction.

The minimum of molten iron held in the wall results in more uniform carbon pick-up, hence better control. The cupola's counterpart, the blast-furnace, has its cast rate controlled on the charging end and not at the notch or spout.

The majority of cupolas today operate with 9 x 6 x 4-in. block, each course of which is 4 in. high. How can 9-in. high tap-hole brick be set in 8 or 12-in. high breast spacing without cutting brick or plugging with clay? When refractories do not fit exactly, it is natural for any one to use his original ideas.

However, individuality in shapes and sizes of refractory blocks is uneconomical. Both producer and user should come to an understanding for mutual advantage. This was the guiding thought of the plans for simplified practices in the early thirties set forth by the Department of Commerce.

In relining the cupola to a certain inner diameter it should be obvious that proper size block or a combination of sizes should be used for closing this circle correctly. Difficulties in relining can be expected if improper sizes of cupola blocks are used in the relining operation.

When block of a larger size than required are used the joints will be open toward the back in the form of a "V" whereas block of too small a size will line up with open joints on the front. Such conditions will tend to decrease the productive life of the cupola lining.

Since 1937, the American Refractories Institute adopted as standard the revised series of cupola block, which in the calliau block of 4½-in. thick sizes retained the same alphabetical designations as before.

However, when procuring A, B, D or E block at the present time, the foundryman receives them 2 in. smaller in diameter than the sizes shown in the sales catalog tables of standard cupola manufacturers. For some years these equipment suppliers have shown lining tables which are not correct on sizes smaller than the No. 5, having a shell diameter of 56 in. or less.

Practically no refractory producer today makes these old sizes unless as hand made shapes, and few are carried in stock. The best cupola block today are machine-made stiff-mud processed under 25 in. or more of vacuum.

The foundryman purchases his cupola and from the tables of the manufacturer he places his order for the block and brick required; then he finds these block close a circle two inches less inside diameter, giving a resulting decrease in cross-sectional area of the melting zone. It also means there is one inch additional back-up to provide all around.

The author feels that every cupola lining, regardless of the size, should have a backup course and this allow-

ance should be made by the manufacturer in determining his shell size.

We have found through experience that sufficient allowance for expansion must be provided in the backup brick or block used in the melting zone area. The use of a cardboard spacer, ⅛ to ¼ in. thick in the joint, at least four times in each course of backup, will prevent undue expansion pressure and resulting distortion of a new cupola shell.

If your iron melting is too fast to be handled and the floors poured off, the best solution is to line down the inside diameter to such a point that the iron can be moved away about as fast as it melts. The normal melting rate suggested in current literature is 10 to 11 lb. of iron per sq. in. of melting zone cross-sectional area per hour. When possible a longer burning-in time between the "light-off" and "wind time" is preferable as it results in better preheating of the lining, and hence longer life.

Too many foundries in the past have slighted the cupola and its attendant equipment, except as the source of molten iron. Improved storage facilities and location for patching brick, block stone, fire clay and cupola block will pay dividends.

Next to pig iron, scrap and coke, air is the largest ingredient in cupola melting practice. It is too often forgotten or minimized. About a pound, or 13 cu. ft., of air is used for every lb. of iron melted.

Many shops are now equipped with constant volume, balanced blast controls, with the latest equipment also controlling the moisture content, another important variable. Blast volume or quantity of air is the important thing; pressures are secondary in importance.

However, a cupola which is being pushed by stepping up the air blast pressure is going to be one in which refractory life is considerably shortened; it would be cheaper to alter the cupola tuyeres.

DISCUSSION

Chairman: C. E. BALES, Ironton Fire Brick Co., Ironton, Ohio.

Co-Chairman: C. S. REED, Chicago Retort & Fire Brick Co., Chicago.

CHAIRMAN BALES: The subject of cupola refractory standardization is of vital importance to the producer and to the user of foundry refractories. If we can have certain standard sizes and shapes, the price is naturally more favorable. The greatest advantage is the fact that the manufacturer will usually have standard size materials in stock, whereas if you use non-standard refractory articles it may take anywhere from four to eight weeks to make them.

J. A. BOWERS[1]: What is the reason for the backup brick back of the cupola block?

MR. STEVENS: Too often it is the line of least resistance in cupola operation to run the block until they burn through. If you like buying new sections of a lower shell, that is one problem. If you want to have your cupola constant in diameter and protected you should use some backup, and then reline when you burn through your inner lining.

The smaller size cupolas, according to the manufacturers' standard tables do not provide for backup. Too often we go into particularly the smaller shops and find many welded spots on their cupola shells. They simply have not provided for that additional margin of safety.

ARNOLD SATZ[2]: How do you prevent burning in above the tuyeres when the blast volume is increased?

MR. STEVENS: Unfortunately, I am not much of a cupola operator. I know there are several types of tuyeres used. It is possible to put more tuyeres facing on a cupola that requires more volume of air. The easiest thing to do is to increase the rpm., and increase your pressure. That is just the wrong thing to do. You are upsetting your equilibrium conditions. You will pay for it through increased refractory expense.

CHAIRMAN BALES: The thing that has often puzzled me is why it is, with cupola blocks rather well standardized, the fabricator of cupola shells is so non-standard as far as the refractory blocks are concerned. In the cupola manufacturer's catalogue you will find a list of non-standard blocks as being necessary to line up a cupola the way they recommend it.

MR. STEVENS: It is foolish to continue to supply cupolas when you go to your refractory supplier and procure blocks which do not fit. That is just exactly what nine out of ten foundrymen are doing, whether you are doing it knowingly or unwittingly. As Mr. Bales says, these cupola catalogues are listing sizes which have been obsolete for 15 years.

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[1]American Cast Iron Pipe Co., Birmingham, Ala.

[2]National Radiator Co., Newcastle, Pa.

MALLEABLE IRON FOUNDRY CORE PRACTICE

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Union Malleable Iron Works
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PRODUCTION OF CORES for any foundry, and for malleable foundries in particular, is of great importance. The quality and characteristics of dry sand cores definitely influence the quality of the castings produced.

The purpose of this article is to outline briefly the production of dry sand cores in a malleable foundry manufacturing agricultural castings and chain links. Castings produced regularly in this foundry vary in weight from approximately one oz. to 25 lb., with an average weight per casting of about ¾-lb.

Raw Materials

Practically all cores in this foundry are made from one of four different sands, or a combination of two of these sands. The sand most commonly used is a locally obtained bank sand containing a small amount of clay. The relatively low cost of this particular sand is one of its chief advantages. The other three sands commonly used are washed and dried Ottawa silica sands. The physical characteristics of these four sands are given in Table 1.

Core oil used in this plant is a petroleum polymer type oil having the following chemical specifications:

Saponification No.	80-85
Iodine No.	195 (min.)
Acid No.	14-17
Specific Gravity	0.940

This core oil has been found quite satisfactory for most work when all of its characteristics are considered. However, a faster baking oil having similar chemical specifications is used for some applications.

Commonly known, commercially prepared cereal binders are used to obtain green strength in most of the core sand mixtures. In a few instances an Albany type molding sand is added in small amounts to produce definite properties. Due to the excessive absorption of oil by the clay in molding sand, this practice is not advisable unless necessary.

Mixing is done in a muller-type mixer. The usual amount of sand mixed is 34 gal. per batch. The dry ingredients, sand and cereal binders, are mixed together for approximately one min. before the oil and water are added. The total mixing time varies with the different mixes (from 4 to 6 min.) and is measured with an interval timer. The moisture content of each batch of sand is carefully determined with a moisture tester. Definite moisture ranges have been set for each mix, and it

Table 1

SIEVE ANALYSES OF SANDS USED

U. S. Bureau Standards Sieve No.	*Retained, per cent*			
	Colona Bank Sand	*Ottawa Sand "A"*	*Ottawa Sand "B"*	*Ottawa Sand "C"*
6	0.0	0.0	0.0	0.0
12	0.0	0.0	0.0	0.0
20	0.0	0.0	0.0	0.0
30	0.0	0.0	0.0	0.0
40	0.7	0.3	0.5	0.2
50	6.0	45.6	24.5	3.6
70	18.8	36.1	40.2	21.0
100	35.2	16.4	24.8	40.2
140	21.7	1.5	7.0	19.1
200	8.8	0.1	2.0	9.1
270	3.5	0.0	1.0	3.6
Pan	2.7	0.1	1.0	3.2
A.F.A. Clay	2.5	None	None	None
A.F.A. Fineness No.	88	51	59	89

Core quality control is an essential factor of quality casting production. Close control of the core-making process—raw materials, mixtures, core baking and finishing—aid the foundryman to meet increasingly rigid castings specifications.

This paper was presented and discussed at a Malleable Iron Founding Session of the Fiftieth Annual Meeting, American Foundrymen's Association, at Cleveland, May 6-10, 1946.

Table 2
CORE SAND MIXTURES

Components	Sand Mixtures "B"	"Bracket"	"Sickle"	"Chain"
Bank Sand, gal.	24	14		
Ottawa Sand "A," gal.	10	20		
Ottawa Sand "B," gal.			34	
Ottawa Sand "C," gal.				34
Cereal Binder, gal.	1	1	1	1
Core Oil, pt.	5	6	6	10*
Kerosene, pt.	1	4	4	4
Water, pt.	5	9	9	10
Mulling Time, min.	4	4	6	6

*Fast drying.

is very important that these be maintained.

Four of the most commonly used mixtures, together with their physical properties, are listed in Tables 2 and 3.

At least 90 per cent of the production cores are made from these mixtures. All four are used either on core blowers or on the bench. The "B" sand mixture is used for approximately 50 per cent of the cores. It produces a core of average hardness and sufficient strength to meet most needs. The "bracket sand" mixture is used to produce cores that require more strength and higher hardness than usual. Cores for mower sickle guards and other applications requiring close tolerances are made from the "sickle sand" mixture. The "chain sand" mixture listed is used on the core blowers to produce cores for a number of different types of malleable chain. A "chain sand" mixture previously used in the plant would not blow satisfactorily, and it was found that by changing the mixture to the one listed, the core blowers could be used. This resulted in a saving in labor for these cores of approximately $10 per ton of core sand used, and greatly increased the quality of the core produced.

Variations Made

However, as in the case of most foundries, variations of these mixtures are made to take care of certain peculiar requirements. The method of making the core, as well as the desired characteristics of the finished core, has a direct influence upon the core mixture.

The use of kerosene in almost all mixtures is a common practice in the plant. It is highly important that there be little sticking in the core boxes. Other materials for prevention of sticking have been tried but, when cost and results obtained are considered, kerosene has proved quite satisfactory. The amount of kerosene used does not greatly affect the dry strength of the cores.

Coremaking

Cores are produced by several different methods. About 75 per cent of the cores are made on core blowers. Because of the nature of the work, most of the cores can easily be made by this method. Sixteen core blowers are operated at 95 psi. air pressure.

Several other coremaking machines of the vibrator type are also in operation. Jolt roll-over machines are used for the production of the largest cores. A small percentage of the production cores are handmade by bench coremakers.

Wherever necessary, aluminum core driers are used, although most of the cores will stand up on plates during baking. The chain cores are not placed on driers, but stand on end, thus producing a rather dense mass on the core plate. For this reason, chain cores receive a double bake by making two trips through the continuous oven.

In a number of instances metal chills are placed in the core box before the cores are formed, and thus become a part of the core. Metal inserts of various types are also used in this manner when the castings produced are to be used as hard iron and in an unannealed condition.

Baking is done, for the most part, in a continuous horizontal-type gas-fired oven. The temperature and time cycle are controlled automatically. The total cycle time is approximately 3½ hours, of which about 1½ hours are spent in the baking zone at a temperature range of 410 to 420° F. The conveyor carrying the cores into the oven passes between two rows of core blowers for convenient loading.

A group of gas-fired, batch-type core ovens is also maintained to take care of the largest cores, and special purposes.

Cores Cooled

As the cores come from the continuous oven they are somewhat above room temperature, and are further cooled on racks before packing. The cores then are inspected and packed onto boards for delivery to the foundry.

Chain cores, cylindrical in shape and varying from ⅛ to ½-in. in diameter and one in. to 3 in. in length, are packed onto boards in an end-to-end manner. The number

Table 3
PROPERTIES OF CORE SAND MIXTURES

Properties	Sand Mixture "B"	"Bracket"	"Sickle"	"Chain"
Moisture, per cent	2.9	2.5	1.5	1.5
Green Permeability	90	115	140	34
Dry Permeability	115	150	155	37
Green Comp. Strength, psi.	0.98	0.66	0.14	0.30
Tensile Strength (after baking at 400° F.), psi.				
1 hr.	183	214	263	360
1½ hr.	194	220	255	360
2 hr.	181	200	255	347
(after baking at 450° F.), psi.				
1 hr.	165	185	229	287
1½ hr.	164	172	213	267
2 hr.	142	151	209	273
Collapsibility (20-lb. load at 2500° F.), sec.				
(baked 2 hr. at 400° F.)	198	155	95	268
(baked 2 hr. at 450° F.)	180	125	127	283

of cores in a row depends upon the number of castings produced in a mold. The cores are set with the aid of a scoop that holds one row of cores. This practice greatly facilitates the setting of cores by the molder.

Because of the type of work produced in the foundry, few cores are sprayed or coated. For the small amount of core pasting that is done, a commercially prepared core paste is used. Where exceptional hardness is required in a particular section of a core, that section of the baked core is dipped in a mixture of core oil and kerosene (one part oil and two parts kerosene) and dried in the continuous oven.

Core Quality Control

As the quality of the castings improves, the control of core quality becomes more essential. The plant endeavors to control the quality of all raw materials used. Sand shipments are checked periodically for fineness and clay content. Each shipment of core oil is sampled and tested for chemical specifications and baking characteristics.

During the mixing of the sand, the moisture content is carefully controlled and mixing time is checked. Needless to say, excessive moisture or mulling time will cause considerable trouble with core blowers. Control of temperature in baking ovens is relatively simple with the controllers and pyrometers available to the industry.

Inspection of the finished cores is done by the packers as the cores are prepared for delivery to the foundry.

This article is not intended to serve as a guide for the successful operation of a core room, but is merely a brief outline of the methods used in one plant. There are many instances in the author's plant where improvements in core practice could be made, and as time and experience indicate these changes are made.

As this is being written the scarcity of cereal binders for core sand mixtures is becoming more and more important. It will soon be necessary to eliminate much of the corn product binders from the mixtures, at least temporarily. An experimental program is now under way, intended to find a substitute that will meet requirements.

DISCUSSION

Chairman: W. D. McMILLAN, International Harvester Co., McCormick Works, Chicago.

Co-Chairman: D. F. SAWTELLE, Malleable Iron Fittings Co., Branford, Conn.

CO-CHAIRMAN SAWTELLE: I do not remember your proportion of cereal binder to oil in the core mixes you use most.

MR. WELANDER: When corn cereal was commercially obtainable we used 1 gal. of cereal binder and 5 pt. of oil. That is almost twice as much binder as oil. The strength was developed by the cereal binder rather than the oil.

MR. SAWTELLE: As the proportion of the oil increases do you decrease the proportion of the cereal binder?

MR. WELANDER: No. In the smallest core we make the amount of oil would be increased and the amount of cereal binder would remain the same. The smaller cores do have a high oil content in proportion to the cereal binder.

MR. SAWTELLE: That agrees quite closely with my foundry.

H. E. HENDERSON[1]: As to moisture control in your core blowing operations, what moisture range do you find most satisfactory?

MR. WELANDER: I mentioned one mixture with 2.75 to 3 per cent moisture. That is the highest range we have. The others run lower than that. As the amount of Ottawa silica sand increases the moisture content is decreased to about 1½ per cent. Keep your moisture content below 2½ per cent, from 1½ to 2 per cent. Best results are obtainable in that range.

CHAIRMAN McMILLAN: Is your sand for cores dried to 1½ to 2½ per cent moisture?

MR. WELANDER: No, that is the core mix.

CHAIRMAN McMILLAN: Do you dry the sand?

MR. WELANDER: The Ottawa sand is purchased in the washed and dried condition. The bank sand that we obtain is not dried; however, it has a relatively small amount of moisture and we can easily hold the core mixes down to 2½ per cent moisture. We have to add water to everything we mix.

CHAIRMAN McMILLAN: Do you buy core oil by the pound or by the gallon?

MR. WELANDER: By the gallon.

CHAIRMAN McMILLAN: What do you pay for a gallon?

MR. WELANDER: At the present time we pay approximately 65 cents per gallon.

A. M. FULTON[2]: I notice you use chills with your cores. What do you put on your chills to cause them to bond well with the core, and what do you put on the outside of the chill to prevent burning in?

MR. WELANDER: Most of these chills are made in such a manner that their shape holds them well to the core. No coating is put on the core after baking to prevent sticking to the metal.

MR. FULTON: I understand you use 6 pt. of oil and 4 pt. of kerosene in your core mixture. Have you tried reducing the amount of core oil to make the core a little weaker and to lessen the amount of kerosene needed?

MR. WELANDER: This is an abnormal amount of kerosene, but we would rather have the hardness and the resistance to breakage of the core than to reduce the quantity of oil and make a weaker core. If we were using a higher grade, more expensive oil, we would probably use less oil. However, by using the lower cost oil, we feel we can afford to put in that much kerosene.

DAVID TAMOR[3]: Regarding coating of chills, we follow the same practice that we do in the making of swivels; that is, the outside of the chill is dipped in shellac and then in sand. That sand is about 70-mesh grain size, and our experience has been that the casting just sheds the chill when the core is shaken out.

MR. HENDERSON: I was very much amazed by the oil ratios of 1:200 and 1:400 that were mentioned at another session. Could you give us some information about those mixtures?

MR. WELANDER: Those particular mixtures were entirely oil mixtures. No dry binder of any kind was used. The only ingredients used were sand, oil, and water. The speaker stated that they were used on very light sections, that they were extremely soft, and that they had to be handled very carefully.

In our plant we could not do that. We require a core that will stand a great deal of handling so we use considerably larger amounts of oil. Those were light-section cores, used on comparatively light section work, where collapsibility was extremely important.

We checked the collapsibility by means of a dilatometer on the mixes I mentioned, particularly the B sand mix, which has about ⅔ bank sand and ⅓ Ottawa sand. In the collapsibility test we made a specimen 1⅛-in. in diameter and 2-in. in length. It is similar to the regular permeability specimen except smaller in diameter.

We baked the test specimen at 400° F. or under conditions which simulated those in the core oven. Then we applied a constant load of 20 lb. After the load had been applied, we lowered the dilatometer furnace over the specimen. In this particular case we set the temperature control at 2500° F. so we exposed that core to a temperature of 2500° F. immediately. We found that collapsibility occurred in about 198 sec. or a little over 3 min. As the amount of Ottawa silica sand in the mix was increased, we found that the specimen collapsed in less time.

[1]H. C. Macaulay Foundry Co., Berkeley, Cal.
[2]Northern Malleable Iron Co., St. Paul, Minn.
[3]American Chain & Cable Co., York, Pa.

CENTRIFUGAL CASTING of a MAGNESIUM PART

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and
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Dow Chemical Co.
Midland, Mich.

A RING CASTING of approximately 13 in. outside diameter, 11 in. inside diameter and ½-in. thickness was chosen for experimental work on centrifugal magnesium casting because it had been made as a commercial permanent mold casting and because of its simplicity of design. It should be realized that a relatively small percentage of commercial shapes can be cast by the true centrifugal process.

Two factors probably have tended to retard the development of a process for casting magnesium alloys centrifugally. The first of these is the chemical reactivity of molten magnesium, or its tendency to oxidize readily in contact with the air. The second factor largely concerns its density. The density of magnesium oxide is greater than the density of magnesium. This factor makes it difficult to separate the dross from the metal by centrifugal force because the heavier MgO particles may be trapped by the freezing alloy.

Another point to mention here in comparing centrifugal casting of magnesium with steel or cast iron is the fact that, in steel or iron alloys, the slag is not only much lighter but is also a liquid at casting temperatures, while the magnesium dross is a solid. This combination of a low density, fluid slag makes it relatively easy to separate the slag from the metal centrifugally in steel practice.

It is interesting to note at this point that these same two factors were a problem in the early days of the sand casting of magnesium alloys. The volume and quality of magnesium sand castings today indicates that these factors were overcome in commercial practice. The balance of this discussion indicates that they can also be overcome in centrifugal casting practice.

Description of Equipment. The centrifugal casting work described here was done on the machine shown in Fig. 1. This view shows the steel mold as it is mounted on the machine. The casting is formed under the cover in an impression cut into the mold. The cover plate also fits into an impression cut into the mold and is held down during the spinning by the three "dogs" shown in the picture.

The machine is mounted so that it can be rotated through 90° allowing for casting from any angle from horizontal to vertical. The machine is driven by a 3 hp. motor equipped with a dynamic drive which allows a continuously varying mold speed with a constant speed A.C. motor. An air brake is used to stop the mold. A tachometer indicates the mold speed.

A 3½-lb. capacity ladle of the conventional teapot spout design was used to pour the castings. The metal was melted in 100-lb. magnesium capacity open pots and heated directly to the casting temperatures used in all cases. The metal was not given any grain refining treatment. The alloy used was ASTM alloy AZ92 (9Al-2Zn-0.2Mn). This is the same alloy, and metal preparation treatment, as used on the static casting to which the centrifugal casting is being compared.

Discussion of Experimental Results. Centrifugal casting of AZ92 alloy requires that a definite technique be followed in order to produce good quality castings. Table 1 indicates the results of various casting temperatures, mold speeds and pouring methods. Pits, microporosity, lapping or cold shutting, and banding can occur under certain conditions.

Pitting is caused by excessive turbulence of the metal as it flows from the ladle into the mold. In normal sand or permanent mold casting the metal is bottom gated or side gated into the mold to eliminate turbulence. In centrifugal casting the metal is, in effect, top gated because it flows into the mold in the direction of the centrifugal force.

In order to reduce to a minimum

Centrifugal casting of cast iron, steel and the heavy non-ferrous alloys is an old practice. Relatively little work, especially in production, has been done on the centrifugal casting of light alloys. Investigation has shown that a commercial magnesium casting can be made successfully by the true centrifugal permanent mold process, the product being equal to or better in quality and efficiency than that obtained in static permanent mold casting.

This paper was presented at an Aluminum and Magnesium Session of the Fiftieth Annual Meeting, American Foundrymen's Association, at Cleveland, May 6, 1946.

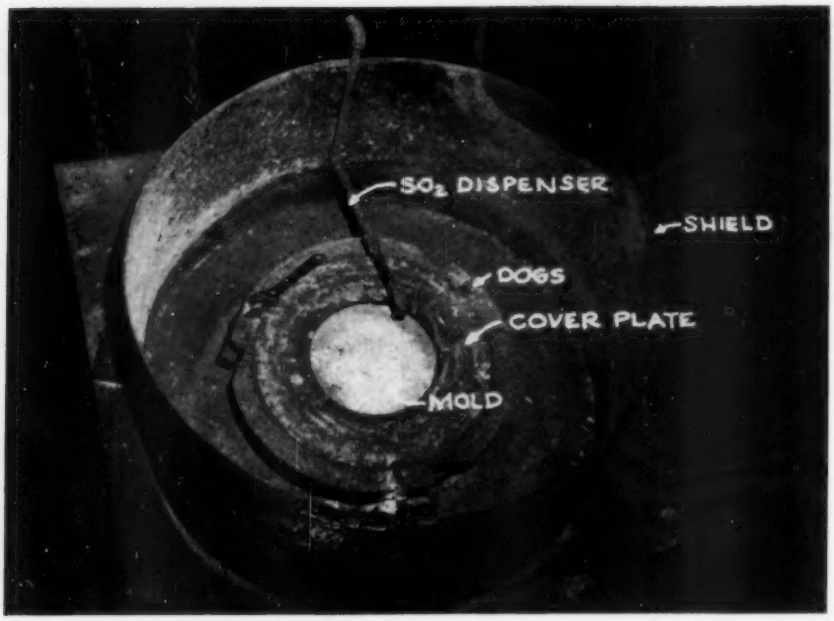


Fig. 1—Front view of centrifugal casting machine.

Table 1

EFFECT OF CASTING CONDITIONS UPON QUALITY OF CENTRIFUGALLY CAST RINGS

Mold Speed, rpm.	*Mold Temp., ° F.*	*Metal Temp., ° F.*	*Type of Pour**	*Quality Rating***	*Defects*	*Physical Properties as Heat Treated and Aged* — *Tensile Strength, 1000 psi.*	*Yield Strength, 1000 psi.*	*Elongation, per cent in 2 in.*	*Grain Size, 0.001 in.*
600	600	1375	P	E	Porosity, Pits, Lapping				
800	600	1240	P	C	Lapping				
		1300	P	C	Lapping, Pits	37.3	26.2		5
		1375	P	C	Porosity, Slight Lapping	39.3	28.7	1.4	5
1000	600	1240	P	B	Slight Lapping and Pits	36.5	28.8	1.0	3
		1300	P	A	None	36.9	25.3	1.8	4.5
			O	D	Pits				
		1375	P	A	None	37.0	24.9	1.4	4.5
			O	D	Pits				
1200	600	1240	P	B	Banding, Lapping	37.0	28.6	1.2	3
		1300	P	D	Banding	38.9	30.3	1.2	3

NOTE: Fast pour used for castings in above table to avoid lapping. Also, a 30° mold angle was used for ladling convenience.

*P—Peripheral pour. O—Center pour.

**Key to Quality Rating: A—Excellent, acceptable for production. B—Good. C—Fair. D—Poor. E—Bad.

Table 2

COMPARISON OF PROPERTIES OF CENTRIFUGALLY CAST AND STATICALLY CAST PERMANENT MOLD RING

Casting Method	*Temp., ° F.*	*Tensile Strength, 1000 psi.*	*Yield Strength, 1000 psi.*	*Elongation, per cent in 2 in.*	*Grain Size, 0.001 in.*	*No. of Values*
Static	1240	31.1	21.6	1.3	9	6
	1300	29.9	21.4	1.0	9	5
	1375	31.3	20.8	1.2	11	5
Centrifugal*	1240	36.5	28.8	1.0	3	4
	1300	36.9	25.3	1.8	4.5	10
	1375	37.0	24.9	1.4	4.5	10

*Only "A" and "B" quality rings considered. All rings spun at 1000 rpm. with peripheral pour.

the "drop" of the metal from the ladle spout to the outside diameter of the mold, it was found that the opening in the cover plate should be as large as possible, leaving only enough room for the casting and the riser. The riser, of course, in true centrifugal casting consists merely of excess metal on the inside diameter of the casting. The spout of the ladle should be placed as close to the inside diameter of the cover plate as possible.

Centrifugal Pour Types

In Table 1 this type of pouring is classed as a "peripheral" pour as compared with a "center" pour. Figure 2 shows diagrammatically the manner in which the metal is scattered with various types of pour. For minimum turbulence the metal should not only enter the mold near the periphery, but the mold should also be spinning in the proper direction consistent with the type of ladle used. It is evident that the method shown in Fig. 2C produces the least amount of pitting.

Another common method for reducing turbulence in centrifugal casting is to start pouring the metal at a slow rotational speed of the mold and to accelerate the mold as the pouring progresses. This method was tried on this casting, but no obvious improvement in quality resulted from the slower initial speeds.

However, the relatively small volume of metal poured and the rather slow acceleration of the mold may have been responsible for the lack of improvement, since the metal probably had frozen before the mold reached its top speed.

Cleaning the Metal

Oxide particles and skins are other defects which are caused by excessive turbulence, especially at the higher pouring temperatures. The same precautions are necessary in centrifugal casting as in sand or permanent mold casting with respect to cleaning the metal in the pot and protection of the metal in the pot and ladle.

Again, the peripheral pour with the mold spinning in the proper direction relative to the ladle design produces the least turbulence and hence the least amount of burning. The mold should be flushed with SO_2 gas prior to each casting.

Lapping or cold shutting is the

term applied to the defect which may occur on the outer diameter of the rings. It is caused by turbulence and low metal temperatures together with low mold speeds. The reason for it is that small globules of metal break away from the main stream of metal and are thrown against the outside of the cavity.

These globules then freeze against the mold surface and do not entirely weld into the main body of the casting. Slow or discontinuous pours also cause lapping because the metal at any point in the casting freezes before the next layer of metal is added to it.

Microporosity is caused by lack of sufficient feeding of the metal from the riser to the casting. Under the conditions of the test, there was a greater tendency of microporosity at the higher casting temperatures. It was found that the microporosity was eliminated at these higher temperatures by going to higher mold speeds, which produce more efficient feeding from the riser.

Banding is caused by vibration of the mold. This is a function of machine design and support. The mold used in this work vibrated noticeably at 1200 rpm. and caused a considerable amount of banding. The banding consisted of concentric rings

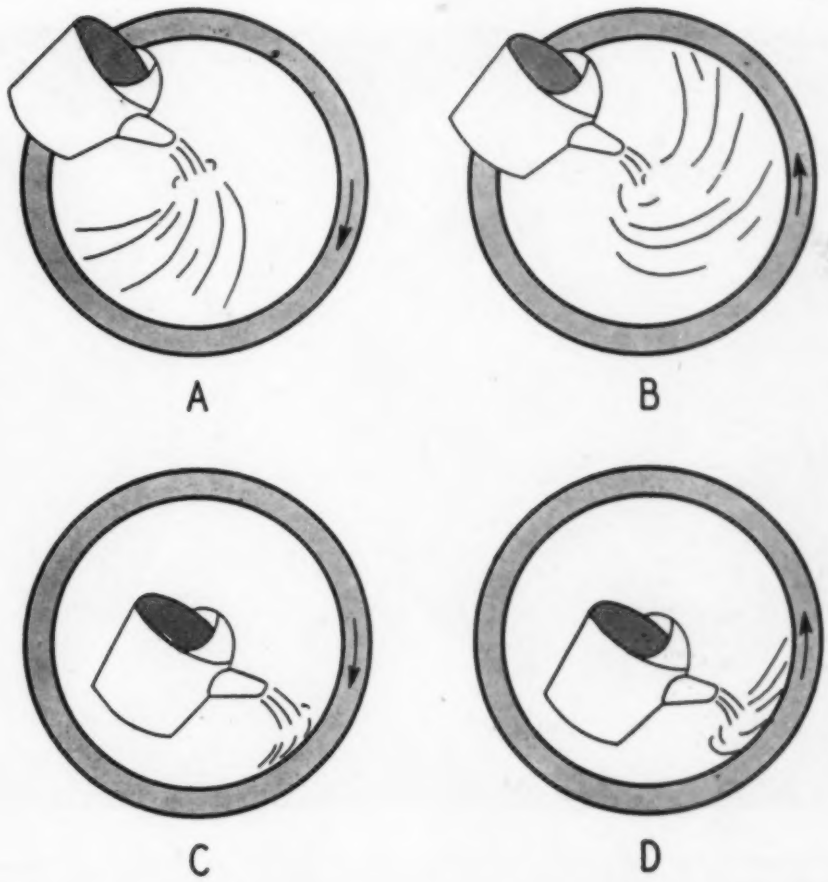


Fig. 2—Sketch showing various methods of pouring. A—Clockwise rotation, center pour. B—Counterclockwise rotation, center pour. C—Clockwise rotation, peripheral pour (least turbulence and exposure of metal to atmosphere). D—Counterclockwise rotation, peripheral pour.

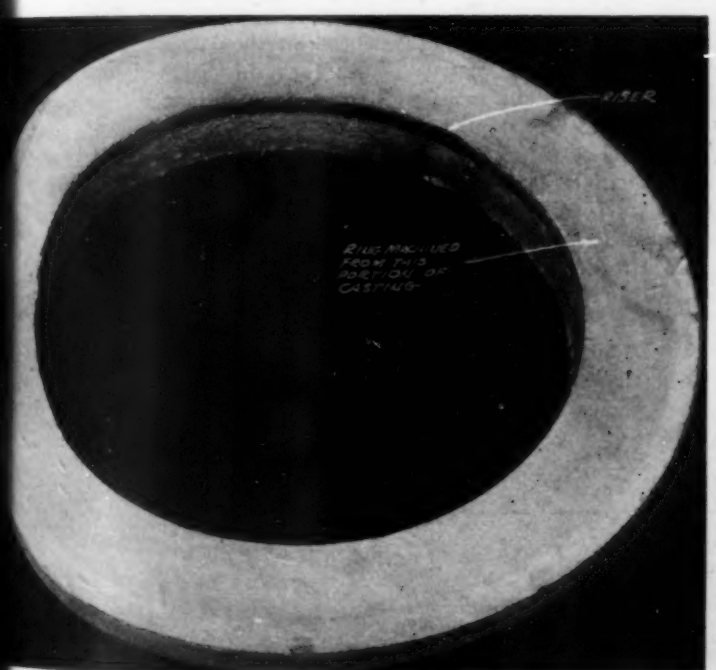


Fig. 3 (above)—Photograph of centrifugally cast ring.

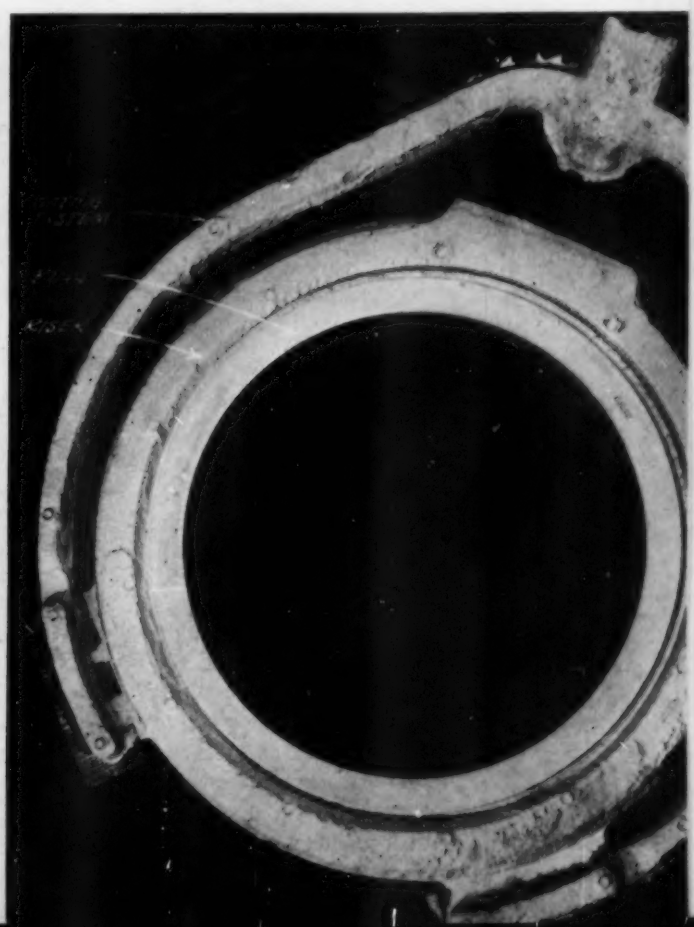


Fig. 4 (right)—Photograph of statically cast ring.

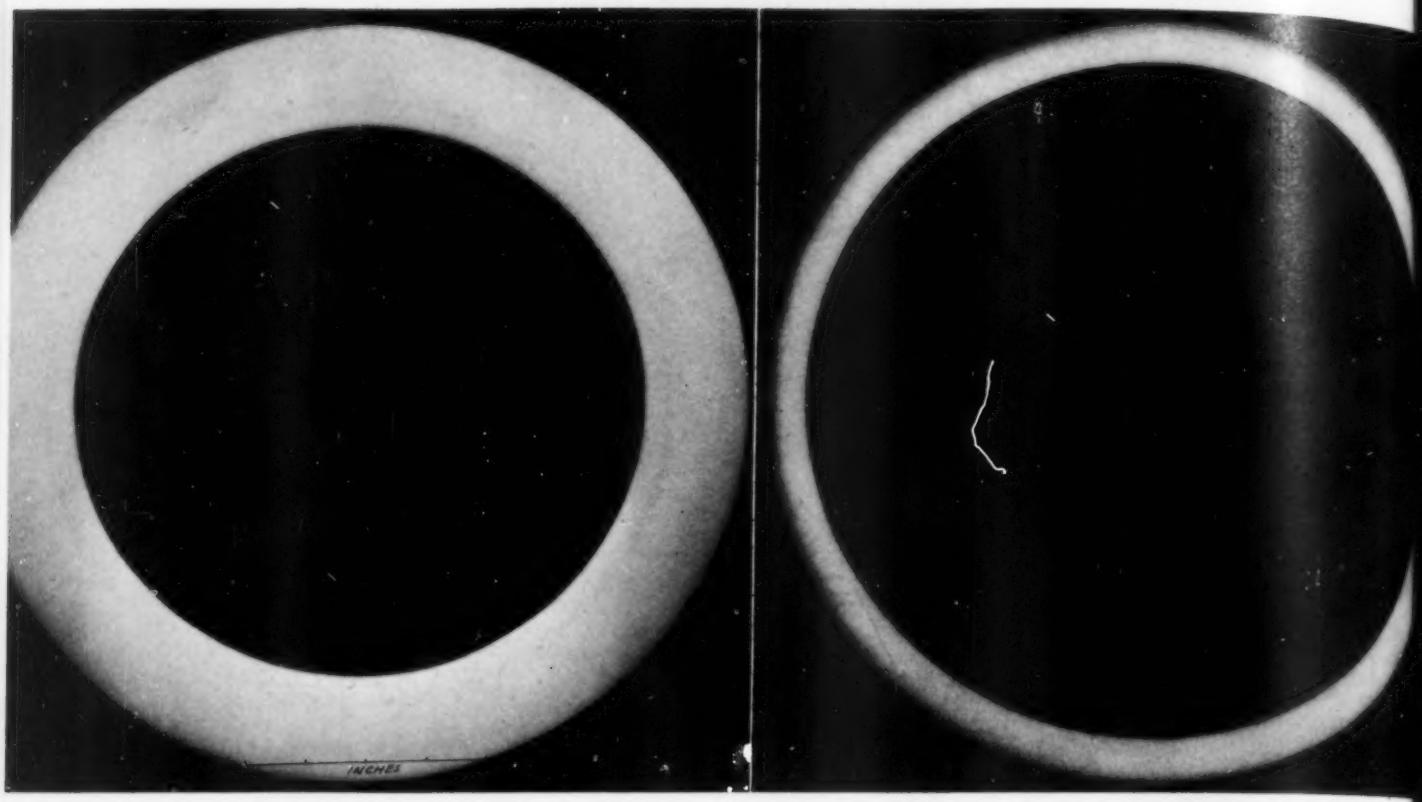


Fig. 5 (left)—Radiograph of centrifugally cast ring. The casting appears perfectly sound. Fig. 6 (right)—Radiograph of statically cast ring. Some porosity is apparent.

of porosity running partially or completely around the casting.

Considering these defects, it may be seen that the best castings can be produced by maintaining a balance of the casting conditions as follows:

1. Mold speed about 1000 rpm.
2. Metal temperature about 1300-1400° F.
3. Mold temperature about 600° F. The mold, under continuous operating conditions without supplying additional heat, operates at this temperature.
4. Peripheral pour with the direction of mold rotation such as to produce a minimum of turbulence (Fig. 2C).
5. Fast pour.
6. Mold angle about 30° from the vertical for ease and convenience in getting the ladle spout close to the mold.
7. Care in holding and ladling metal from the pot to obtain clean, flux-free metal, as is required in all casting methods.

Under these conditions, all of the defects considered are reduced to a minimum and castings of satisfactory quality are consistently obtained.

Comparison of Static and Centrifugal Casting. The same casting was produced statically in a permanent mold for comparison. A comparison of the quality and economy of the two methods of making the same casting indicates that the centrifugal casting can be made at least as inexpensively as the permanent mold casting, and it is of a better quality.

The pouring temperature of the static casting is about 1200-1250° F. in production. However, some castings were made at 1300-1400° F. in the static mold in order to compare the two methods at equal temperature.

Figures 3 and 4 are photographs of the centrifugally cast ring and the statically cast ring, respectively, showing the gates and risers in both cases. A saving in the amount of metal cast centrifugally over the amount required statically is evidenced in these two pictures. The casting as shipped weighs 1.53 lb. The permanent mold casting weighs 6.6 lb. for a casting yield of 23 per cent, while the centrifugal casting weighs 2.8 lb. for a casting yield of 54 per cent.

Table 3

COMPARISON OF CENTRIFUGAL TO STATIC METHOD OF CASTING

	Centrifugal
Grain Size	Better
Tensile Properties	Better
Casting Yield	Better
Casting Speed	Equal
Radiographic Quality	Better
Cleanup	Equal
Fracture Cleanliness	Equal

Both types of castings were poured with a refractory wash on the molds. The thickness of the wash and the variation in thickness from riser to casting is very important in the permanent mold casting. On the centrifugal mold the wash is far less critical.

Typical prints of radiographs of the rings cast by the two methods are shown in Figs. 5 and 6. While the prints are not as clear as desired, some porosity can be seen in the static casting. The centrifugal casting appears perfectly sound.

Comparative Properties

A comparison of the heat treated and aged tensile properties and grain size of the two casting methods is shown in Table 2. This table indicates a considerable improvement in

the grain size as a result of the centrifugal casting process. The metal was not given any grain refining treatment, and yet the grain size is equivalent to that obtained statically only when the metal is intentionally grain refined.

The grain refinement probably is caused by the additional chilling effect obtained in centrifugal casting since the casting is held more tightly against the mold walls. The improvement in tensile properties is due to a combination of the improvement in soundness as well as the refinement of the grains.

About 24 to 30 static castings can be made per hr. by one man on one mold under the present operating conditions. One man can cast the same number of rings at the present time on the centrifugal machine. By designing the casting cavity into the cover plate, and by using two or more cover plates, one man probably can make considerably more castings per hr. by this method.

It is estimated that the cost of machining the centrifugal ring is about the same as for the static ring. The centrifugal casting requires a lathe cut to remove the inside material which acts as a riser, while the static casting requires a saw cut around the outside to remove the combination gate and riser.

A summation of the comparison between the permanent mold casting and the centrifugal casting indicates that the centrifugally cast ring is superior to the statically cast ring in quality, and is somewhat better than the static casting in economy of production. Table 3 indicates the comparison between the two methods. A large scale production run was not made with the centrifugal process because of contract cancellations following the end of the war.

DISCUSSION

Chairman: Leslie Brown, Magnesium Fabricators Div., Bohn Aluminum & Brass Corp., Adrian, Mich.

Co-Chairman: G. G. Glider, Ebaloy Foundries, Inc., Rockford, Ill.

R. F. Hauser[2]: Table 3 shows a better grain size with centrifugal castings as compared with the static casting. I would like to ask what method was used for grain size determination.

Mr. Strieter: Grain sizes in Table 2 were about 0.009 to 0.011 in. as the average grain diameter.

Mr. Hauser: Is that taken in one direction?

Mr. Strieter: The grain diameter is determined by a visual comparison of the grains at a magnification of 100 diameters with the grains of a standard grain size chart.

Mr. Hauser: Was that taken in the direction of flow of metal as well as transverse to the flow of metal?

Mr. Strieter: There was no difference. There is no orientation of the grains. They were equiaxed. The grain size was the same in all directions.

Dr. Basch[1]: The feature that interested me was that Mr. Strieter's conclusion practically completely coincided with what we of the AFA Committee on Centrifugal Casting of Light Metals so far have found in our work. Of course, one swallow does not make a summer and the experience with one casting may not necessarily indicate what we can get in general production.

Member: Has any other work been done on centrifuging, semi-centrifugal or precision casting by centrifugal force in magnesium?

Dr. Basch: We ourselves considered the centrifuge casting as probably having a greater commercial appeal than the straight centrifugal casting on account of the greater variety and flexibility of design and certain economic features. I must admit that we so far cannot contribute any accomplishments. We are in the midst of investigation and experiment. What we have done so far simply indicates that tremendous advantages can be obtained, but how we will obtain them is still in the offing. That still has to be determined and it is the objective of the committee's work to find out how it is to be done.

Member: I would like to add that there are two or three companies that are making some small magnesium parts by the centrifuge method.

Chairman Brown: The centrifugal casting of magnesium and aluminum alloys at the present time is quite young and there is considerably more work to be done before we are assured that we have a commercial process that can be used on various types of castings.

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[1]Almin Co., Schenectady, N. Y.
[2]Bendix Aviation Corp., Teterboro, N. J.

PATTERNS in a PRODUCTION FOUNDRY

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IT IS PROBABLE THAT the first pattern was made by an artist in an attempt to capture the beauty of a form or figure. In doing this the artist was thinking of only one casting. It did not occur to him to make the pattern so that it would mold easily. His aim was to reproduce all the irregularities of his model and, in fact, to add a few in order to demonstrate his consummate skill.

Cost Unimportant

Undoubtedly the pattern was a solid, one-piece affair made of clay. The molding element probably was a clay or plaster of sorts that dried or set over the pattern and was removed by breaking away—later to be reassembled and filled with metal. The artist got results, but the man hours involved were tremendous. The value of the completed casting was not measured in dollars, but in the pride and awe of a masterpiece created.

In this endeavor one person was at one and the same time creator, designer, patternmaker, molder, melter and finisher. His limits were beauty and proportion; absolute size and shape were not required. Today, those who cast metal for commercial use are faced with a much different problem, summing up somewhat in the following manner:

1. Exact requirements of metal specifications.
2. Exactness of size and shape.
3. Need of a competitive cost.
4. Delivery date to be met.

Item No. 1 deals in shrinkages of metal. All patternmakers have rules, but some have difficulty in using them properly. On small patterns the shrinkage is a minor consideration, but on large work it sometimes becomes involved. A long, thin casting will not shrink the same as one of the same length but several times thicker. Metal will not shrink the same over dry sand as it does over green sand, and this calls for a combination of shrink rules used on the same pattern. This shrink factor is the first step, and an important one.

Item No. 2 is a large order and one requiring the best thought in any organization. The patternmaker must be the guiding hand in any pattern equipment conference.

Most pattern purchasers send a print to the pattern shop with orders to make a pattern. A few purchasers specify the kind of pattern, and still fewer detail the kind of equipment and just how it is to be made. This type of purchaser usually is one of the larger companies having in its organization a planning department. Some of them do have the pattern planning "know how," but some fall far short. Therefore, it is the patternmaker's job to consider the pattern equipment to be made, and to suggest any change in part design that might help to produce a better or lower cost casting.

Suggestions Valuable

In some cases this is a difficult thing to do as some engineers feel that a suggestion of change is a reflection upon their ability and good judgment. However, the greater number know that the patternmaker's experience can be valuable.

Most patternmakers have had, at one time or another, the desire to extend a pad to a wall to eliminate a loose piece, or add a larger fillet

A patternmaker is an exalted craftsman, the greatest common denominator, as well as the least common multiple of all industrial production. A patternmaker must have the creative conception of a draughtsman designer, the practical ability of a molder, the precise skill of a machinist, the analytical judgment of a mathematician. He must create a plan, or design, with vision and ingenuity and build the idea from trade to trade with practical knowledge: thinking and forming inside and out with length, breadth and thickness—adjusting accurately all values and dimensions and producing with dextrous finality any conceivable form to be cast in metal. The products of the patternmaker's skill are truly surrounded by an aura of greatness which dignifies his right to assume a place of confidence, trust, and honor in all industrial advance and national progress.—Edward Leslie.

Presented at a Patternmaking Session of the Fiftieth Annual Meeting, American Foundrymen's Association, at Cleveland, May 9, 1946.

to reduce possible casting shrink or strain cracks. However, far too many patternmakers blindly make the pattern to the drawing, with the self-righteous, smug thought that the pattern is to the drawing, regardless of the fact that it is a poor pattern entailing a high molding cost.

It should be emphasized that a good engineer will thank the patternmaker for saving money or making a better, less costly casting possible.

Item No. 3 deals with competitive cost, not only in the pattern shop but in the foundry as well. The cost of the pattern in most cases, and certainly in production work, is a small item in the overall cost picture. Therefore, a pattern equipment which will mold readily and with the foundry equipment on hand must be designed if at all possible, a factor prone to be overlooked. As an illustration: A customer sends a drawing to several shops, asking for quotations on equipment to produce 200 castings.

Shop No. 1 quotes $25 for a single, one-piece pattern.

Shop No. 2, noting that the pattern can be split, quotes $35 for a two-wood pattern.

Shop No. 3, noting that 200 castings are wanted, checks the foundry for flask sizes on hand in the pattern range, and quotes $65 for a hardwood matchboard of one pattern to fit a certain flask and, upon inquiry, finds that the casting is to be a repeat number. Shop No. 3 also quotes $165 for an aluminum matchplate of two patterns, also to fit a flask on hand.

Overall Picture

In checking costs for the various equipment, the purchaser finds that the casting cost works out as follows:

Equipment No.	*Cost*	*Casting Cost*	*Total Cost (equipment + 200 castings)*
1	$25	$1.00	$225
2	35	0.92	219
3 (wood)	65	0.70	205
3 (metal)	165	0.60	285

Now, it will take no time to discard the quotations of shops No. 1 and No. 2. The only thing to decide is whether or not to spend $80 more on the first order to insure a low cost on all subsequent casting orders and have a permanent metal equipment, or take the lower first cost with a fair re-order casting cost. It is unnecessary to point out that shop No. 3, while quoting a higher pattern cost, got the job and made a good impression on the purchasing agent.

Item No. 4—delivery date. It will be found that in most cases a purchaser can and will extend the time limit on patterns in order to gain a good unit cost. This is particularly true when multiple pattern equipment will enable the foundry to produce at a much faster rate.

Customer Limitations

A word of caution must be added. It would be the height of folly to design an equipment, for example, to be used with jolt squeeze machines and core blowers when the foundry is a hand shop; or to make an equipment which will produce 1,000 cores a day when the production will not be able to use 100 a day. In other words, not only production but also the mechanization of the foundry using the pattern equipment must be considered.

It has been the author's experience that cores are expensive and should not be used unless made necessary by the design. However, if a drawback can be eliminated or a parting made easier, there should be no hesitation in using a core. The patternmaker should ask himself, "If I were the molder, would I like to use this pattern?"

Molders prefer a pattern with sufficient draft, no loose pieces, and if cores are used, prints having telltales so that cores cannot be misplaced. Also, clearance strips to prevent crushing should be provided.

The company with which the author is associated has a planning procedure that works fairly well. When a request for a quotation is received, it is referred to a planning board consisting of the pattern and molding superintendents who, after deciding which foundry department will make the job, call in the mold, core and cleaning foremen involved. This group then checks for present and future production and decides on the type of equipment. This information is then entered on an estimate sheet, going into detail as to just how equipment is to be made, giving flask size, number of core boxes, special arbors and core driers.

Each foreman then enters his respective cost. The pattern superintendent then reviews the equipment and enters time and material needed to produce the desired equipment, together with the cost. This card is then given to the cost department, which figures final casting cost and submits the quotation to the customer. If the order is obtained, the reverse side of the cost card becomes a job record card, so that both estimated and actual cost are on one card. This is desirable as it gives a ready check to the pattern foreman at all times.

The patternmaker assigned to the job is encouraged to offer criticism and, if the foreman believes that a change is desirable, the planning group is called to the pattern shop and again goes over the job. Good ideas are often obtained in this way. In planning the job, rough shapes are sometimes cut out to better show partings, etc.

High production patterns are mostly of the smaller types, that is, flask work, so in this discussion pit work can be passed over with only a word of caution. The pattern must be removed from the sand, and the patternmaker should provide a means of doing this without distorting either the pattern or the mold.

Cores

Medium size work—say 40 to 200 lbs.—is a casting field requiring considerable mechanization. The type and number of castings wanted largely dictate whether the pattern will be made for a stripping, a pin life, or a rollover machine. A prime point to keep in mind is that cores cost real money and can develop into a bottleneck, but if the part can be molded faster, then cores should be added (ram-up cores if possible, since a properly made ram-up core blends with the contour of the casting and shows hardly a joint, not to mention that they are easy to locate and hold in correct position). Then too, by adding a loose piece which can be removed when the mold is nearly rammed, and placing a core before completing the ramming, it often is possible to change a difficult three-part flask job into an easily molded cope and drag, with a good saving in equipment and in cost of casting.

When planning for a set core, it should be considered that the molder must set the core properly, without a lot of filing and fitting. It is false

economy to skimp on print material. A core must have its proper foundation, the same as a building. Also, the prints should have proper tell-tales. If two tell-tales can be used without extra coremaking cost, the patternmaker should not hesitate to use them. Also, the cope print cannot center a large core. If necessary, use a ram-up core ring in the green sand cope in which to enter the set core. Core prints in the drag should have beads on the edges to receive any shaved sand. In this way the core setting is speeded up.

When designing the pattern, a few extra dollars should be spent to provide fillets even if a few loose pieces are added to pattern and core box. It should be remembered that the molder's time is all production and that cutting fillets takes time.

When making a pattern for a production job using cores, it is well to remember that cores do not always run true to size, due to core box wear and sand and oil conditions. To overcome this, the patternmaker should provide metal core rubbing frames so that all cores must be fitted. This sounds costly, but on a production job the molding machine sets the pace and no time can be spent by the core setters to rub, fit and try outsize cores.

Another practice some patternmakers follow is that of placing a loose piece on a pattern with one or two dowell pins—an inexpensive operation in the pattern shop, but one that should not be tolerated by the foundry unless the pattern is for emergency use for a limited run.

Loose pieces should be firmly secured to a backing piece, that is, dovetailed into the pattern. If six loose pieces of different sizes are used, then six quite different sizes or shapes of backing pieces should also be used, in this way insuring proper location of parts.

However, if the six loose pieces are duplicates, then the backing pieces should all be identical so as to save the molder's time in finding proper location.

In jolting the mold and in straight ramming, loose pieces will sometimes lift unless a pin is added to hold them down. When the pattern is large enough to allow it, mechanical means should be provided.

DISCUSSION

Chairman: FRANK C. CECH, Cleveland Trade School, Cleveland.

Co-Chairman: V. J. SEDLON, Master Pattern Co., Cleveland.

MEMBER: Occasionally a customer will send in a large pattern which must be handled with a crane. However, no provisions are made for drawing that pattern out of the sand. There are no draw straps. Draw straps should be both on the cope and on the drag for handling to set the pattern in the mold, and also for removing the pattern from the mold. Too much of this expense is being needlessly borne by the foundry.

MR. PEALER: In large patterns we usually have three or four draw straps on the outside of the pattern and turning in on the bottom so as to give a good steady pull to the pattern; cross braces usually in the top, so that the means of drawing them out with a three-legged or four-legged chain will not crush the pattern.

On cope and drag work mounted on boards, we put inserts in the corners of the boards so that the foundry can, if they have the equipment, have a hook that fits in. Unfortunately, all foundries do not have that facility. Our foundry has a very elaborate way of handling patterns. They have a bail that fits over and grabs the patterns on the corners. If we would make a pattern for another foundry that did not have this bail, it would be useless. It is up to the foundry to state what kind of drawing equipment they have, whether they have a three-legged or four-legged chain.

I have found, also, that if you put one draw plate in the center of the pattern that can be bolted through to the bottom, it is a great help in the foundry to draw that pattern because of the necessity of drawing straight, and the one draw strap will do it.

If you are working in a shop that has a foundry connected with it, you will get that information directly from the foundry supervisory staff. If you are building the pattern for another plant and do not know where the pattern is to be used, you cannot very well say just what is wanted without contacting the foundry. If it is a small pattern, we very often leave the draw straps and draw plates off. If it is to be mounted on a machine, you do not need draw plates. If it is to be hand molded draw straps can be put in at very small cost at the foundry. We often contact the customer and get that information from him.

MEMBER: You mentioned that the patterns would have to compensate for the dry sand core because of the different shrinkage of metal. Is there a fixed amount?

MR. PEALER: No. I have never found that it was absolutely fixed. On very small cores we do not have to worry about it. But on a large core you make your core box to standard rule and the pattern to a standard shrink. That very often answers the question. I know all patternmakers do not agree on that. They all have their own ideas, but it is something that has to be thought of.

MEMBER: I would like to stress the cost of adapting equipment from the foundry facilities. We buy numerous patterns from outside sources. Consequently, in placing orders for patterns, we contact the foundry as to design to fit their facilities. There is a certain amount of work that has to be done on the patterns even after they are received by the casting men. However, that cost is reverted against the original purchase order for the pattern equipment. It is necessary for the casting vendor and the pattern shop vendor to get together and go over the designs.

MR. PEALER: That is a very good point. We have never found two foundrymen that use the same gating practice, for instance. We do not know why, but they all make good castings, and they all use different gates. When we ship a pattern from one plant to another, we have a charge for changing the gates, or fitting it to a different flask. If the purchaser would always tell the pattern shop where the pattern is going to be made, in most cases, the pattern shop would get in touch with that foundry to determine their requirements.

MEMBER: You suggest a rapping plate in the center or several rapping plates as a rapping device for the pattern. You use a 2x4-in. block to rap the pattern sufficiently to draw it out. Would you suggest that angle irons be placed there instead of the center rapping plates?

MR. PEALER: That is why I said the rapping bolts should be tightened to the bottom. There are on the market many facilities for rapping the patterns without damaging them. It is not a good idea for a molder to use a 10 or 15-lb. steel hammer on the pattern, or to put a hard wood rapping block against some delicate part of the pattern and to rap with it. There should be rapping plates. There are several fixtures on the market whereby you put a heavy rapping plate in the center of the pattern, and that is also the draw plate. There is a fixture that fits into it and locks in place. When you begin to draw the pattern, the rapping fixture vibrates and removes the pattern without excessive rapping.

E. T. KINDT[1]: Pattern coloring has been controversially discussed at another meeting. The opinion of those present at the pattern committee meeting was that the present standard of colors should be changed. In my opinion, the committee that is supposed to do the recommending should not have the last word. I also feel that this matter should be brought to the attention of various pattern groups and Chapters throughout the country, for their opinions for the purpose of standardizing on the most satisfactory universal color for patterns and core boxes throughout the United States.

I noticed in some parts of the country patterns are painted brown, and the core prints black. In other parts of the country pattern shops conform exactly to the standards of the American Foundrymen's Association's present specifications. If,

however, we are to set up a new standard on colors for painting patterns and core boxes, the committee will be able to do a much better job by getting the opinion of various pattern groups.

Now, it was also brought up during the discussion of the committee that the body of the pattern should be painted clear orange in order to be able to inspect the construction of the pattern itself. By so doing the wearing qualities of the pattern would be jeopardized because a pigmented color has considerably more wearing qualities. Here wearing qualities are sacrificed for eye appeal.

I would further like to suggest that in order to get standard colors throughout the country, the vendors or manufacturers of pattern coatings should have a standard color chart which would be adhered to rigidly. The cans or packages should be marked "Colors Approved By The American Foundrymen's Association."

MR. CECH: Would you mind reading the colors that were suggested for the stop off?

MR. KINDT: There is one color that was ignored entirely at the meeting and that color is chrome yellow. The committee suggested painting the body of the pattern clear orange, the core prints black, the finished part of the casting red or vermilion, the loose pieces green, and the stop-offs with aluminum.

[1] The Kindt-Collins Company, Cleveland.

Effect of Section Size and Chills on Physical Properties at Central Portion Of Sand Cast Bronze Test Coupons

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DESIGN OF STRUCTURAL BRONZE castings is based to some extent on the inherent tensile properties of the alloy used. These properties are generally determined by testing specimens from a standardized test casting. However, it is interesting to note that for some alloys, the properties of test castings are generally far superior to the properties of commercial castings, particularly those with heavy cross-sections.

This may be interpreted in another way, namely, that even though the quality of metal is satisfactory based on the tensile properties of the test casting, the commercial item may leak or lack the required strength because of the spongy or porous structure found in heavy sections of certain alloys.

Intent of Paper

The intent of this paper is to show the effect of section size and chills on the properties at the central portion of sand cast test coupons of four commercial bronze alloys. These four alloys, namely, red brass, gun metal, regular and high strength manganese bronze, were selected because they are representative of distinct types of bronzes commonly used in industry.

One type, into which category red brass and gun metal fall, is characterized by its long freezing range, critical pouring temperature, tendency to form intercrystalline shrinkage voids and inability to feed properly even though the principles of directional solidification are followed. This type of alloy, which will be called Type 1 for simplification, generally produces spongy or porous metal in the center of heavy sections.

In contrast to this is the other type of bronze, which will be called Type II, into which category regular and high strength manganese bronze fall. Type II alloy is characterized by its short freezing range, high shrinkage, ability to feed directionally, and soundness in the center of adequately fed heavy sections.

Experimental Procedure

Two match-plate patterns were prepared, one for red brass and gun metal (Fig. 1) and the other for manganese bronze (Fig. 2). Each pattern comprised four test castings (2x2x5 in., 1½x1½x5 in., 1¼x1¼x5 in. and 1x1x5 in.) gated and fed from a common runner and riser.

The pattern designed for red brass and gun metal (Fig. 1), in addition to the four test castings, contained one standard Crown type 0.505-in. diameter cast-to-size test bar and one standard Crown type 0.505-in. diameter machined-to-size test bar, both of which were gated into the common riser.

The pattern designed for manganese bronze (Fig. 2), in addition to the four test castings, contained a solid piece, 1x1x10⅝ in., gated along its full length into the common riser, and from which two standard 0.505-in. diameter test bars could be machined. The riser for the manganese bronze test mold was designed larger than the riser for the red brass and gun metal test mold to compensate for the excessive shrinkage of manganese bronze.

The standard test bars were attached to the castings to provide a check on the quality of the metal and to provide data for a comparison with the properties of the heavier sections. The molds were prepared with an Albany green sand of about 20 permeability which was rammed to have a 30 A.F.A. mold hardness.

Density and Tensiles

Density and tensile properties of red brass and gun metal vary with pouring temperature, and for this reason a series of six test molds were prepared for each alloy. Four of these same molds were cast at 100° F., intervals beginning at 2200° F. and ending at 1900° F., and the two remaining molds were poured at 2100° F.

The test castings in one of the remaining sand molds were chilled with a metallic chill equal to the size and volume of the test casting, thereby providing sufficient material poured at 2100° F. for a comparison of the density and tensile properties between chilled and sand cast material varying from 1 to 4 sq. in. in

Effect of section size and chills on properties at central portion of sand cast test coupons of four commercial bronze alloys is shown. These four alloys, namely, red brass, gun metal, regular and high-strength manganese bronze, were selected because they are representative of distinct types commonly used in industry.

Presented at a Brass and Bronze Session during the Fiftieth Annual Meeting of the American Foundrymen's Association in Cleveland, May 8, 1946.

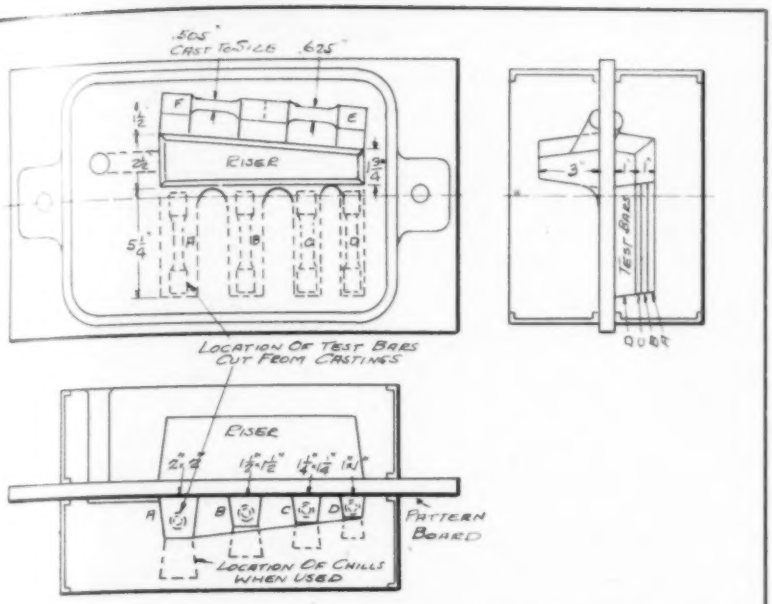


Fig. 1—Drawing of match plate pattern for red brass and gun metal test bar castings.

cross-sectional area.

The procedure followed for regular and high strength manganese bronze differed from the procedure for red brass and gun metal in that the two molds required for the study of chilled and sand cast material were cast at 1900° F., and only one mold was poured at a lower temperature (1780° F. for regular manganese bronze and 1750° F. for high strength manganese bronze).

In order to determine the variation in soundness of each casting, the density of sections A, B, C, D, E, and F shown in Fig. 1 and of sections A, B, C and D shown in Fig. 2 was determined. Following the density determinations of the entire rough casting, a standard 0.505-in. diameter test bar was machined from the central portion of sections A, B, C and D from each casting.

Section F, present only in red brass and gun metal castings, was tested cast-to-size, and section E in red brass and gun metal castings was machined-to-size prior to determining density and tensile properties. Since the density in sections A, B, C, and D of the manganese bronze castings did not vary, it was not determined for section E.

Discussion of Results

Test results are tabulated in Tables 1, 2, 3, and 4, and data for each alloy cast at the optimum temperature into both chill and sand molds are expressed graphically in Fig. 3, 4, 5, and 6, in which density, elongation, yield strength, and tensile strength are plotted versus cross-sectional area of each specimen.

Except for density determinations, the properties of sections A, B, C, and D were obtained from the central portion of each section, but are represented as the properties of the entire cross-section. This is undoubtedly true for the sections with small cross-sectional areas, i.e., 1x1 in. (D), 1¼x1¼ in. (C), but the average properties of sections with large cross-sectional areas, particularly those which were sand cast, may be somewhat higher in view of the fact that the dense and strong skin of such castings has been completely removed by machining.

However, this does not detract from the value of the data. Variation in tensile properties in castings of variant cross-sectional areas is less pronounced in gun metal than in red brass. Moreover, in all cases there is a striking and significant difference between the density and tensile properties of castings which were sand cast and castings which were chilled in sand molds.

Sections E and F

Data for sections E and F represents more nearly the data normally obtained from a standard test casting, such as the Crown-type casting used for red brass and gun metal and the Keel-block used for manganese bronze. These data are not joined with the remainder of the curve because the scale is varied for these points and the castings are a different shape than the sections A, B, C, and D.

However, the differences between the properties obtained from standard test castings and those obtained in a commercial casting of variant section thicknesses are suggested by careful study of the data presented in the accompanying figures.

Red Brass (Fig. 3) cast in sand molds appears to be more susceptible

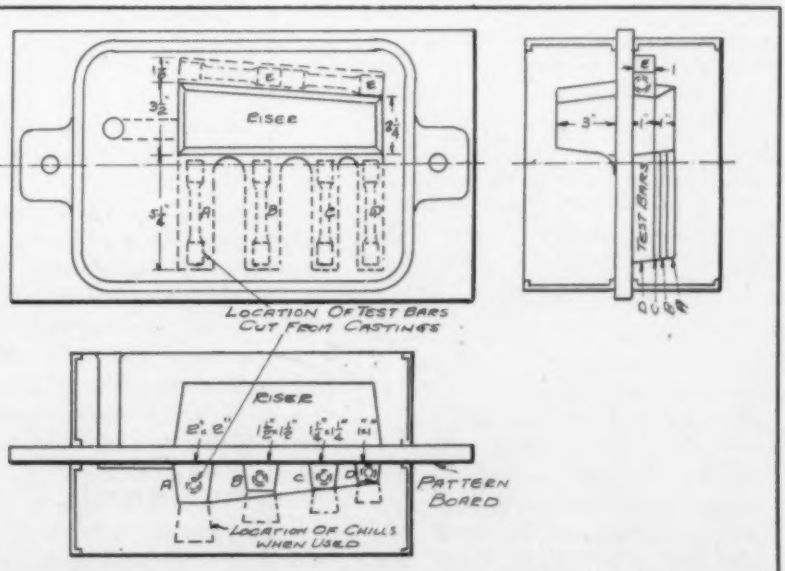


Fig. 2—Drawing of match plate pattern for manganese bronze test bar castings.

to changes in section thicknesses than gun metal and the two manganese bronze alloys. The tensile strength quite obviously decreases with increases in cross-sectional area. As it is to be expected, the density follows the same trend, while contrary to what was expected, very little if any change in yield strength and elongation is noted with variations in cross sectional area.

There are two outstanding observations in the case of red brass and gun metal (Fig. 4): the properties of the test bars (Sections E and F) are far superior to the properties of sections A, B, C, and D, and the density and tensile properties of sections A, B, C, and D are remarkably improved by chilling. Properties of the chilled castings also approach more nearly the properties obtained from the standard test bars (Sections E and F).

Tensile Requirements

While tensile properties of the standard test castings and the chilled sections A, B, C, and D of both red brass and gun metal meet the tensile requirements, the tensile properties of the same sections without chills fall short of the minimum requirements.

Complete details are presented on red brass and gun metal in Tables 1 and 2, respectively, where it is indicated that of the two alloys, red brass is far more sensitive to changes in pouring temperature and section thickness than gun metal.

Both manganese bronze alloys do not show a significant change in properties which could be related to cross-sectional area. (Fig. 5 and 6, and Tables 3 and 4). Densities of sections A, B, C, and D are relatively the same for both sand cast and chilled sections, though there is a very slight decrease in density with an increase in cross-sectional area.

Chilling Benefits

Two important observations can be made, namely that chilling has a beneficial effect on the tensile strength, yield strength, and elongation of the high strength manganese bronze, and while it improves the tensile strength and elongation of the regular manganese bronze, it does not appear to influence the yield strength of this alloy.

The tensile properties of the 0.505 in. standard test bars machined from the 1x1x10⅝ in. section E are approximately the same as the tensile properties of specimens obtained from sections A, B, C, and D. Chilling did not improve the tensile properties of manganese bronze to the extent that it improved the same properties of red brass and gun metal.

The structure of red brass and gun metal is considerably refined by the use of chills. Use of chills equivalent in size and volume appears to be adequate for complete chilling. It is possible that the size of the chill could be increased or reduced and the same depth of refinement of structure obtained.

However, since the structure appears to be refined completely throughout the cross section it is doubtful if an increase in the size of the chill would render a further improvement in the material. It would seem that the dense structure obtained by chilling coupled with the improvement in tensile properties should, among other things, be conducive to the production of leakproof heavy section pressure castings.

Table 1

PROPERTIES OF CHILL AND SAND CAST RED BRASS 85-5-5-5[1]

Specimen Number	Heat Number	Pouring Temperature, °F.	Density, GMS./cc.	Yield[2] Strength, psi.	Tensile Strength, psi.	Elongation, % in 2 in.
1A	A	2200	8.47	11,900	12,400	6
B			8.54	14,100	18,800	12
C			8.53	14,200	16,500	8
D			8.48	16,400	24,400	11
E			8.72	17,800	36,200	28
F			8.81	16,700	40,000	38
2A	A	2100	8.66	15,900	22,400	8
B			8.71	15,900	24,800	13
C			8.70	16,400	27,200	15
D			8.73	13,500	28,000	10
E			8.86	17,800	39,600	41
F			8.82	16,800	38,000	31
6A[3]	B	2050	8.89	19,200	39,600	32
B			8.89	19,000	39,000	28
C			8.87		37,800	22
D			8.91	19,200	34,800	16
E			8.93		38,300	30
F			8.94	16,000	41,300	41
3A	A	2000	8.74	15,900	20,500	10
B			8.82	16,800	30,800	16
C			8.81	17,400	33,800	21
D			8.88	17,700	23,900	9
E			8.99	17,600	40,100	41
F			8.85	18,200	41,300	44
4A	A	1900	8.77	18,300	25,700	9
B			8.87	18,800	29,000	13
C			8.86	18,900	28,800	13
D			8.91	19,400	30,600	13
E			8.91	19,500	38,400	22
F			8.92	15,500	40,500	27

[1] Composition, per cent: Cu—84.45, Sn—4.42, Pb—4.71, Fe—0.22, Sb—0.12, Ni—0.66, S—0.054, Zn—Bal.
[2] 0.010 in. deflection in 2 in.
[3] A, B, C, D, chilled with casting, equivalent in size and volume.

Summary

In view of the foregoing presentation, it seems that the following generalizations can be made.

1. The density and tensile properties at the central portion of test coupons of sand cast red brass decrease to a great extent as the cross-sectional area increases. Though the same trend is indicated for gun metal, the properties vary to a limited extent only, excluding the properties of the attached standard test bars. The pouring temperature of red brass seems to be more important than the pouring temperature of gun metal.

2. The density and tensile properties at the central portion of test coupons of regular and high strength manganese bronze alloys do not vary with changes in cross-sectional area to and including 4 sq. in., nor do these same properties change with variations in pouring temperature.

Table 2.

Properties of Chill and Sand Cast Gun Metal 88-8-4[1]

Specimen Number	Heat Number	Pouring Temperature, °F.	Density, GMS./cc.	Yield[2] Strength, psi.	Tensile Strength, psi.	Elongation, % in 2 in.
16A	F	2140	8.64	17,200	35,300	25
B			8.57	16,700	34,000	27
C			8.63	16,700	35,900	30
D			8.67	19,400	36,200	23
E			8.77	20,300	47,000	48
F			8.55	21,600	51,500	56
17A	F	2080	8.55	17,400	35,000	23
B			8.58	18,100	37,500	30
C			8.64	16,600	34,300	30
D			8.70	18,500	35,200	27
E			8.76	20,800	48,700	55
F			8.83	20,200	50,500	53
21A[3]	G	2100	8.76	21,500	47,500	44
B			8.78	20,800	48,700	50
C			8.74	21,000	49,800	48
D			8.77	21,200	45,200	30
E			8.77	22,000	49,000	44
F			8.83	20,000	50,500	44
18A	F	1990	8.58	18,200	31,800	20
B			8.67	18,800	34,600	20
C			8.69	19,900	36,400	22
D			8.76		20,700	13
E			8.78	21,600	51,300	50
F			8.77	20,000	54,700	66
19A	F	1890	8.55	19,100	29,300	13
B			8.63	18,500	31,100	14
C			8.69	20,000	32,600	13
D			8.73	20,500	32,000	13
E			8.82	23,400	50,300	33
F			8.77	21,900	54,300	44

[1] Composition, per cent: Cu—87.2, Sn—7.87, Pb—0.05, Ni—0.59, Sb—0.01, Fe—0.07, S—0.018, Zn—Bal.
[2] 0.010 in. deflection in 2 in.
[3] A, B, C, D, chilled with casting, equivalent in size and volume.

3. The properties of the attached standard test bars of red brass and gun metal are far superior to the properties at the central portion of the heavier sand cast sections (A, B, C, and D). The properties of the attached standard test bars of these same alloys do, however, approach the properties of the chilled heavy sections (A, B, C, and D). The properties of the attached test bar 1x1x10⅝ in.) of manganese bronze are about the same as the sand cast heavy sections (A, B, C, and D).

4. Chilling heavy sections of red brass and gun metal has a remarkable beneficial effect on density and tensile properties, and appears to be slightly beneficial to the tensile strength and elongation of regular manganese bronze and the density and tensile properties of high strength manganese bronze.

Appendix

Review of the Literature. Few references are found which present data to show the effect of section size and use of chills on bronze. A recent paper, however, by Brouk, Hardy, and Loring[1] shows the distribution of mechanical properties in "G", "M", red brass, and special nickel-containing tin bronzes.

Their investigation covered a study of the properties of these alloys at the corners and center of a heavy section casting, 3x6x8 in. The results of their tests indicate that the density and tensile properties of "G", "M", and red brass are highest at the casting corners and lowest in the center of each casting. This agrees with the results of this investigation particularly in the case of red brass.

The fact that these investigators used a casting with a greater cross-sectional area than the one used for this investigation may account for a wider spread in properties found in the alloys. An interesting point which was brought out in their paper was the very small difference in properties throughout the cross-section of "G" type alloys in which 3 and 6 per cent nickel replaced an equivalent amount of tin as compared with the nickel-free "G" metal.

These nickel-containing "G" metals have comparatively short freezing ranges and because of this, approach the desirable characteristics (i.e., lack of intercrystalline shrinkage) of the manganese bronze type of alloy. This may account for the rather uniform properties between skin and center of castings prepared with this alloy, or manganese bronze.

Bryant's[2] descriptive article on chills in bronze castings states that chills frequently equal the weight of castings, that some are plain or flat, whereas other follow the contour of the casting, and that chills should be at least as thick, or preferably twice as thick, as the section in the casting to be chilled.

References

1. R. H. Brouk, R. G. Hardy and B. M. Loring, "Distribution of Mechanical Properties in Sand Cast Bronzes," American Foundryman, pp. 61 - 66, July, 1945.

2. J. W. Bryant, "Chills Used on Bronze Castings," *The Foundry*, vol. 71, no. 12, pp. 109, 168 and 169.

DISCUSSION

Chairman: B. A. Miller, Baldwin Locomotive Works, Cramp Brass & Iron Foundries Div., Philadelphia, Pa.

Co-Chairman: C. Louis Lane: Florence Pipe Foundry & Machine Co., Florence, N. J.

B. M. Loring[1]: This was a very timely paper and it was very well presented. In some of our work in heavy sections, we have noticed that melting practice is quite important. It is possible to increase sometimes the mass effect by differences in melting practice. Undoubtedly, though, there was a very uniform melting practice in all of this work.

In manganese bronze the mechanical properties are less dependent on section size. In this paper only a certain range of section sizes were considered. If the author had gone up to 10 or more in. in thickness, then the section size would have become more important, perhaps accounting for 10 per cent of the mechanical properties. Just why this should be, I am not quite sure. Probably the grain size of the manganese bronze as well as the distribution of the iron particles may have something to do with it.

John Bolton[2]: The authors should be complimented for bravely tackling a difficult problem. However, they have done things in their testing technique that, in my opinion, are in error. They have taken four different sizes of blocks and some conventional test bars; they have made them on a common gating

[1] Naval Research Laboratory, Washington, D. C.
[2] Lunkenheimer Corp., Cincinnati.

Table 3

PROPERTIES OF CHILL AND SAND CAST REGULAR MANGANESE BRONZE[1]

Specimen Number	Heat Number	Pouring Temperature, °F.	Density, GMS./cc.	Yield[2] Strength, psi.	Tensile Strength, psi.	Elongation, % in 2 in.
25A	H	1900	8.23	30,400	77,400	23
B			8.23	30,800	77,700	23
C			8.25	31,400	77,100	22
D			8.25	31,500	78,100	22
E				30,900	78,200	23
24A[3]	H	1900	8.16	30,000	79,800	31
B			8.24	30,800	80,600	34
C			8.24	30,400	80,700	36
D			8.27	32,200	80,600	34
E				29,500	78,500	28
26A	H	1780	8.23	30,300	79,000	27
B			8.24	30,800	78,500	25
C			8.23	31,900	77,700	25
D			8.28	30,600	79,600	25
E				30,900	78,200	23

[1] Composition, per cent: Cu—57.17, Pb—0.03, Fe—0.96, Mn—0.36, Al—0.85, Zn—Bal.
[2] 0.010 in. deflection in 2 in.
[3] A, B, C, D, chilled with casting, equivalent in size and volume.

system; they have used a common riser; and they have used a common pouring temperature. Now, how many of you would take four different castings in your foundry, of different sizes, and cast them all at the same temperature, with the same riser? Perhaps the authors will supplement their paper by showing the relationship of the structures they produced to the properties.

Some years ago Dr. Smith did this on somewhat similar alloys and followed the relationships of pouring temperature to the relative dendritic versus equiaxed crystallization.

I do take exception to some of the implications made in the verbal presentation. We, as foundrymen, are dealing with engineers, designers, who want certain properties in their castings. To make a blanket statement that the test bar properties are vastly superior to the properties in the castings is misleading. There is a degree of truth to it, we know that, but to use the method of comparison that is used and still emphasize disparities that are partly due to techniques that were not suited for all types of bars leads to an exaggerted picture in the matter.

I think there is no question but that as the Non-Ferrous Division of A.F.A. gets more into fundamentals, that the horrible picture that is now alleged will be seen to be not as bad as alleged. In the non-ferrous foundry field, more of the foundation work is needed, the relationship of the structure, for example, and the things I pointed out as shortcomings in the technique of the authors are things that need further work. Too few people are interested in the mechanics of these things. Until we find out more about the WHYS, we are never going to discover the proper HOWS.

MR. GEORGE: We know that the furnace atmosphere has a considerable bearing on the structure of the castings. We know when we are running about 1 per cent of oxygen in the atmosphere, we get the best physical properties, which we have tried to do in this case. In fact, before these bars were ever run, we ran a number of tests, a number of fractures, to see that the bars were right.

If we have to meet certain physical requirements on a certain job and manganese bronze is suited for it, why specify 85-5-5-5 and get 40,000 psi. tensile in the bar and 22,000 psi. in the casting, when we could substitute manganese bronze and get 70,000 psi. tensile in the bar and 70,000 psi. tensile in the casting? We tried to bring out the fact, that some test bars represent the castings which they are made for and some do not.

CHAIRMAN MILLER: I personally would more or less take the middle road between Mr. Bolton and Mr. George. I am glad that Mr. Loring made a remark in reference to getting up to a 10-in. cross-section. I am sure that the Navy know that they do not have in propeller wheels, especially in the hub of propeller wheels, a 65,000 psi. tensile and 20 per cent elongation. I think they have understood that for a long time. It is true that in the instance of propeller wheels, the test bars are attached to the hub as well as to the blades, but I know in the innermost part of that particular hub, they do not expect 65,000 psi. tensile and 20 per cent elongation.

MR. LORING: That is quite right.

W. ROMANOFF[2]: The discussers before me expressed the same opinions I had when I saw and heard the paper. We know definitely that the melting conditions have much to do with physical properties. As Mr. Bolton mentioned, various section thicknesses should not be poured at the same temperature with the same riser.

A statement was made whether to use a manganese bronze or some high strength alloy in place of 85-5-5-5, because the physicals were greater. Of course there are other considerations. Suppose we were making a pressure casting and desired high resistance to leakage. Certainly, we would not necessarily want an alloy merely because it had a high tensile strength and high elongation, without taking into consideration its castability.

I do not know whether Mr. Bolton was present at one of the ASTM meetings a number of years ago where a number of competent men made the suggestion that the test bar method was entirely inadequate when trying to determine whether a metal would be pressure resistant in a valve or friction resistant in a bearing. It was felt that some simple pressure or anti-friction test would be worked out and it would not be a question of tensile strength or elongation. I still think that method of testing for alloy determination is not too far off.

MR. BOLTON: May I make one more remark to make this matter entirely clear? Some years ago, back 25 years

[2] H. Kramer & Co., Chicago.

Table 4

PROPERTIES OF CHILL AND SAND CAST HIGH TENSILE MANGANESE BRONZE

Specimen Number	Heat Number	Pouring Temperature, °F.	Density, GMS./cc.	Yield[1] Strength, psi.	Tensile Strength, psi.	Elongation, % in 2 in.
27A	I	1910	7.80	65,900	107,000	14
B			7.81	63,100	104,000	14
C			7.81	64,000	106,000	13
D			7.81	64,000	110,000	19
E				63,200	108,000	19
29A[2]	I	1910	7.81	63,400	110,000	22
B			7.79		110,000	22
C			7.84	65,000	110,000	22
D			7.84		111,500	22
E				65,200	109,000	20
30A	I	1750	7.81	64,200	106,000	14
B			7.82	63,500	106,000	16
C			7.83	62,700	95,500	8
D			7.82	66,200	72,700	3
E				62,700	105,500	16

[1] 0.010 in. deflection in 2 in.
[2] A, B, C, D, chilled with casting, equivalent in size and volume.

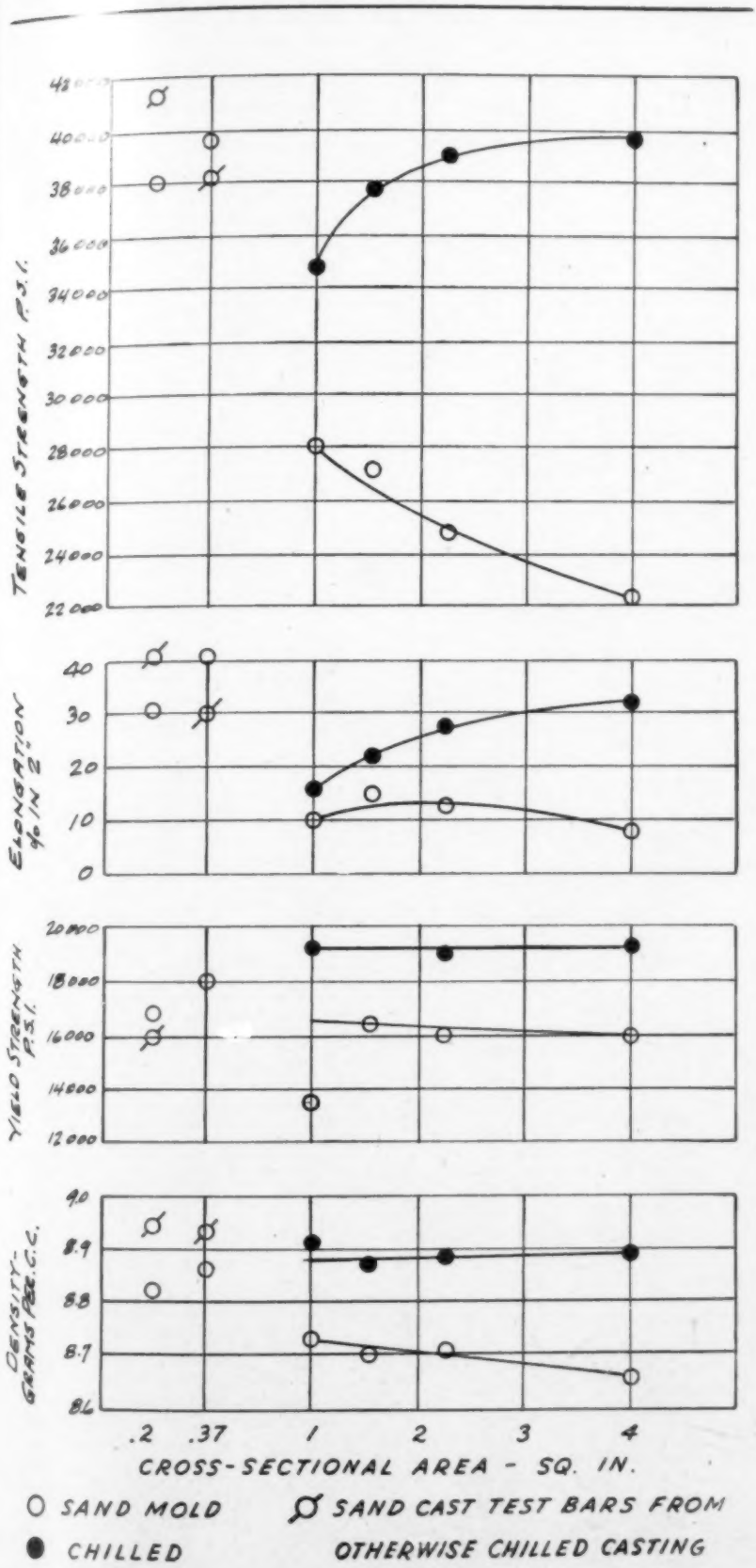


Fig. 3—Graphs show effect of section size and chills on the physical properties at the central portion of sand cast test coupons of 85-5-5-5 alloy.

or better, I had just exactly the same viewpoint expressed concerning cast iron as I have expressed today concerning bronze. We know and have known for years that the test bar may not represent the casting. We in the iron foundries got busy and found out how much more nearly alike we could make the test bar and the casting, and I am going to tell you a little story about that.

We had recently a cast iron 24 in. valve body with a 2-in. wall thickness. The regular 1¼-in. test bar had a tensile of approximately 40,000 psi. We took bars all over that casting after it had been broken up and made stress analysis tests. I think the lowest tensile we got on the bars from the casting was 38,000 or 39,000 psi., and some ran up to 40,000 psi. or better.

But what had heppened in the research field! There was a symposium, a series of papers. Much research was done, and many of us found a lot of divergence, but we worked out methods and means of producing results in cast irons. We got the proper cast irons and devised proper heat treatments and proper gatings. I would not ever guarantee you will get 44,000 psi. tensile average, in bronze castings to which Mr. George refers, but I will say we should not think either of going to the other uneconomical and tricky method of chilling, which is, to my mind, rather impracticable for many castings. Although you may not get 40,000 psi. tensile, we may find some method of inoculation that will enable us to get 35,000 or 38,000 psi. even in heavy brass castings.

Mr. Weigand and I worked on some heavy stem castings years ago that were worthless even though the test bars were good. We used certain little tricks including proper directional solidification by means of re-gating and we got the casting tensiles up nearer to the test bar values.

If you have not read early ASTM papers on this subject, I suggest that you read Dr. Smith's review, covering the whole test bar situation from the time it was started back in 1900 or before. Your point was well known at that time and figures similar to yours were published at that time, but metallurgists and foundrymen have not done enough about it in the intervening time.

MR. GEORGE: It was my opportunity during World War II to visit nearly all the non-ferrous foundries of this country and it was surprising to see the different test bars used in the 85-5-5-5 alloy and others of the red brass series. I believe there were about 500 of them. We had some foundries that consistently obtained 40,000 psi. tensile with the 85-5-5-5 alloy, while others could not get acceptable values on a single bar. Some foundries had 30 lb. risers on the 85-5-5-5 alloy bar and they still could not get acceptable tensiles. We had other foundries that had only a 1¼-lb. riser per bar and they obtained acceptable tensiles on practically every one of them. It is time for somebody to try to correlate cooling rate between the metal and the bar. We know the bars cast to

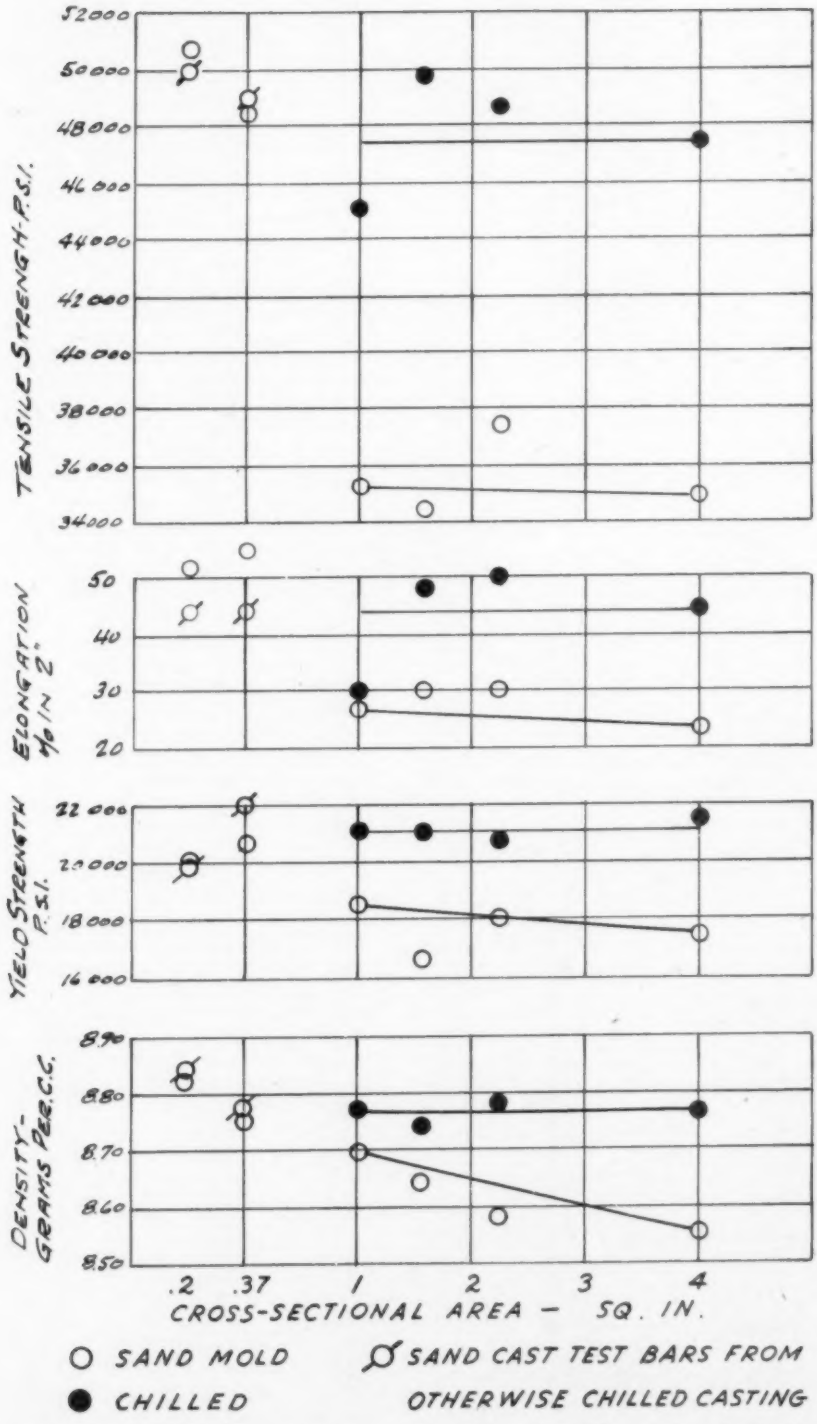


Fig. 4—Graphs show effect of section size and chills on the physical properties at the central portion of sand cast test coupons of gun metal.

size pull the highest, due to the cooling rate in the sand.

W. M. BALL, JR.[4]: I do not believe the test bar means much to the engineer. Having made test bars for a number of years I am positive that test bars do not give the engineer what he wants to know. We recently had a paper given at the Cincinnati Chapter of A.F.A. on "What is Strength" by J. B. Caine of the Sawbrook Steel Castings Co. He gave a lot of good figures and much data on the physical testing of castings.

In his paper he definitely showed that many castings were not failing from tension but were failing due to fatigue which is caused by faulty design. Resulting fatigue forces concentrating in the weakly-designed part of the casting cause a hair-line crack to develop.

He also showed results of tests that were made from the part of the casting within ½ in. of the crack and these results showed the casting to be physically correct. I could give many specific examples of experience where the test bar was in no way indicative of the quality of the casting I had made. I believe that Mr. Caine has in his papers contributed a fine thought for guidance in our future physical testing.

MR. GEORGE: I agree with Mr. Ball to the extent that the test bar should be disconnected from the casting. However, I believe the designers do not realize the facts presented here. Men like Mr. Bolton and Mr. Ball who have been in the field for years and around in different technical societies know these facts. However, they are only a small minority. We tried to bring it out in this paper for those who do not know it and those who are designing machinery, that there is a certain type of casting where the tensile of the bars comes near the tensile of the casting and if they want certain physical properties, to recognize that type of alloy and not another type alloy, like 85-5-5-5, where there is no representation whatever.

MEMBER: It is very interesting to me to see how closely the situation which evidently prevails in the brass and bronze industry parallels that which prevails in the aluminum industry.

During the first World War, many aluminum castings were made with test bars cast all over and a great deal of effort and time were spent on this very situation and, as you have, the aluminum industry found, after much argument with the government services and so on, that we just could not tell what the properties of the casting were going to be by pulling a bar that was attached to the casting.

Many studies were made correlating the mechanical properties of the bars attached to the casting with those actually machined from the casting, and many different types and sizes of castings were investigated. As a result of that and subsequent work, the cast-attached test bar has been substantially eliminated, because it gradually has been realized, at least on aluminum castings, that the properties of the cast-attached test bar seldom represented a good picture of what the properties were that were found in the casting itself, for the reason that if you attached the bar near the gate, you got rather low properties in cast-attached parts. If you poured the casting in exactly the same way, you could attach the bar to various other parts of the casting and substantially get any properties you wanted.

As a result of those studies and of considerable customer experience, the cast-attached bar has practically disappeared and been discarded. Now, separately-cast test bars are used almost entirely. There may be some few cases where attached bars are used on castings but I do not know of any today. Of course, that does not mean there may not be some. It has been decided substantially that tensile test bars of standard

[4]Magnus Brass Div., National Lead Co., Cincinnati.

sizes can represent only the properties of a definite aluminum alloy as they are affected by melting conditions, the action of grain refiners, heat treatment, and composition.

The foundry practice, therefore, for each type of work or each job is developed to produce the most satisfactory casting for the particular application. The separately-cast test bars, when cast, mean only that the melting conditions have been within the specification; the composition is within the specification. That, however, is a rough check, because, chemical analyses are also made, and then, too, it is quite true that the chemical composition can be somewhat outside specified limits in some alloys and still obtain satisfactory properties if the heat treatment has been correct.

So in the aluminum industry today, I believe I am correct in saying that the separately-cast test bar is an indication of the efficacy of the melting practice, a general indication of composition, and an indication of correct heat treatment and nothing else.

A great deal of information has been built up on the mechanical properties obtained from various sections of aluminum alloy castings and, as a result of a mass of such information, government specifications are set up today to the effect that bars machined from the thinnest section, an intermediate section and the thickest section of an aluminum alloy casting covered by that specification shall average 75 per cent of the separately-cast test bar minimum tensile strength and 24 per cent of the minimum elongation as established by separately-cast test bars. So the government services, such as Army, Navy and Air Corps, realize that the tensile properties, elongation, tensile strength of aluminum alloy castings are not going to be the same as those in the casting and are only an indication of the efficiency of the melting operation, roughly of the chemical composition, and of the heat treating conditions.

MR. BOLTON: Under the ASME Boiler Code for many of these castings, what stress is permissible in the casting?

MR. GEORGE: I could not answer that.

MR. BOLTON: There lies the fundamental reason for our objection to your presenting your data in the negative way. You will find the design stresses are in the neighborhood of 2500 to 3500 psi. Therefore, if your material in the casting is only 22,000 psi. that tensile strength would give you a very large safety factor.

MR. GEORGE: Not necessarily, because the yield strength does not follow the tensile strength. The tensile strength can fail and the yield strength still stays up very well.

MR. BOLTON: Yes, the engineer is working to the yield strength, below which distortion does not take place in any permanent degree. Therefore, if you are getting 22,000 psi. tensile and the yield has dropped down to but less in proportion let us say, 15,000 or 16,000 psi. as compared to normal 18,000 psi. you still have a big margin, even at a better ratio (variation, bar to casting) than your tensile figures suggest.

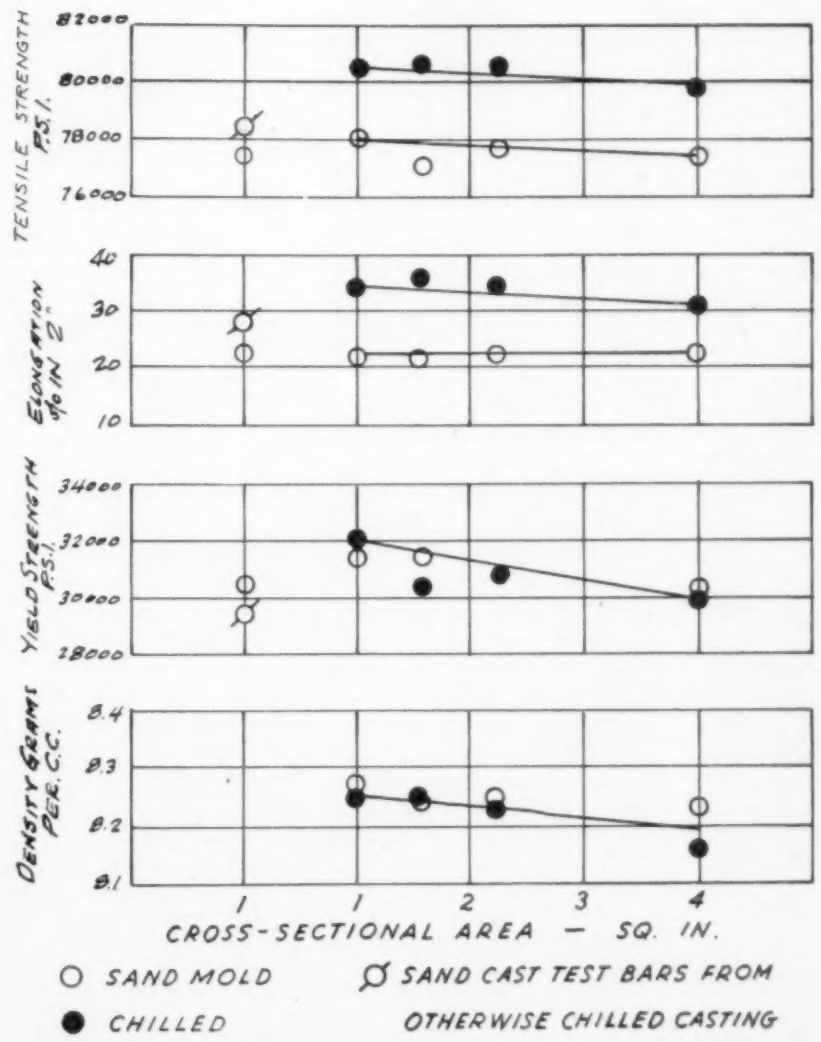


Fig. 5—Graphs show effect of section size and chills on the physical properties at the central portion of sand cast test coupons of regular manganese bronze.

MR. GEORGE: How many of our Navy specifications call for yield strength? Practically none. Up to this time, it has only been talked about. I know from contacting many engineers that they talk about yield but they pay no attention to it when it comes to making their product.

MEMBER: I want to say that the test bar does not definitely represent the casting. In some parts of the foundry you make the test bar and in other parts of the foundry you make the casting. You are working under absolutely different conditions. If you have an ideal condition, where you can set up a definite practice to make a test bar and the same practice to make the casting, then your physical properties should be equally the same, under normal conditions. During the war many foundries were forced to throw out castings which were actually perfect because the test bar did not meet physical properties.

What I am trying to bring out here is that we have many factors entering into the maximum physical properties of a test bar. The most important one is melting practice. If you do not have good melting practice you cannot produce a good bar of any design. Poor baking or coremaking facilities also are factors. Ingredients that go into making a core or into the sand has something to do with the production of gas within the mold. If you are not able to free that, you are not going to get physical properties. So you have many factors entering into the casting of a test bar, yet that test bar is used in some specifications for the acceptance of castings. We must make some standard which the foundryman can use to say whether the physical properties of his metal are satisfactory.

I have taken several of these so-called test bars of various designs and I have gotten perfect test bars with all of them, simply because we try to control our melting, our core-making and our molding practice. In other words, we eliminate the possibility of so-called gas defects, which includes gating as well, to get a good test bar, and any one of these test bars is satisfactory provided you can set

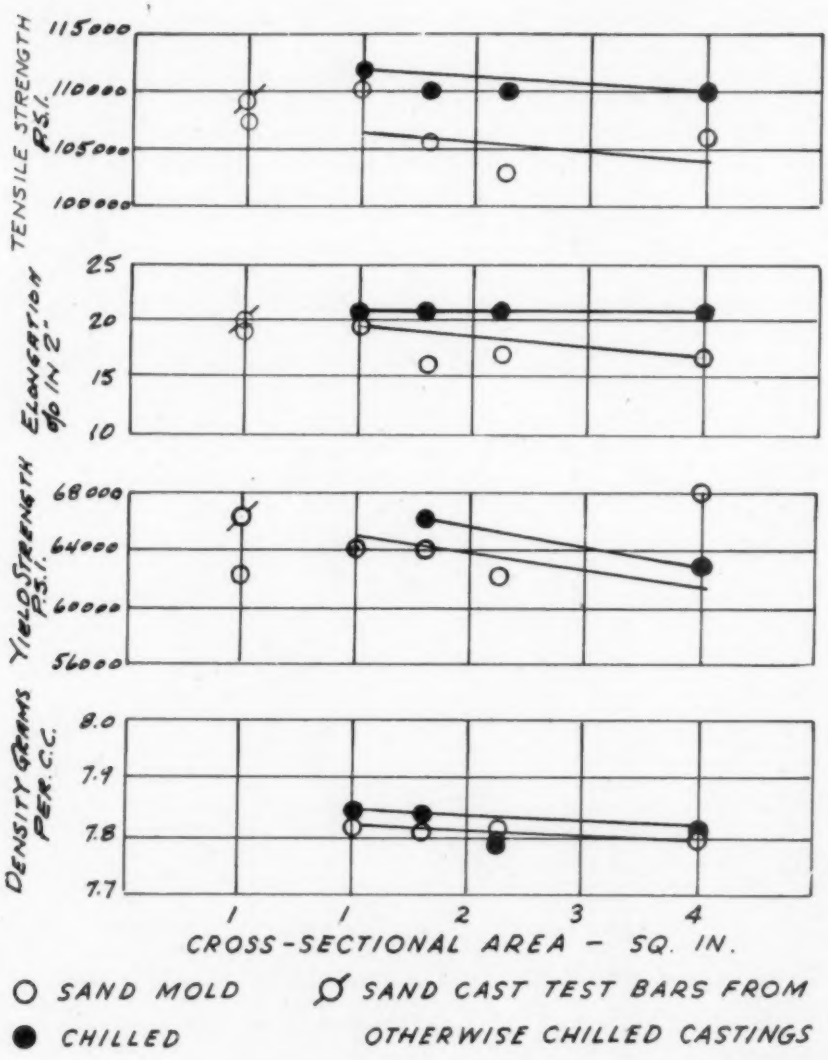


Fig. 6—Graphs show effect of section size and chills on the physical properties at the central portion of sand cast test coupons of high strength manganese bronze.

up a good foundry practice.

MR. ROMANOFF: There is one statement that was made during this discussion that I think should be corrected. That is the statement that test bars cast to size give the best physical properties. They do when the metal is poor. If the metal has been melted and poured right and all foundry practice corrected, that is not true. It is probably more economical and more convenient to make them cast to size but it does not give the best results if the metal is right and the foundry conditions are right.

C. L. FREAR[5]: The question of test bars appears to me to be tied up with the design and placement of risers on castings. Considerable discussion pointed to the fact that the properties existing in castings do not agree with those found in the test bar. I do not see how, with the present state of the art, we could expect the test bar to show the actual strength of the casting. Test bars in most cases are poured under conditions as nearly ideal as possible. With this in mind, why should not we get a good test bar? On the other hand, I doubt if we could find one casting out of a thousand which can be molded, poured and cooled under such ideal conditions. How then can we expect the casting to have the same strength as the test bar?

The best thing that we can do is to admit that we do not obtain the same properties, but at the same time try every means at our disposal to determine the cause of the variations, which are certainly greater with many of the nonferrous metals than we can explain by differences in section. There must be some other causes for such variations amounting oftentimes to discrepancies, due to variations in foundry procedure and probably in a number of cases to inherent tendencies to unsoundness in the metal itself. This leads to the question of the melting procedure and certain properties of the metal which we do not know at the present time but on which we will make rapid progress and, I hope, will find some of the answers within the next year or two.

In the meantime, I am strongly in favor of test bars, because if a foundry cannot pour a good bar it can hardly be expected that castings made under much less ideal conditions could be good.

In connection with the question of obtaining satisfactory test bars, it is my opinion that one of the variables which is often overlooked is the necessity of keeping the risers open. As I picture the action of metal solidifying in a mold, an open riser will feed as long as it is open to the atmosphere. The instant that it freezes over, thus sealing off any connection with the atmosphere, we might as well not have a riser, in fact, such a sealed-off riser probably does more harm than good.

The method described in a paper entitled "New Type Exothermic Powders For Making Sound Castings" by C. G. Lutts, J. P. Hickey, and M. Bock II in the Aug. 1946 issue of the AMERICAN FOUNDRYMAN appears to be an excellent method of keeping risers open, but the efficacy of this and other methods which have been described is dependent on other things such as properly melted metal, correct design and using the right molding procedure.

MEMBER: I believe that your test bar is an excellent check on your foundry practice as far as melting and possibly the molding of the test bar are concerned. Beyond that, I cannot see how it can be correlated with the casting. A few years ago we actually made test bars and castings and tested them to destruction. We could find no relationship between good test bars and good castings. We made good test bars, but many times our castings did not have as good physical properties as our test bars, and that is the case in 95 per cent of the cases where test bars and castings are made.

H. J. ROAST[6]: I think that the key to the situation is design. The foundry is frequently called upon to make castings that cannot, by the very nature of the differences in cross-section, ever simulate the test bar. Of course, I entirely agree that there is only one job for a test bar and that is to state that the metal going into the casting is good. I can see no other use for it. The engineer is primarily interested in buying physical constants in the casting, or should be. Yet the engineer will design in a manner to make it impossible to get the results of the test bar in the casting.

I suppose I am already known as the proponent though not the originator of the keel block test bar of about 40-lb. weight. I believe that this makes a good selling point to the engineer because he believes at once that you are not trying to improve the metal by casting to size but that you are willing to give it a fair chance in an average cross-section.

G. P. HALLIWELL[7]: It seems to me you throw this meeting wide open by asking for the purpose of a test bar. Mr. Roast has stated it very clearly, namely, to test

[5]Bureau of Ships, U. S. Navy Dept., Washington, D. C.

[6]Canadian Bronze Co. Ltd., Montreal, Quebec, Canada.

[7]H. Kramer & Co., Chicago.

the quality [illegible]he metal in the pot. What you do with [illegible] metal after that by pouring into mol[illegible] containing chills or made of poor qu[illegible]y sand is something else. The function [illegible]f a test bar must be clearly understoo[illegible]and the properties obtained properly interpreted.

G. K. EGGLESTON[8]: There is one point that struck me as significant. In gun metal and in manganese bronze, the variation between the test bar and cross-sectional size seemed to have little effect. But in the 85-5-5-5 alloy, it had a marked effect. I think what we should know, or try to find out, is the reason for that effect, if there is one. I think Mr. Bolton dwelt on that when he said they found the same thing in iron and by correction of analysis and constituents the effect of variation in sectional size was greatly reduced. I wondered if anybody has any ideas why the 85-5-5-5 alloy acts so differently than the other two alloys, as it was discussed in that paper.

MR. GEORGE: What you say has a lot of merit. We do know that certain elements now in the 85-5-5-5 alloy that we considered as impurities are beneficial from the fact they do act as inhibitors and grain refiners. There is chance for considerable debate among foundrymen at times about some batches that will cast solid and some that will not. Many foundrymen have the idea that when the tin is low in the alloy, we get porous castings, and some have the idea that when the tin is high and the zinc is high, we will get tight castings. The 85-5-5-5 alloy range, being wide as to analysis and impurities, I think will stand considerable research, to the end of bringing the properties we show here into a higher range.

MR. EGGLESTON: The only point that might be an indication of the reason is that in your gun metal and manganese bronze you might consider them as single constituent alloys. When you get into the 85-5-5-5 alloy, you have copper, tin, lead and zinc and on the larger sections, where the cooling rate is slower, I think there may be a possibility of a wide variation in the constituents formed. This variation might have a marked effect on the resulting physical properties.

A. H. HESSE (author's closure): The response to the paper is very gatifying and Mr. George and I wish to express our appreciation to those who participated in the discussion. There is one point, however, which has not been mentioned in the oral discussion, and which we feel is sufficiently important to set forth in the following paragraphs.

It seems to be generally agreed that the physical properties obtained from standard test castings of red brass and gun metal do not represent the properties of the sand casting. Let us assume this to be correct and consider the data from another angle. Suppose there is a job on hand to make a casting containing some heavy sections with red brass or gun metal. This particular casting is to be sound throughout, to withstand hydraulic pressure, and to merit government tensile requirements.

The data in the paper show that the use of chills equivalent to the size and volume of the section to be chilled increases density and tensile properties to a considerable extent, thereby increasing pressure tightness, soundness, and tensile strength of castings normally cast against sand. It would therefore seem reasonable to consider the use of chills in the manufacture of the above-mentioned castings other factors governing the production of a satisfactory casting being favorable to the use of chills.

Mr. Eggleston has brought out an important point concerning the reason why the physical properties of 85-5-5-5 alloy was very susceptible while the physical properties of gun metal and manganese bronze varied to a limited extent only with changes in cross sectional areas. The results of an extensive investigation on gun metal, "M" metal, and red brass were reported in the literature in 1943*. The data were obtained from a number of heats so that a statistical analysis could be made. It was shown that changes in pouring temperature affected the properties of red brass to the greatest extent, while "M" metal and gun metal were affected the least, gun metal being the least affected.

The alloys seem to line up in accordance with lead content, with the 85-5-5-5 alloy, carrying the most lead, being affected to the greatest extent. It is quite possible that the lead content of the alloy has an important bearing on Mr. Eggleston's question. It would be interesting to continue this study on a lead-free 85-5-5-5 alloy to check this point.

[8]Barnes Manufacturing Co., Mansfield, Ohio.

*A. H. Hesse and J. L. Basil, "Nickel in Bronze," JOURNAL OF THE AMERICAN SOCIETY OF NAVAL ENGINEERS, vol. 55, no. 1, Feb. (1943).

MOLDING SAND FINENESS TEST

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AT THE PRESENT TIME the accepted test for evaluating the clay substance in a molding sand is the American Foundrymen's Association standard sedimentation method. This test is of limited value because the definition of clay substance as a material composed of particles less than 20 microns is purely arbitrary. In addition, the test yields no information on the distribution of the subsieve particles contained in the sand. These limitations are recognized by the A.F.A. and, according to the FOUNDRY SAND TESTING HANDBOOK[1], this method "has as its principal defect the inability to separate fine silt from true clay. This accounts for the fact that two sands with the same A.F.A. clay content may have different properties."

Morey and Taylor[2] have ably demonstrated that the present A.F.A. method does not give an adequate evaluation of molding sand.

As far as it is known to the authors the first contribution on the subject of determining the size distribution of the subsieve particles in molding sand was presented by Jackson and Saeger[3]. In their pioneer work they used a pipette method and showed that this method was readily adaptable to the determination of the fineness of molding sand. The objective of the present investigation was to compare three different methods of making fineness determinations on molding sands.

Albany and Lumberton molding sands were employed in this investigation. The samples were thoroughly mixed, riddled, mulled at a low moisture content, and stored in mason jars prior to use. The specific gravities of the sands were determined by the standard method for specific gravity of soil[4]. The average values for the specific gravities of Albany and Lumberton sands were found to be 2.698 and 2.676, respectively. The sands tested were dispersed in distilled water and a sodium hydroxide solution used as a deflocculating agent.

Methods and Results

The essential principle of the fineness test methods used in this investigation is sedimentation. When solid particles of various sizes are dispersed in a liquid medium and then allowed to settle freely, a definite relationship exists between the diameter of the particles, the density of the particles, the density of the liquid, the viscosity of the liquid, and the distance that each particle settles in unit time. This relationship is expressed mathematically by Stokes' law which may be written:

$$d = \sqrt{\frac{30\,nL}{980(G-G_1)T}}$$

where

- d = maximum diameter of particle in millimeters,
- n = coefficient of viscosity of the suspending medium in poises (varies with changes in temperature of the suspending medium),
- L = depth of settling in centimeters.
- T = time in minutes (period of sedimentation),
- G = specific gravity of sand particles,
- G_1 = specific gravity of the suspending medium.

Although Stokes' law applies particularly to spherical particles, An-

It is recognized that the present American Foundrymen's Association sand fineness test method is not entirely satisfactory since sands with the same A.F.A. classification may have different properties. In view of this fact other methods have been proposed, but little has been published on this subject. The present investigation was prompted by the need for more data, the primary purpose being to evaluate the merits of the ordinary pipette, the Andreasen pipette, and the hydrometer methods in making fineness determinations on Albany and Lumberton molding sands. The hydrometer method, which is described in detail, was found to be preferable because it yields satisfactory results conveniently in a minimum of time.

Presented at a Sand Research Session of the Fiftieth Annual Meeting, American Foundrymen's Association, at Cleveland, May 8, 1946.

Table 1

RESULTS OF THE REGULAR PIPETTE TEST ON LUMBERTON SAND

Temperature, C	Viscosity of water in poises	Time, min	Diameter "d" in microns — Read from nomograph	Diameter "d" in microns — Corrected according to sp gr of sand	W_1 grams	W_1—0.0095, grams	P, %	100—P, %
28	0.0084	2	31.8	31.5	0.2672	0.2577	20.6	79.4
28	0.0084	5	20.0	19.8	0.2509	0.2414	19.3	80.7
28	0.0084	15	11.5	11.4	0.2365	0.2270	18.2	81.8
28	0.0084	30	8.2	8.1	0.2314	0.2219	17.7	82.3
27	0.0085	60	5.8	5.7	0.2243	0.2148	17.2	82.8
27	0.0085	250	2.9	2.9	0.2152	0.2057	16.4	83.6
27	0.0085	1440	1.2	1.2	0.2081	0.1986	15.9	84.1

dreasen[5] has shown that it can be applied to angular or cubical particles of the same weight.

Regular pipette, Andreasen pipette, and hydrometer methods were employed in this investigation.

Pipette Method. When solid particles of various sizes are dispersed in a liquid medium and then allowed to settle freely, the relative distances that individual particles will settle in a fixed time depends upon their size provided that the particles are of the same density and the temperature remains constant. If a sample is taken after a stated time from a specified level, the concentration of the solid particles in the sample will be less than the original concentration by the weight of the larger particles that have settled out of the test zone. The maximum particle size present in the sample can be calculated with the aid of the Stokes formula and, if several samples are taken at suitable time intervals, data may be obtained for plotting a particle size distribution curve for the material undergoing test.

Regular Pipette Method. This method was used by Jackson and Saeger[3] and has been described in detail in their paper.

Test Procedure

Experimental Details. A 50-gram sample of oven-dried sand was placed in a 1-qt. mason jar with 25 ml of one per cent sodium hydroxide solution and 475 ml distilled water and dispersed for 5 min. with an electric stirrer equipped with vertical baffles. The mixture was then transferred into a 1-liter cylinder by repeated washings with distilled water. Additional distilled water was added to bring the volume of the suspension up to 1 liter. The mixture was shaken thoroughly for 1 min. and the cylinder placed on a level surface to allow uniform undisturbed settling. A stop watch was immediately started and samples were withdrawn at intervals of 2, 5, 15, 30, 60, 250, and 1440 min. The tip of the pipette was inserted in the mixture to a settling depth "L" of 5 in. (12.7 cm). After the sample had been withdrawn, the pipette was rinsed with distilled water and the rinsings added to the sample in a weighed evaporating dish. The temperature of the mixture was determined at the time that each sample was withdrawn.

Correction Factors

The samples were evaporated to dryness in an oven at 105 C, cooled in a desiccator, and weighed for the determination of solids. The weight of each dried sample was corrected for the amount of NaOH which was present in the sample. This correction could not be based on the calculated concentration of NaOH in the sample but had to be determined experimentally. It was found to be —0.0095 grams for each sample. The percentage of particles remaining in suspension at the time the sample was taken was determined from the simple formula:

$$P = \frac{W_1 - 0.0095}{W} \times 100$$

where

W_1 = weight of dried sample.

W = weight of sand originally dispersed in 25 ml. Since 50 grams were dispersed in 1000 ml and a 25-ml sample was withdrawn,

$$W = \frac{50 \times 25}{1000} = 1.25 \text{ grams.}$$

The maximum particle size present in each sample was calculated from Stokes' law by substituting in the correct values for time, temperature, settling depth, specific gravity of the sand, etc. This computation was simplified considerably by the use of a nomographic chart such as that used by Jackson and Saeger. This nomograph was constructed for use with a sand having a specific gravity of 2.650. Since the specific gravities of the sands used were slightly higher than this value (Albany, 2.698; Lumberton, 2.676), a correction was made for this difference. This correction was calculated from the simple relationship derived from the Stokes formula:

$$\text{Specific gravity correction} = \sqrt{\frac{1.65}{\text{sp gr} - 1}}$$

For each of the sands used the correction factor amounted to 0.99. This factor multiplied by the diameter obtained from the nomograph gives the true value of "d." Obviously this correction is quite small and need not be made except when extreme accuracy is required.

After the last sample was withdrawn the mixture remaining in the cylinder was wet-screened on a No. 270 sieve under a light stream of tap water for 5 min. and then oven-dried on the same screen. The dried sample was removed from the screen, weighed, sieved, and the weights of the fractions collected on each sieve were determined in the usual way.

The complete distribution of particle sizes as determined both by the sieving test and by the pipette determination is represented most conveniently by a cumulative curve.

Table 2

RESULTS OF SIEVE ANALYSIS AFTER COMPLETING PIPETTE TEST ON LUMBERTON SAND (TABLE 1)

Sieve No.	Retained on Sieve — grams	Retained on Sieve — %	Accumulative %
12	0.24	0.5	0.5
20	2.31	4.6	5.1
30	2.57	5.1	10.2
40	4.30	8.6	18.8
50	5.60	11.2	30.0
70	8.24	16.5	46.5
100	8.01	16.0	62.5
140	3.96	5.9	68.4
200	2.66	5.3	73.7
270	0.66	1.3	75.0
Pan	0.16	0.3	
	39.23		

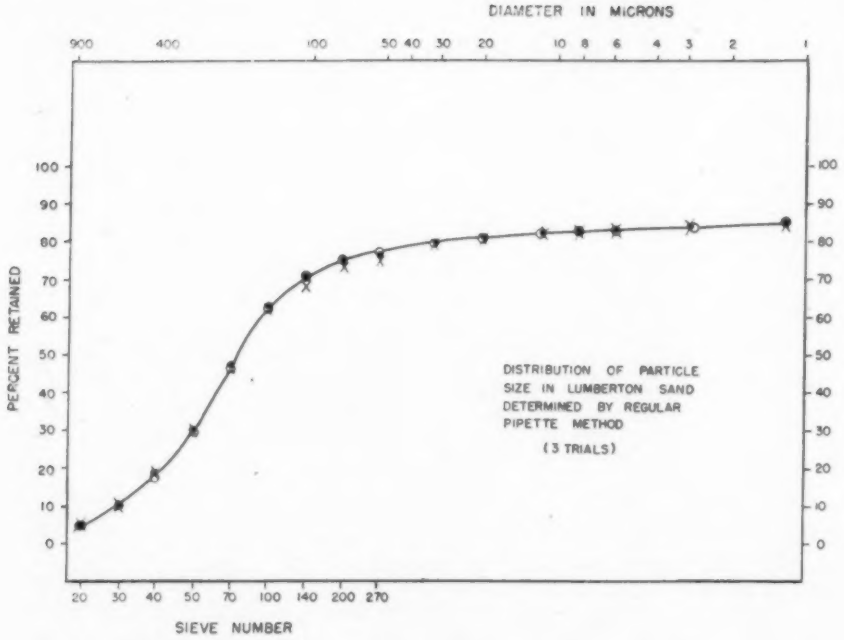


Fig. 1—Results of fineness tests on Lumberton molding sand by the regular pipette method.

The accumulated amount (expressed in percentages) that would be retained on each successively finer sieve is plotted on a semilogarithmic graph paper against the diameter of the sieve opening. The percentage in each case refers to the particles *coarser* than the specified size. In the subsieve range the data obtained by the pipette method specify the percentage (P) of the test sample representing particles *finer* than a specified diameter. *Therefore, in each instance the value, 100%—P, represents the fraction of particles coarser than the specified size.*

Results Obtained. Three determinations were made on both the Albany and the Lumberton sands by the regular pipette method. The data were tabulated and cumulative curves plotted. A sample data sheet showing the results obtained in one of these tests is presented in Tables 1 and 2. The cumulative curves related to the regular pipette tests are shown in Figs. 1 and 2.

Andreasen Pipette Method. The Andreasen apparatus (Fig. 3) consists of a glass cylinder provided with a ground glass stopper through which passes the stem of a pipette. The cylinder has a capacity of approximately 643 ml when filled to the upper mark on the scale. The pipette extends 20 cm beneath the upper scale mark and ends at the level of the zero mark (4 cm from the bottom). The pipette has a capacity of approximately 10 ml, and is provided with a three-way stopcock and spout. The volume delivered by the Andreasen pipette bulb used in these tests was found to be 10.02 ml. The Andreasen pipette method is theoretically more accurate than the regular pipette method since the pipette remains in the cylinder throughout the test and the suspension is not disturbed.

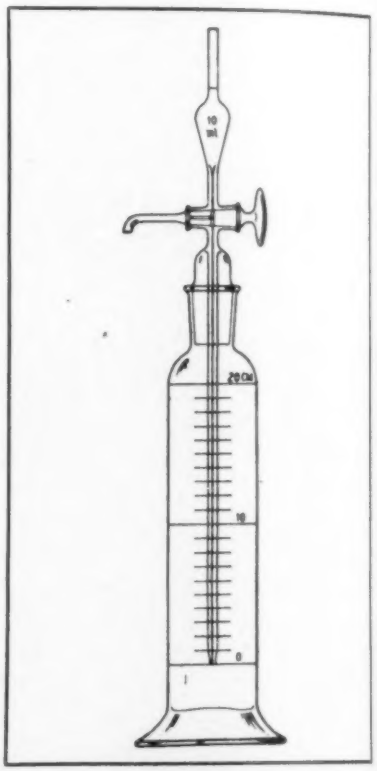


Fig. 3—Sketch of Andreasen pipette.

Experimental Details. A 25-gram

Fig. 2—Results of fineness tests on Albany molding sand by the regular pipette method.

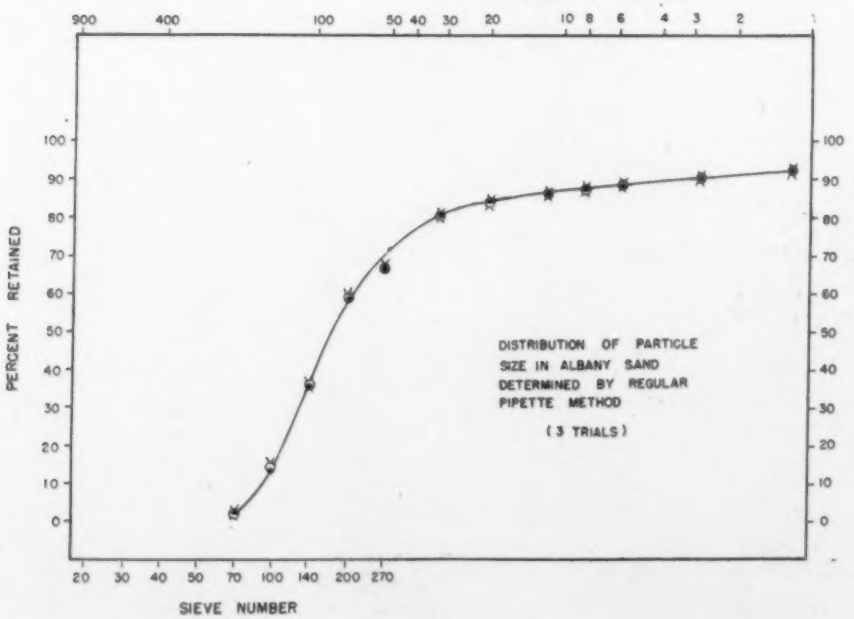


Table 3

RESULTS OF ANDREASEN PIPETTE TEST ON LUMBERTON SAND

Time, min	*L, cm*	*d, microns*	*W_1, grams*	*$W_1-0.0031$, grams*	*P, %*	*100−P, %*
2	20.00	43.1	0.0861	0.0830	21.2	78.8
5	19.65	27.0	0.0808	0.0777	20.0	80.0
15	19.30	15.5	0.0760	0.0759	19.5	80.5
30	18.95	10.8	0.0726	0.0693	17.9	82.1
60	18.60	7.6	0.0712	0.0681	17.5	82.5
250	18.25	3.7	0.0681	0.0650	16.7	83.3
1440	17.90	1.5	0.0644	0.0613	15.8	84.2

sample of oven-dried sand was placed in a mason jar with 300 ml of distilled water and 15 ml of 1% sodium hydroxide solution. After 5 min. of agitation the mixture was transferred carefully into the Andreasen cylinder and additional water was added up to the 20-cm mark.

The cylinder was placed in a constant-temperature bath and allowed to remain there until the mixture reached the required temperature (67 F). The cylinder was then removed from the water bath and, with the stopper in place, the mixture was shaken for 1 min. Immediately after this operation the cylinder was returned to the constant-temperature bath and a stop watch started. Samples were then withdrawn at specified time intervals and placed in an evaporating dish just as in the regular pipette method. The bulb of the pipette was rinsed with distilled water and these rinsings added to the sample in the evaporating dish. Each sample was then evaporated to dryness, cooled, and weighed. The weight of each dried sample was corrected for the amount of sodium hydroxide present in the sample (determined experimentally to be —0.0031 grams).

The percentage of material still in suspension was obtained by considering the weight of solids in the sample as compared with the amount present in a similar volume immediately after the sand was dispersed. Since 25 grams of sand were dispersed in a volume of 643 ml, the amount of sand in 10.02 ml was

$$\frac{25}{643}\times 10.02 = 0.389 \text{ gram.}$$

The percentage of material in suspension was obtained from the formula

$$P = \frac{W_1 - 0.0031}{W} \times 100,$$

as was shown in the discussion of the regular pipette method.

The maximum diameter of the particles remaining in suspension at the time of taking the sample was calculated from Stokes' formula, *consideration being given to the fact that the level of the suspension changes after each withdrawal.*

The mixture remaining in the pipette cylinder after withdrawing the last (1440 min.) sample was washed on a No. 270 sieve for 5 min., dried, and sieved in the usual manner. The cumulative curve was then plotted as described previously.

Results Obtained. A sample data sheet showing the results obtained in one of the Andreasen pipette tests is given in Tables 3 and 4 and the cumulative curves are presented on Figs. 4 and 5.

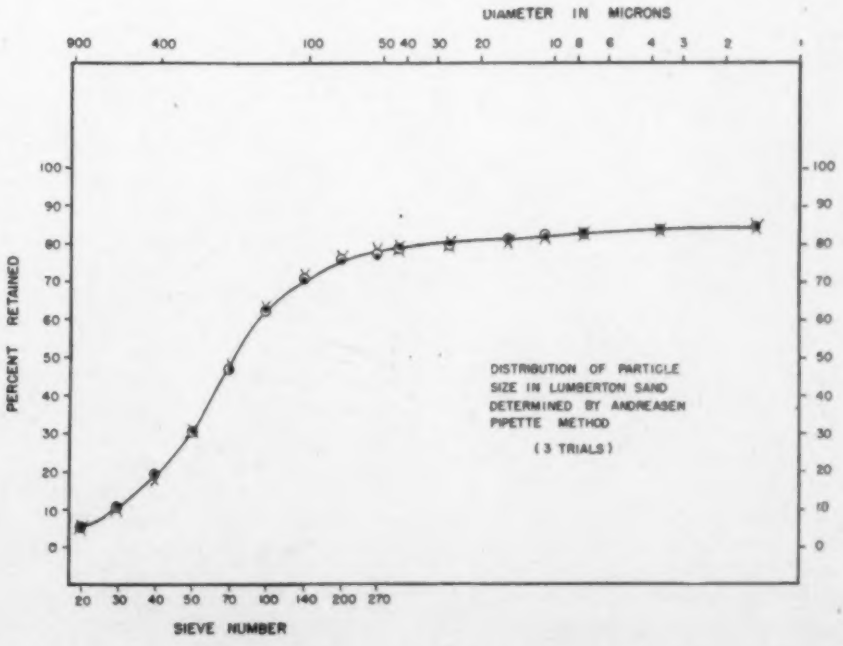
Hydrometer Method. The hydrometer method was introduced as a means of making fineness determination of soils by Bouyoucos[6] in 1927. The procedure is relatively simple and consists of making hydrometer readings on a suspension of the sand in question at predetermined intervals of time. Both the settling depth and the percentage of material in suspension can be obtained from the hydrometer reading. It should be pointed out that in a suspension containing particles of various sizes, a density gradient will develop during settling. Consequently, a hydrometer which measures the specific gravity of the suspension will give a reading which represents the

Table 4

RESULTS OF SIEVE ANALYSIS AFTER COMPLETING ANDREASEN PIPETTE TEST ON LUMBERTON SAND

Sieve No.	*—Retained on Sieve— grams*	*%*	*Accumulative %*
12	0.10	0.4	0.4
20	1.24	5.0	5.4
30	1.27	5.1	10.5
40	1.94	7.8	18.3
50	3.16	12.6	30.9
70	4.28	17.1	48.0
100	3.83	15.3	63.3
140	2.00	8.3	71.6
200	1.18	4.7	76.3
270	0.51	2.0	78.3
Pan	0.14	0.6	
	19.67		

Fig. 4—Results of fineness tests on Lumberton molding sand by the Andreasen pipette method.



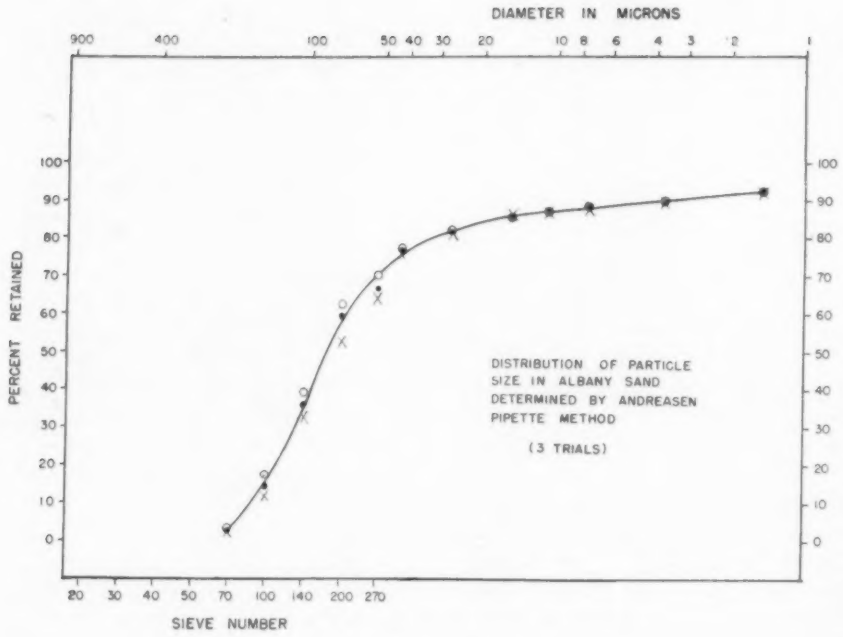


Fig. 5—Results of fineness tests on Albany molding sand by the Andreasen pipette method.

average specific gravity and, therefore, the average composition of the vertical zone occupied by the bulb. The settling depth depends on the distance from the surface of the suspension to the center of buoyancy of the hydrometer, and this distance changes during the course of a determination. The method of calibrating the hydrometer for evaluating the settling depths for different scale readings will be discussed later.

The standard equipment consists of a hydrometer and a glass cylinder of 18-in. height, 2.34-in. (5.85-cm) diameter, and graduated for a volume of 1000 ml.

Two types of hydrometers are available for this work; (1) a gram concentration hydrometer, graduated in grams of soil or sand per liter, and (2) a specific gravity hydrometer. In the present work all of the hydrometer tests were made with the gram concentration hydrometer.

Experimental Details. A 50-gram sample of oven-dried sand was placed in a 1-qt. mason jar with 25 ml of 3% sodium hydroxide solution and 475 ml distilled water and dispersed for 5 min. with an electric stirrer equipped with vertical baffles. The mixture was then transferred carefully into the liter cylinder. Additional distilled water was added until the level of the mixture reached the liter mark on the cylinder. The cylinder was placed in a constant-temperature bath and allowed to remain there until the mixture reached the required temperature (67 F). The cylinder was then removed from the water bath and, with the top closed by the palm of the operator's hand, the mixture shaken thoroughly for 1 min. by turning the cylinder end over end. Immediately after this operation the cylinder was returned to the constant-temperature bath and a stop watch started. Successive hydrometer readings were taken at the top of the meniscus at the same time intervals as in the pipette tests.

After each reading the hydrometer was removed from the mixture, cleaned, and placed in a distilled water container in the constant-temperature bath until the time for the next reading. This removal was necessary to permit free settling of the suspension and to prevent adherence of the particles to the hydrometer.

Determination of the Maximum Diameter of Sand Particles in the Sampling Zone. The computation of results involves two separate operations: (a) the determination of the maximum size of the particles in suspension, and (b) the determination of the percentage of dispersed particles remaining in suspension at a given time.

The settling depth in the hydrometer determination depends upon the dimensions of the hydrometer and upon the level at which it comes to rest in a given suspension.

The effective depth of immersion of the hydrometer at any sedimentation time was defined by Schuh-

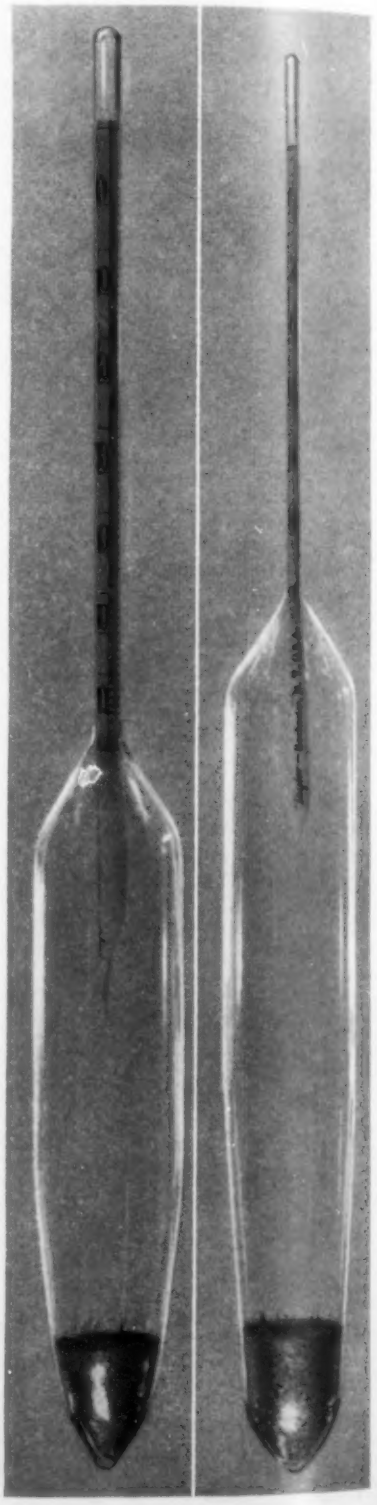


Fig. 6—Photographs showing gram concentration hydrometers.

mann[7] as the depth of the center of volume of the hydrometer bulb, as it would be measured from the liquid surface with the hydrometer absent. Thus, the evaluation of this depth for any hydrometer reading requires data as to volume and dimensions of the hydrometer and the cross-sectional area of the container used. Computation of the effective depth of settling is given by Klein[8] and Schuhmann[7] as

$$H = H_1 + \tfrac{1}{2}\left(h - \frac{V}{A}\right)$$

where

H = effective depth of immersion,

H_1 = distance between the suspension level and the top of the hydrometer bulb,

h = length of the hydrometer bulb,

V = volume of the hydrometer bulb,

A = cross-sectional area of container.

The term ($-\frac{1}{2}$ V over A) is a correction for the change in the distance from the surface of the suspension to the center of the sampling zone when the hydrometer is removed from the suspension. The distance from the surface of the suspension to the center of volume is a true settling depth only when the hydrometer is in suspension. It may be seen that the hydrometer bulb is here assumed to be of a symmetrical shape and the center of its volume is taken at the midpoint of the bulb. Actually the hydrometer bulbs are seldom symmetrical (Fig. 6) and, therefore, the center of volume must be determined experimentally.

Center of Volume

The following method[9] for locating the center of volume was used for this purpose in the present investigation. A graduated cylinder was filled with water and the volume of water recorded. The hydrometer was then immersed in steps (usually 2 cm) and the water level at each step recorded. The volume readings were plotted against corresponding depth of immersion readings and a "depth of immersion versus volume" curve drawn. The center of volume of a given hydrometer was determined as a point on this curve, corresponding to one-half of the total volume of water displaced by the hydrometer.

The method for locating the center of volume of one of the hydrometers is shown by a curve (Fig. 7).

If we let "l" represent the distance from the surface of the suspension to the center of volume as determined in this manner, and "L" the effective depth of settling, the following relationship will hold:

$$L = l - \frac{V}{2A}$$

where "V" is the volume of the hydrometer bulb and "A" is the cross-sectional area of the container.

In the example (Fig. 7) discussed in the foregoing, V = 89 cc (for the particular hydrometer used) and A = 27.03 cm². Hence:

$$L = l - 1.65 \text{ cm}$$

This correction of −1.65 cm for the particular hydrometer used is incorporated into the scale at the right of Fig. 7 so that values for "L" may be obtained directly from the hydrometer readings. For example, for the hydrometer reading of 2.6 "L" is 12.0 cm; for a reading of 9 the "L" distance is 8.0 cm, etc.

Grain Diameter

Knowing the settling depth "L" the maximum grain diameter can be calculated according to Stokes' formula given in the foregoing.

In order to simplify the computation of the maximum diameter of the particles present at a distance "L" from the surface, the nomograph shown in Fig. 8 may be employed. This nomograph furnishes a convenient means for determining

Fig. 7—Determination of settling depth "L" of sand particles in the hydrometer test.

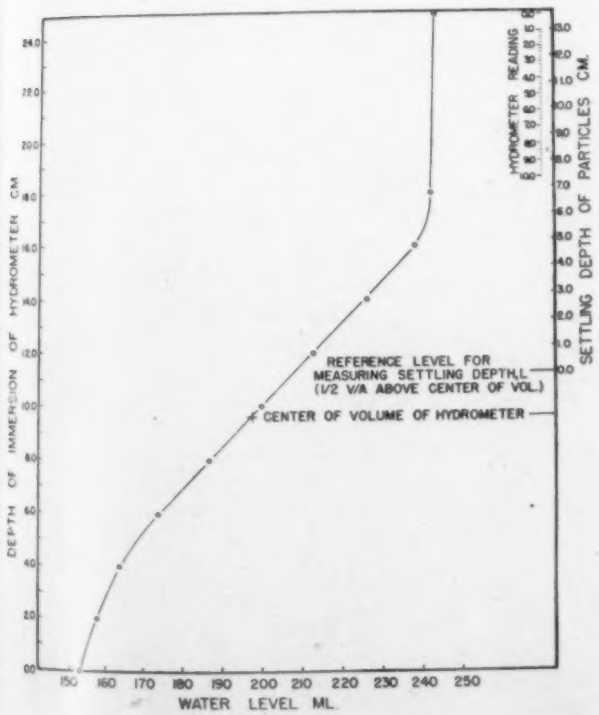


Fig. 8—Nomograph showing relationship between settling depth, time of settling and particle diameter of sand with reference to hydrometer test.

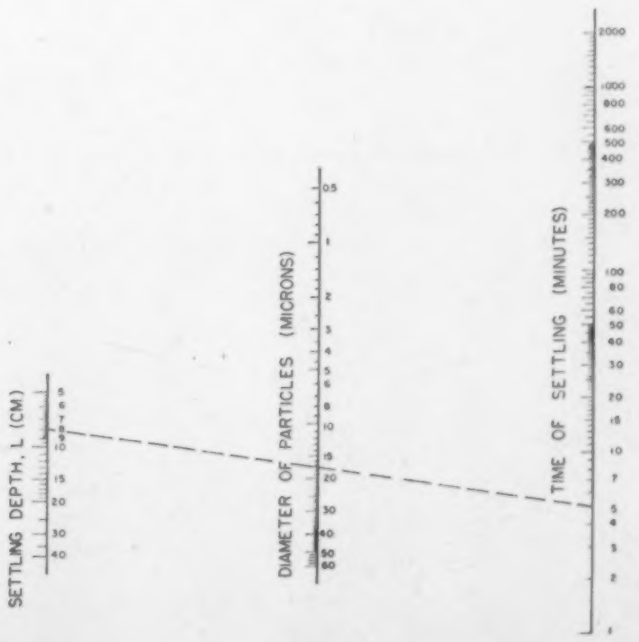


Table 5

VALUES[10] OF CONSTANT "A"

Specific Gravity of Sand "G"	*Constant "a"*
2.95	0.94
2.85	0.96
2.75	0.98
2.65	1.00
2.55	1.02
2.45	1.05
2.35	1.08

Table 6

RESULTS OF HYDROMETER TEST ON LUMBERTON SAND

Time, min	*Reading*	*Scale Correction*	*Electrolyte Correction*	*Reading Corrected*	*L, cm*	*Diameter "d" in microns — Read from nomograph*	*Diameter "d" in microns — Corrected According to sp gr of sand*	*P, %*	*100—P, %*
2	11.6	−0.9	−1.4	9.3	6.4	25.4	25.1	18.4	81.6
5	11.1	−1.0	−1.4	8.7	6.7	16.3	16.1	17.2	82.8
15	10.6	−1.0	−1.4	8.2	7.0	9.4	9.3	16.3	83.7
30	10.4	−1.0	−1.4	8.0	7.1	6.7	6.6	15.9	84.1
60	10.0	−1.0	−1.4	7.6	7.4	4.8	4.8	15.0	85.0
250	9.7	−1.0	−1.4	7.3	7.6	2.3	2.3	14.5	85.5
1440	9.4	−1.0	−1.4	7.0	7.8	1.0	1.0	13.9	86.1

the interrelation of time of settling, "T" (min.), distance, "L" (cm), and particle diameter "d" (microns). A straight line across the three branches of the nomograph indicates the foregoing interrelation; for example, the broken line in Fig. 8 shows that after 5 min. at a settling depth of 8 cm, the suspension at that depth will be free from all particles larger than 17 microns in maximum diameter.

It should be noted that in constructing this nomograph it was assumed that the test was carried out at the constant temperature of 67 F, and that the specific gravity of the sand was 2.65. Since the specific gravities of the sands used were somewhat different from this value, the maximum diameters "d" should be corrected, as explained previously (with reference to the regular pipette test).

Determination of the Percentage of Dispersed Sand Particles Remaining in Suspension. For the gram concentration hydrometer the percentages of dispersed sand in suspensions corresponding to different hydrometer readings are calculated[10] according to the following formula:

$$P = \frac{(R \pm \Delta R)a}{W} \times 100$$

where

P = percentage of originally dispersed sand remaining in suspension at a distance "L" below the surface,

R = corrected hydrometer reading. (Before substituting "R" into this formula, it was necessary to correct for the sodium hydroxide that was used as deflocculating agent for the suspension. This correction was calculated and also checked experimentally and found to be −1.4 when 25 ml of 3% sodium hydroxide was present in 1 liter.)

ΔR = temperature correction for hydrometer reading. (For a temperature of 67 F this correction is zero.)

W = weight in grams of oven-dried sand originally dispersed (usually 50 grams),

a = constant depending on the specific gravity of the sand particles.

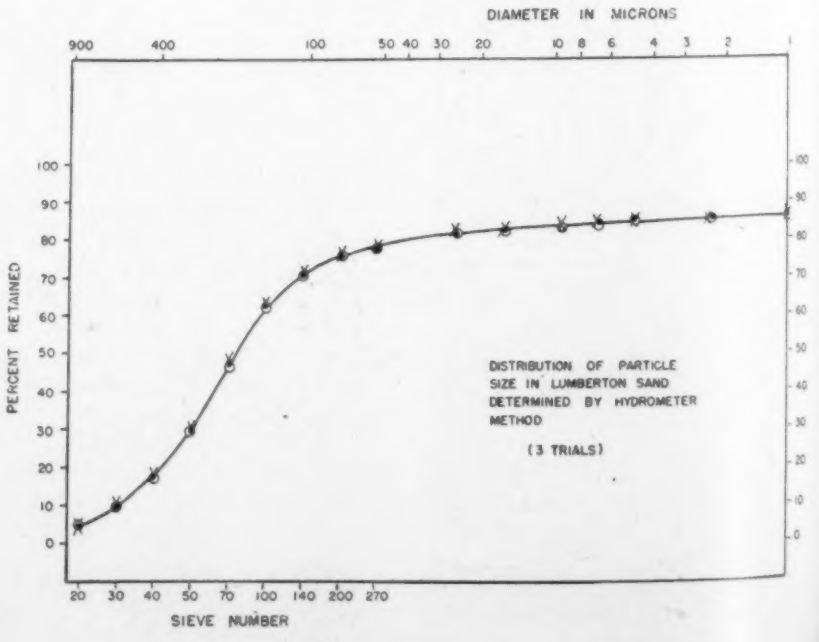
The values "a" when referred to different values of the specific gravity of soil or sand particles "G" are given in Table 5[10]. It is sufficiently accurate for ordinary sand tests to select a value for "a" for the specific gravity nearest to that of the par-

Table 7

RESULTS OF SIEVE ANALYSIS AFTER COMPLETING HYDROMETER TEST ON LUMBERTON SAND

Sieve No.	*Retained on Sieve — grams*	*%*	*Accumulative %*
12	0.23	0.5	0.5
20	2.14	4.3	4.8
30	2.57	5.1	9.9
40	3.78	7.6	17.5
50	6.06	12.1	29.6
70	8.77	17.5	47.1
100	7.79	15.6	62.7
140	4.17	8.3	71.0
200	2.41	4.8	75.8
270	.99	2.0	77.8
Pan	.47	0.9	
	39.46		

Fig. 9—Results of fineness tests on Lumberton molding sand by the hydrometer method.



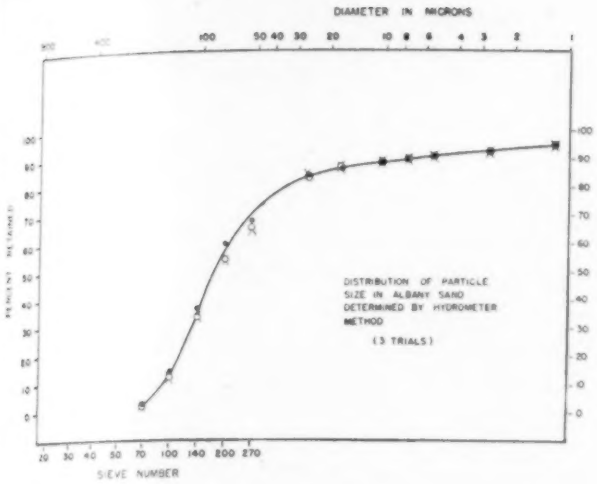


Fig. 10—Results of fineness tests on Albany molding sand by the hydrometer method.

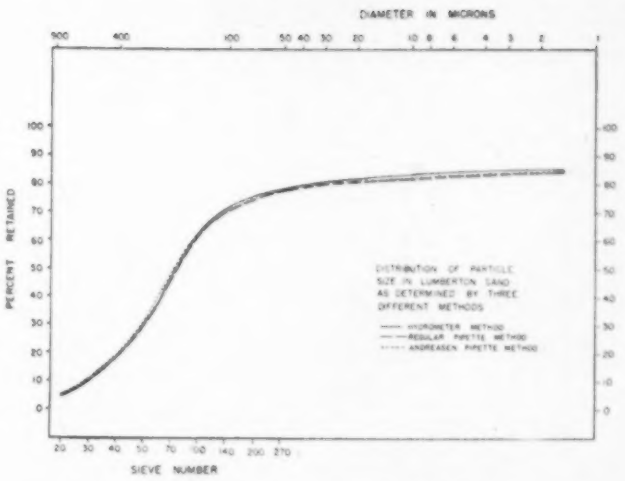


Fig. 11—Cumulative curves showing results of fineness tests on Lumberton molding sand by three methods.

ticular sand tested. For both the Albany and Lumberton sands this value was 0.99. Therefore,

$$P = \frac{0.99R}{50} \times 100 = 1.98R$$

where

P = percentage of originally dispersed sand remaining in suspension at a distance "L" below the surface,

R = corrected hydrometer reading.

Results Obtained. Three tests were run on each of the two molding sands. Cumulative curves for these tests are shown in Figs. 9 and 10. A sample data sheet giving in detail the results obtained in one of the hydrometer tests is presented in Tables 6 and 7.

The scale of the hydrometer used in these tests was calibrated and a suitable correction was made (see Appendix).

Summary and Conclusion

Examination of the results obtained on each sand in three determinations by the same method (Figs. 1, 2, 4, 5, 9 and 10) indicates a rather high degree of precision. For ordinary purposes, therefore, a single determination by a reasonably careful operator would be sufficient to establish the distribution of particle size for a given sand in the subsieve range.

In order to compare the fineness test results obtained by the three different methods used in this study, charts were prepared showing the average results by each method for both Albany and Lumberton sands.

These graphs (Figs. 11 and 12) indicate that there was a good agreement between the average cumulative curves representing the different methods of analysis.

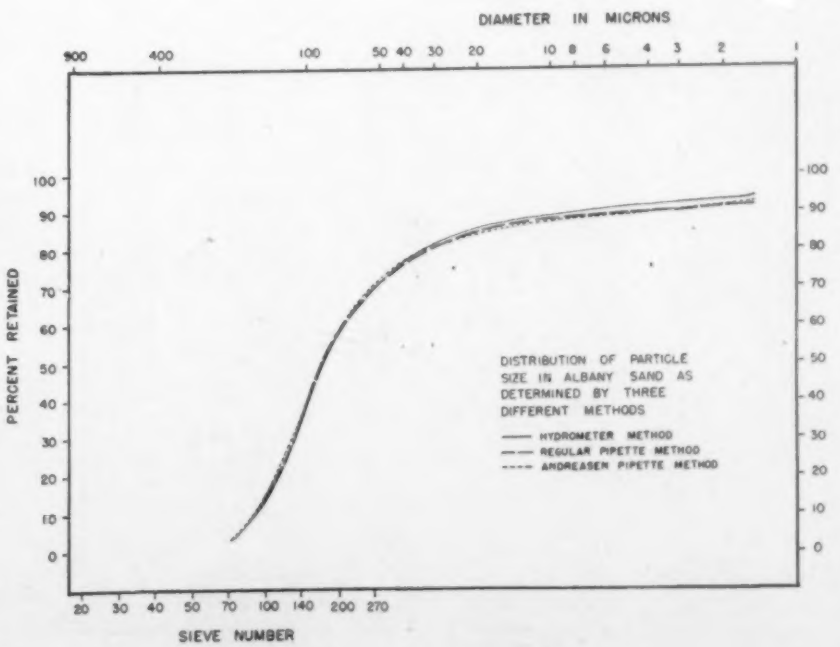
The authors believe, therefore, that any of the methods employed is satisfactory for this purpose. In selecting a method to be used, one should consider the laboratory skill and technique required for each procedure, the time involved for testing, and equipment available.

The hydrometer method seems to be the most convenient since it involves simply taking a hydrometer reading and then referring to simple charts for obtaining the values for the limiting particle size according to Stokes' law, while the percentage of material still in suspension is obtained from a simple calculation.

The pipette tests are more time consuming chiefly because of the time required in evaporating, drying, cooling, and weighing. However, in the regular pipette test no special equipment is required other than that in the chemical laboratory.

The Andreasen pipette method,

Fig. 12—Cumulative curves showing results of fineness tests on Albany molding sand by three methods.



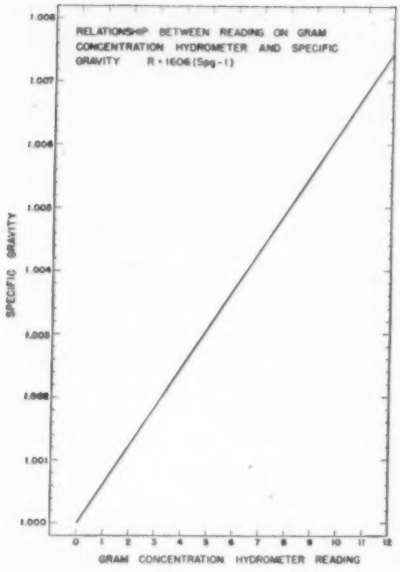


Fig. 13—Graph for converting specific gravity values to equivalent readings on the gram concentration scale.

which involves special apparatus, is theoretically more accurate than the regular pipette method since the stem remains in the cylinder throughout the test and the suspension is not disturbed. The fact that the stem of the pipette is stationary, however, might appear to introduce an error, since some liquid remains in the stem after each sampling and is always drawn up into the bulb as a part of the next sample. The possible error due to this source would, of course, be quite small. This was demonstrated in a special experiment in which the determination was carried out in the usual manner except that two samples instead of one were withdrawn after each settling period. The results were in good agreement indicating that no appreciable error was introduced from this source.

Experience in the authors' laboratory with the different methods of determining distribution of particle size in the clay fractions of foundry sands indicates that the hydrometer method is preferable because it yields satisfactory results conveniently in a minimum of time.

Appendix

Calibration of Hydrometer Scale. The scale of each hydrometer should be calibrated for precise work. Essentially this involves the testing of the hydrometer in solutions of known specific gravities in the range in which the instrument is to be used. The relationship between the "R" readings of the gram concentration hydrometer and the corresponding values for specific gravity may be obtained from the formulas[10] that are used for calculating the percentage of material in suspension from readings on the gram concentration hydrometer

$$P = \frac{Ra}{W} \times 100$$

and on the specific gravity hydrometer

$$P = \frac{1606\ (Sp\ g - 1)a}{W} \times 100$$

Thus $R = 1606\ (sp\ gr - 1)$. This relationship between "R" and specific gravity is shown for a range of values by the graph in Fig. 13.

The zero point of the gram concentration hydrometer may be checked by making a reading in boiled distilled water at 67 F, whereas sodium chloride solutions of known concentrations are suitable for checking other points on the scale.

Density values for aqueous solutions of sodium chloride (1, 2, and 4% NaCl) at temperatures of 10, 20, and 25 C obtained from the International Critical Tables are shown in Table 8. From these values, the densities at 67 F (19.4 C) were obtained by interpolation and the corresponding specific gravities calculated (Table 9). The relationship between concentration of sodium chloride and specific gravity at 67 F is shown in Fig. 14.

Three or more sodium chloride solutions of the desired concentrations are carefully prepared, using dried C.P. sodium chloride and boiled distilled water. The exact "R" values (gram concentration readings) at 67 F for each solution are obtained by referring to Figs. 14 and 13, respectively. These solutions are brought to a temperature of 67 F and hydrometer readings made, exercising the same precautions recommended for the regular use of the hydrometer. The actual hydrometer readings are plotted against the true

Table 8

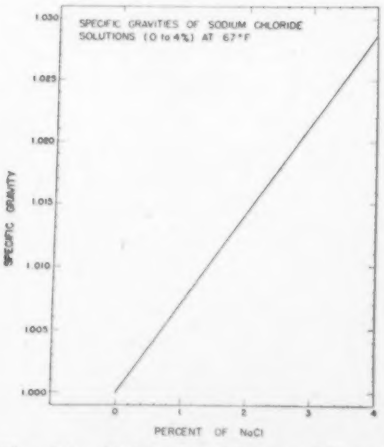
DENSITY VALUES OF SODIUM CHLORIDE SOLUTIONS AT DIFFERENT TEMPERATURES[11]

NaCl Concentration by Weight, %	*Density Values, grams/ml*		
	10 C (50 F)	*20 C (68 F)*	*25 C (77 F)*
1	1.00707	1.00534	1.00409
2	1.01442	1.01246	1.01112
4	1.02920	1.02680	1.02530

Table 9

DENSITY AND SPECIFIC GRAVITY VALUES OF SODIUM CHLORIDE SOLUTIONS AT TEMPERATURE OF 67 F (19.4 C)

NaCl Concentration, %	*Densities, grams/ml*	*Specific Gravities* $\left(sp\ gr = \frac{density}{0.99835}\right)$
0	0.99835	1.00000
1	1.00546	1.00712
2	1.01259	1.01426
4	1.02696	1.02866



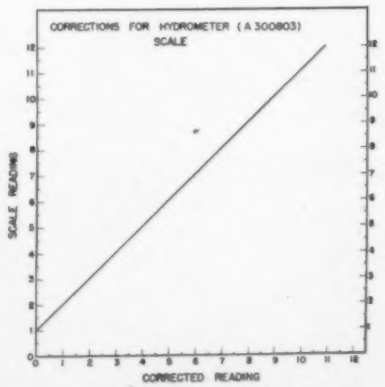


Fig. 14—(left) Specific gravities of aqueous sodium chloride solution at 67 F.

Fig. 15—(right) Graph for obtaining corrected readings from scale readings on gram concentration hydrometer.

or correct readings, as shown in Fig. 15, and the scale correction for any scale reading of the hydrometer may be obtained from this graph.

Acknowledgment

Grateful acknowledgment is made to Dr. Vernon C. F. Holm for suggestions and assistance, and to Mrs. Lura F. Roehl for preparing the charts and graphs.

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DISCUSSION

Chairman: B. H. BOOTH, Carpenter Brothers, Inc., Milwaukee, Wis.

Co-Chairman: R. E. MOREY, Naval Research Laboratory, Washington, D. C.

CHAIRMAN BOOTH: We have tried these hydrometer tests at various times in our laboratory. I would like to ask if you have noticed any difference between the grade of sand, e.g., a heavy grade of sand with high clay content, and a light grade like Albany sand with low clay content? Would there be any difference in the accuracy?

CO-CHAIRMAN MOREY: As long as your technique is correct, you can test the one as accurately as the other. We have tested successfully materials with as much as 40 to 50 per cent of a very strong clay, and others with as low as one or two per cent clay. If the clay goes below that, we do not attempt to make many tests.

WELDING AND SALVAGE of GRAY IRON CASTINGS

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Pontiac Motor Co.
Pontiac, Mich.

IT IS THE AUTHOR'S intention to bring out certain procedures and tried methods of salvaging gray iron castings through welding, brazing, arc, cylinder bore welding, soldering and plugging.

The tests for break load and deflection of welded and brazed specimens proved to be extremely interesting. The simple method used in the first experiment should be valuable in measuring the value of welders and materials used.

Brazing of cast iron, especially finished and semi-finished cylinder blocks, a method being used by the author as a reliable means of repair, will be discussed in detail. The photographs shown in the paper are those of actual repairs which returned the castings to production.

The photomicrographs show just what happens in the welding of cast iron. The discussion on the presence of high phosphorus in the welding rod and the weld itself should clear up some ideas as to its harmful effects from a physical standpoint.

Furnaces. The preheat furnace should be of the continuous or pusher type. It should be long and narrow, the length being sufficient to allow at least 1½ hr. heating time. The temperature at the end where the castings are entered should be as low as possible (about 350° F.). It should have no lighted burners closer than 8 ft. from the entrance. There are two heating zones, the first at about 800° F., and then the final zone or end of the furnace (1000 to 1100° F.).

This need not be an expensive furnace; in fact, one of the furnaces used by the writer consists of three batch-type furnaces placed end to end. The cold end was then added. Two rows of rails were placed through the furnace, allowing the cylinder blocks to slide, and trays were made for smaller castings. Two air pushers are used.

The cold end is the most important part of the furnace. This was added after many unsuccessful attempts at welding castings of complex design, although baffle plates and other means of diverting the heat from the light or heavy sections were used. Double-jacket castings are the most difficult to weld, but with the previously described furnace no difficulty was encountered with castings weighing up to 560 lb.

The furnace may be fired either by oil or gas, equally good results being obtained. The steel rails must be replaced almost every year. However, these may be cast in the foundry in castings of 4-ft. lengths. The rail has male and female ends to interlock for prevention of buckling.

The dimensions are smaller at one end to allow blocks to slide from one to another without binding. As much alloy as possible should be used to prevent corrosion and scale. These rails will give added wear and may be replaced in sections as needed. One of these rail castings is shown in Fig. 1.

Postheating Furnace. This furnace should be as long as possible. The lengths are not mentioned because they depend upon the production expected; however, the postheating furnace should be long enough to allow the castings to come out at temperatures of not more than 700° F., and much lower if possible. This depends entirely upon the type of castings being welded. A single line of rails will be sufficient with one pusher. The temperature at the front end of this furnace, for a distance of 6-8 ft., should be 1000 to 1100° F.

The best results are obtained if the welders load the furnace and push the trays or cylinder blocks as needed. This gives them a chance to remark the defect and get a good look at it while cold, which has several advantages. It gives the welders full responsibility for the casting from the time it is placed in preheat until it is welded and placed in postheat.

Castings are taken directly from the hot zone and placed on the welding bench with the aid of a hoist. Welding should start immediately. Slow welders should not be placed on this work because the time spent in welding and in getting castings into the postheating furnace is very important. For jacket welds the time

How good is a weld? Is it as good as the parent metal? Methods of gray iron casting repair and salvage have been developed to the point of providing satisfactory answers to these questions.

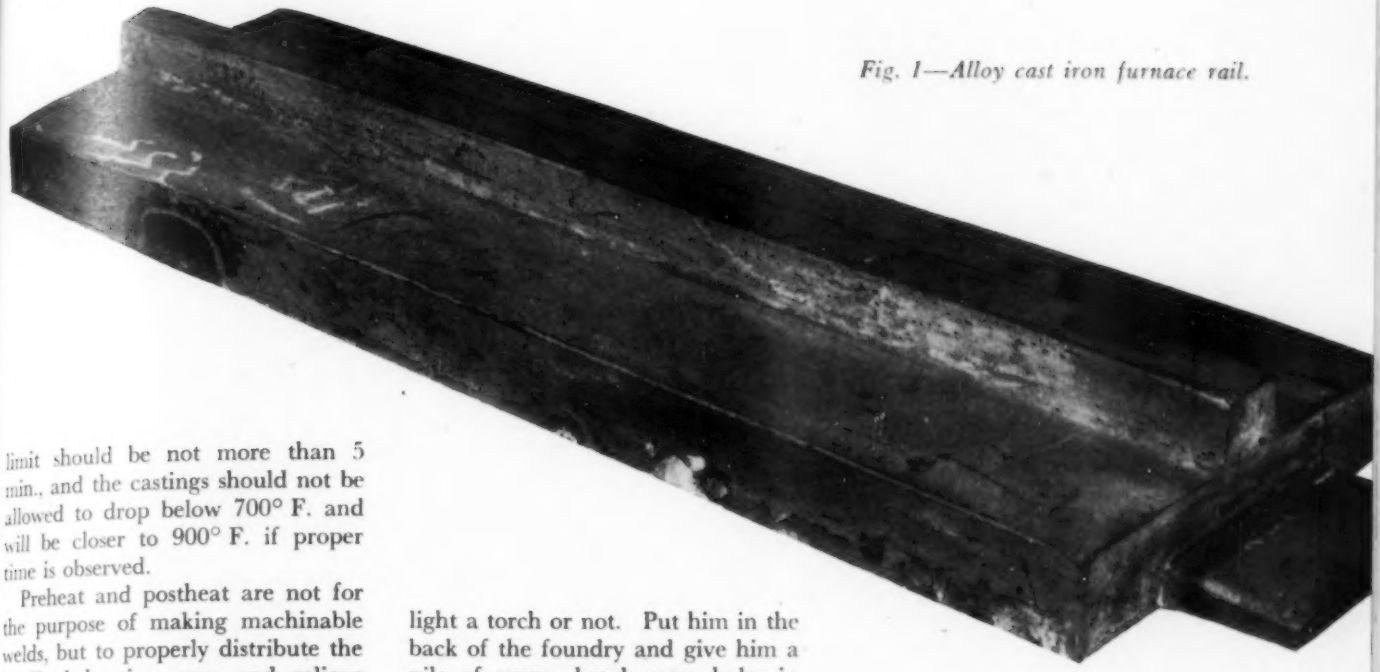


Fig. 1—Alloy cast iron furnace rail.

limit should be not more than 5 min., and the castings should not be allowed to drop below 700° F. and will be closer to 900° F. if proper time is observed.

Preheat and postheat are not for the purpose of making machinable welds, but to properly distribute the localized heating area and relieve stresses which may result. Making welds machinable is entirely the welder's responsibility except in castings so designed that the welder must work straight down into heavy sections such as tappet bosses in cylinder blocks. In such places the welder is unable to properly fuse the metal and welding rod, much less float off slag and impurities.

Such cases should be studied and other means of salvage used. It is not necessary to postheat all castings, or even preheat them. This depends entirely upon design; however, all jackets to be welded should be preheated and postheated. All jacket welds should be burned through and thoroughly cleaned with end of rod before welding.

This method of cleaning is entirely satisfactory except for bad scabs or buckles. A welder should never be allowed to plaster over a defect. He should follow these general rules:

(1) Use a good commercial flux.
(2) For heavy castings use No. 8 or No. 10 welding tip.
(3) Always use the same torch.
(4) Use rod free from chill of the approximate analysis shown in Table 1.

Having proper equipment, next in importance are good welders and a foreman capable of training them. A good method is to select a man who is interested in welding and has been doing a good job at grinding and finishing welds. It makes little difference whether he knows how to light a torch or not. Put him in the back of the foundry and give him a pile of scrap—break some holes in small castings and show him how they should be welded.

He should be kept at this for several hours a day, the rest of the time working on his own job. If after two weeks of welding, file testing and grinding this scrap he still shows a desire to become a welder, and if the foreman is satisfied that he will make a good welder, he should be given a chance. Place him on the job with a good welder or two working the kind of castings on which he received his training. Let him pick the easier welds for 30 days.

After this training period he should be able to take the work as it comes. However, it will be some time before he is ready for cylinder blocks and other jacketed castings. Welders trained in this manner become good welders and are always on the job. Experienced welders employed should go through at least a part of this training, not only to assure the foreman that they are capable but also to familiarize themselves with the castings being produced.

Welders trained in this manner have been checked in one foundry, and absenteeism for all causes, including sickness, vacations, etc., was less than 2 per cent for the year of 1944. Welding errors were less than 0.1 per cent.

Value of Good Welding. In recent meetings of engineers and foundrymen these questions have been asked: How good is a weld? Is it as good as the parent metal? If not, why not? All these questions and many more prompted the author to make the following tests.

A 3/4-in. hole was drilled in a casting. The casting was then preheated, welded, and postheated. Analyses were made of the parent metal, the welding rod (Table 1) and of the welded area. The difference in composition between the parent metal and the welded area is shown in Table 2.

This size of weld, although larger than the average weld, when prop-

Table 1

WELDING ROD ANALYSIS

Component	*Per Cent*
Total carbon	3.47
Silicon	3.07
Manganese	0.55
Sulphur	0.08
Phosphorus	0.61
Copper	None
Nickel	None
Chromium	None

Table 2

EFFECT OF WELDING ON COMPOSITION

	Composition, per cent			
Component	*Parent Metal*	*Welding Rod*	*Weld Area*	*Gain or Loss*
T. C.	3.11	3.47	3.33	0.22+
Silicon	2.59	3.07	2.72	0.13+
Manganese	0.64	0.55	0.55	0.09—
Sulphur	0.18	0.08	0.118	0.062—
Phosphorus	0.14	0.61	0.42	0.28+
Copper	0.5	None	0.13	0.37—
Chromium	0.17	None	0.07	0.10—

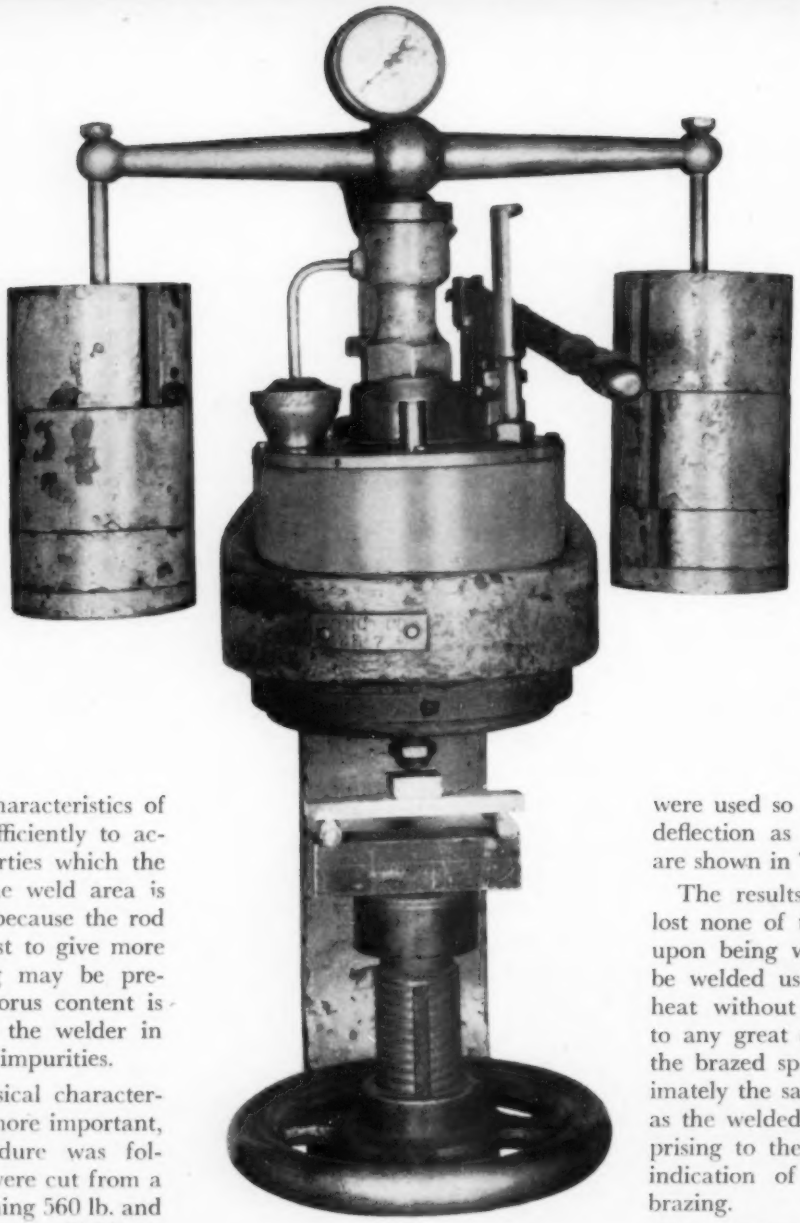


Fig. 2—Apparatus for testing relative values of welds.

erly fused absorbs characteristics of the parent metal sufficiently to acquire physical properties which the rod itself lacks. The weld area is high in phosphorus because the rod itself is purposely cast to give more fluidity that chilling may be prevented. The phosphorus content is also of some aid to the welder in floating off slag and impurities.

To check the physical characteristics which may be more important, the following procedure was followed: Six samples were cut from a cylinder casting weighing 560 lb. and of the analysis shown in Table 3. The specimens were of approximately 5-in. length, 3/4-in. width and 3/16-in. thickness, and are designated as *A, B, C, D, E* and *F*.

Specimens *A* and *B* were broken, preheated, and welded with a rod of the analysis shown in Table 1. Specimens *C* and *D* were preheated and postheated but not welded, *E* was left in the green state, and *F* was cut, edges ground and brazed together. The welds and the braze were as near the center as possible. The testing apparatus used is shown in Fig. 2, and consists of a steel block, two 1/2-in. steel pins, which may be moved to any centers, and a breaking block which is 1⅛-in. wide with corners rounded. This allows the sample to break at its weakest point.

The sample is placed as shown in Fig. 2, pressure is applied and dial noted when sample breaks. An indicator is placed at the breaking block and the deflection noted, which is important in evaluating a weld. The indicator is not shown, because it may be of any type and mounted in any way which will record the deflection or break. Three-in. centers were used so as to register as much deflection as possible. The results are shown in Table 4.

The results show that specimens lost none of their physical qualities upon being welded. Castings may be welded using preheat and postheat without changing the Brinell to any great extent. The fact that the brazed specimen shows approximately the same physical properties as the welded specimens is not surprising to the author: it is a clear indication of the merits of good brazing.

Some users of castings do not like the difference in color, and believe the quality has been depreciated. Various other brazing tests were made and all showed the same comparative results. This does not mean that brazing should replace welding, but it has a place in the salvaging of castings, as will be shown later.

Considering the specimens (Table 4) from a metallurgical point of view, sample *A* was prepared for photomicrographs. Figures 3 and 4 are of the weld area taken at 100 X

Table 3

ANALYSIS OF TEST SPECIMENS

Component	*Per Cent*
Total carbon	3.37
Graphitic carbon	2.70
Combined carbon	0.67
Manganese	0.71
Silicon	2.15
Sulphur	0.127
Phosphorus	0.092
Nickel	0.22
Chromium	0.12
Copper	0.20
Molybdenum	0.07

Table 4

PHYSICAL PROPERTIES OF WELDS

Specimen	*Break Load, lb.*	*Deflection, in.*	*Brinell Hardness No.*
A	375	0.070	182
B	400	0.068	194
C	400	0.078	182
D	370	0.070	194
E	375	0.070	194

and 500 ×, respectively. These show the amount of steadite and its distribution. Figure 5 at 1500 ×, shows the lamellar pearlite which is near the nodular state.

The weld junction shown in Fig. 6 at 500 × reveals the mottled structure of pearlite, pine graphite in rather thick, short flakes, and some free ferrite. Some particles of cementite are present, but most of the white constituent is ferrite.

Covering the weld-affected area, which includes up to 3/4-in. from the weld proper, the photomicrograph (Fig. 7) taken about 1/2-in. from the weld shows free ferrite patches intermixed with graphite. This shows that this area was somewhat in the plastic stage during the process of welding. Little can be done to overcome this and, judging from the checks on physical properties, it has little if any effect. Figure 8 at 100 × shows the parent metal as having a slightly dendritic structure. Figure 9 at 1500 × shows the nature of the pearlite and the amount of free ferrite.

While the simple testing device used in connection with the hardness testing machine (Fig. 2) gave a good comparative check, it was not accurate from a laboratory point of view. To satisfy the more critical, tests were made on a tensile testing machine. The same device was used except that the tensile machine was used so that the break load could be recorded accurately in lb. Another group of samples were prepared from a cylinder block of the analysis shown in Table 5.

These samples were from a 180-lb. cylinder block which had remained in the mold 12 min. Samples *A* and *B* were broken, preheated, postheated and welded; *C* and *D* were preheated and postheated only; *E* left in the green state; and *F* was preheated, cut in two and brazed. Results are shown in Table 6.

From these results it can be stated that a weld, properly executed, is as good as the parent metal. The reason samples of a

Table 5

ANALYSIS OF TEST SPECIMENS

Components	*Per Cent*
Total carbon	3.24
Graphitic carbon	2.74
Combined carbon	0.50
Manganese	0.70
Silicon	2.43
Sulphur	0.142
Phosphorus	0.113
Nickel	0.15
Chromium	0.13
Copper	0.25
Molybdenum	0.04

Table 6

PHYSICAL PROPERTIES OF WELDS

Specimen	*Break Load, lb.*	*Deflection, in.*	*Brinell Hardness No.*
A	645	0.062	205
B	592	0.068	200
C	535	0.060	205
D	537	0.061	200
E	685	0.072	214
F	540	0.068	200

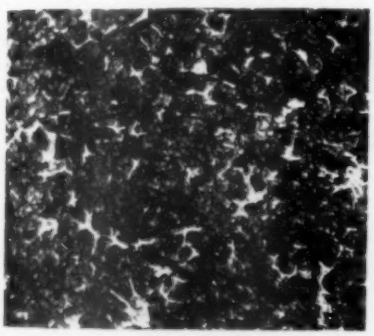


Fig. 3—Photomicrograph of weld area. 100X.

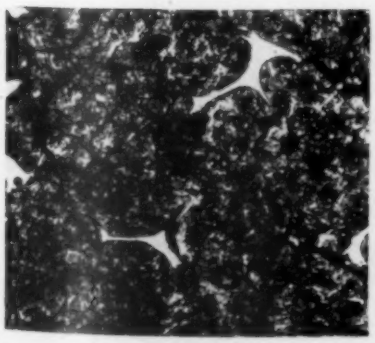


Fig. 4—Same area shown in Fig. 3 except 500X.

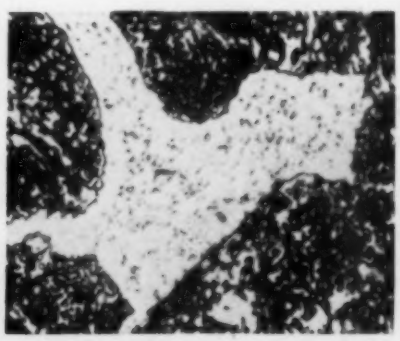


Fig. 5—Lamellar pearlite shown at 1500X.

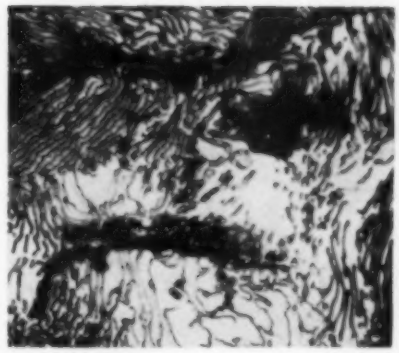


Fig. 9—Same as Fig. 8 except 1500X.

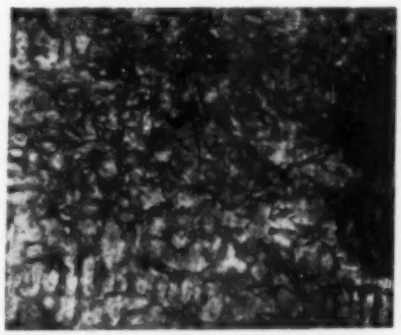


Fig. 8—Photomicrograph showing parent metal to have slightly dendritic structure. 100X.

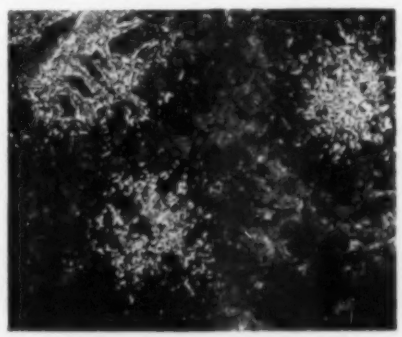


Fig. 7—Area ½ in. from weld shows free ferrite and graphite. 100X.

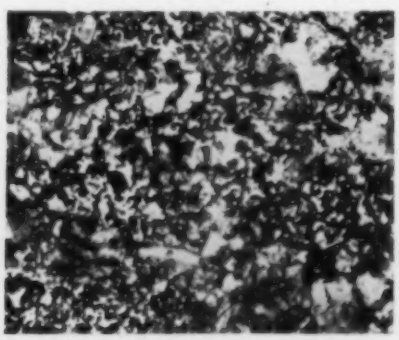


Fig. 6—Weld junction. Pearlite, graphite, and some free ferrite. 500X.

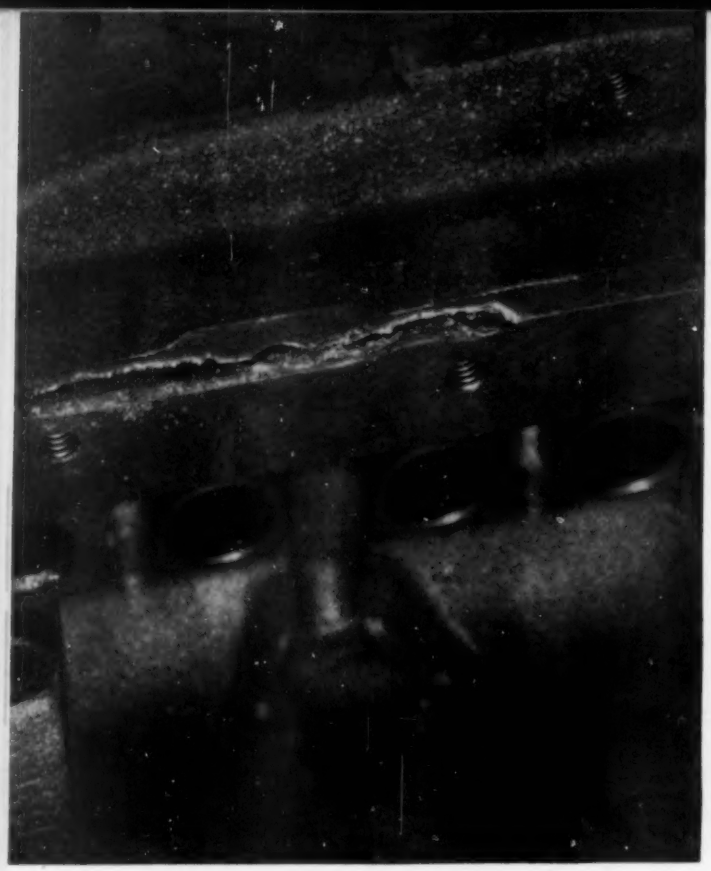


Fig. 10—Cylinder block casting showing blow at cope pan rail.



Fig. 11—Cylinder block casting shown in Fig. 10 after repair by brazing.

Fig. 12—Cylinder block casting rejected because of shy locating pads.

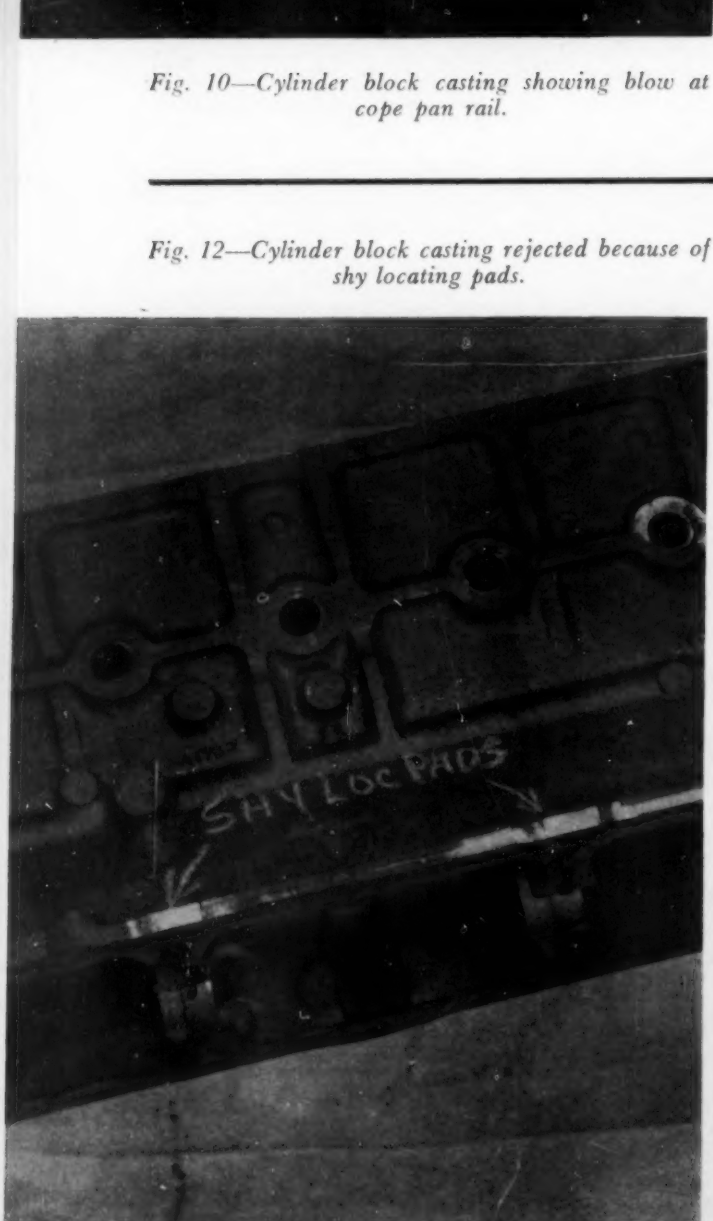


Fig. 13—Same casting shown in Fig. 12 with brazed locating pads.

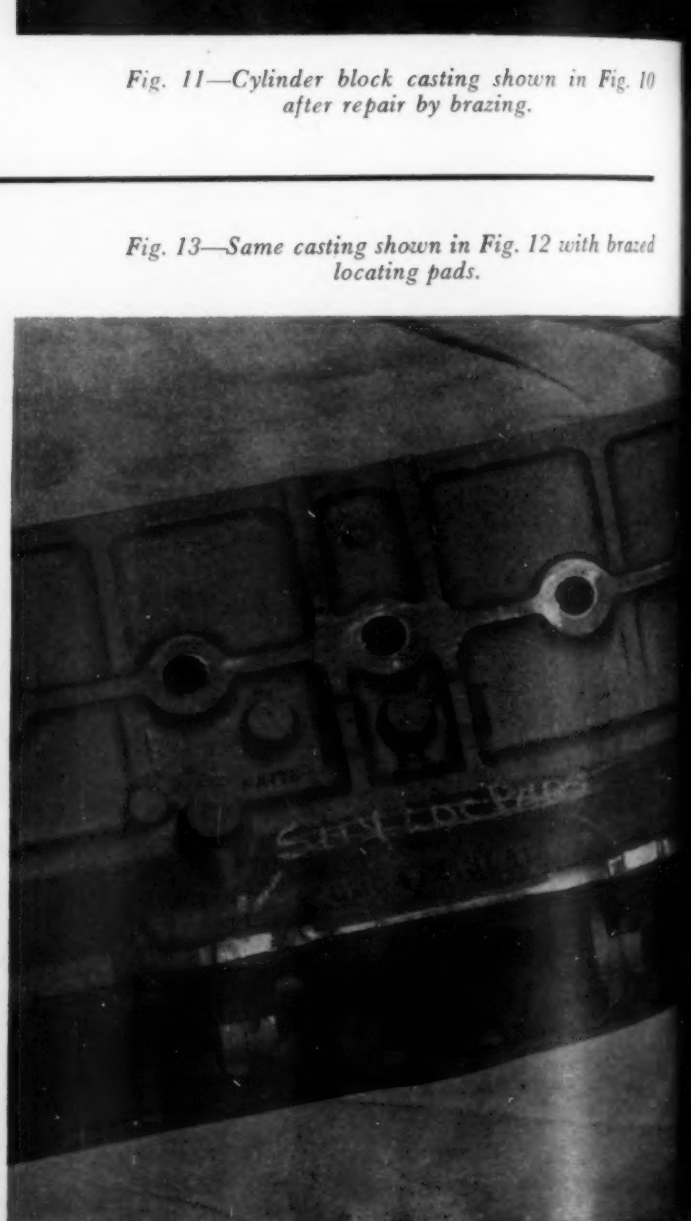








Fig. 14 (Upper Right)—Water pump casting rejected because of chipped snout. Fig. 15 (Above)—Cold arc welding the water pump casting (Fig. 14) with cast-iron flux-coated rod. Fig. 16 (Lower Right)—Water pump casting (Figs. 14 and 15) after being welded and ground.

thickness of approximately 3/16-in. were used is that about 85 per cent of the welding in this foundry is done with metal of that thickness. The brazed sample again showed the same comparative results. Both brazed samples broke about 1/4-in. away from the braze and neither showed any fracture at the bond.

Brazing Procedure. Castings to be repaired in this manner must be ground to solid metal. Flux is applied to the welding rod, and good quality flux is important. The casting is heated to a dull red at the point to be brazed, keeping heat away from the parts liable to be affected. The brass then is bonded to the parent metal and built up in a plastic state as fast as possible. An operator without sufficient practice to assure a good bond should not be allowed to braze. It requires an entirely different technique than cast iron welding. Once an operator has mastered brazing and uses it in the proper places, it is a great aid in salvaging cast iron.

Figure 10 shows a blow at the cope pan rail of a finished cylinder block. After cleaning out the defect for repair it was found to run into the bolt hole, which makes repairing difficult by most methods. This casting was brazed, and the repaired casting is shown in Fig. 11. A carbon was placed in the bolt hole, not screwed in but leaving enough stock for retapping without any further operation. The entire repair job was accomplished in 18 min.

Casting Finish

Figure 12 shows a rough cylinder block with shy locating pads. These could have been arc welded except for the fact that the rail, being high in the cope, is rather hard. Arc welding aggravates this condition and may cause a rough finish, and that finish being the main bearing race, no chances must be taken. Machining through the junction of a braze may be done without danger of encountering hard spots. The repaired cylinder block is shown in Fig. 13.

The author has used both Tobin and manganese bronze rods in brazing, with but little difference in the results. Either rod is recommended.

Arc Welding of Cast Iron. This operation is of a critical nature because of the many types of electrodes used and the many claims of their merits on castings of all designs and almost any state of completion. Arc welding, when properly used, has a valuable place in the salvaging of castings. However, it should be held well within its boundaries.

Arc welding may be divided into five types of repairs: (1) Preheat arc welding. (2) Cold arc welding rough castings. (3) Cold arc welding finished castings. (4) Cylinder bore welding or tack welding. (5) Intermittent welding, rough and finish.

Preheat Arc Welding. This is used to weld small defects which are clean. It is done by the welder after removing castings from the preheat furnace. Several small defects, not extending completely through the casting, may be present. He may arc weld these with a cast-iron flux-coated rod (note that this rod is of cast iron). He may then torch weld the major defect and stay well within the time limit. Many uses exist for this type of



Fig. 17 — Finished cylinder block casting showing blow at top face.

welding and this rod because its penetration is much greater than that of other rods and the weld remains soft.

Loose chaplets may be welded with assurance of being leakproof. The author does not recommend any type of arc welding of rough castings on machined faces except preheat arc with flux-coated cast-iron rod. This statement will meet with much opposition.

Cold Arc Welding. The use of this method on rough castings is limited. It should be used only on surfaces which are sufficiently heavy or designed to withstand the shock. Various kinds of electrodes are used, too many for the author to mention.

In Fig. 14 is shown a water pump with the snout chipped in handling. This casting is worth salvaging if a good, fast method is found. Figure 15 shows this casting being arc welded. The use of cast-iron flux-coated rod makes this weld extremely hard when not preheated, but the hardness makes no difference at this point, which is a hose connection and usually is hard when cast. Note that a carbon is used to eliminate as much grinding as possible.

Figure 16 shows one of these castings finished. They can be welded at the rate of 70 per hour. About six different electrodes will fill the needs for cold arc welding, but the one mentioned should certainly be included.

Arc Welding. On finished castings arc welding is done with one or more of the more expensive rods high in nickel, copper, etc., and flux coated. Castings should not be done too rapidly and, of course, at as low a heat as possible, using proper precautions. Repairs may be made close to cylinder bores and bearings without danger of distortion.

Figure 17 shows such a defect on a boss used for the number pad. A copper-covered flux-coated rod was used for this job. The results are shown in Fig. 18. Much more could be said on this subject, but the extreme example shows the merits of this type of repair, when properly used.

Cylinder Bore Welding. This method, although not new, has not been accepted by all engineers and manufacturers. Many millions of dollars worth of cylinder blocks have been salvaged by bore welding. The author would like to cover this subject as thoroughly as possible in order to clarify some of the doubt. The author has never encountered a single failure due to this method of repair. A good and successful procedure follows:

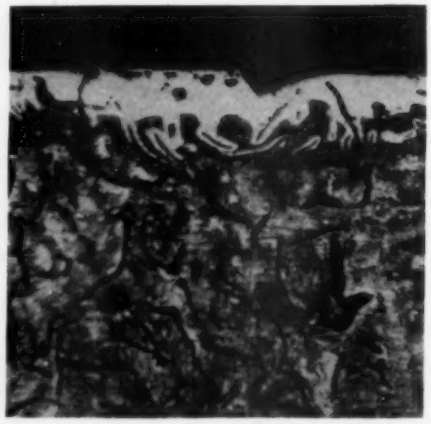


Fig. 19—Photomicrograph showing cylinder bore weld at weld junction. Top—100X. Bottom—500X.





Fig. 18 — Same casting (Fig. 17) after arc welding.

A. The flow should be chipped or cleaned out by a right-angle drill. The casting should be scrapped if the flaw extends more than one-half the thickness of the wall. Flaws as large as 3/16-in. in ring travel and as large as 5/16-in. outside of ring travel may be repaired. Of course, there is a difference of opinion on these limits.

B. Bore-nickel rod of not greater than 5/32-in. diameter should be used. The rod should be of commercial purity, or not less than 94 per cent.

C. The welding transformer should provide 6 to 12 volts with high heat (350 amp.) such that the nickel rod is melted and scratched off the wire while in a plastic form. No continuous arc is held as this would result in a hard, heat-affected zone.

D. This welding operation should be performed by an operator of considerable experience who has demonstrated his ability with samples which have been cut and examined for hardness and bond. A micro-examination of a cylinder-bore weld at the weld junction is shown in Fig. 19. The structure shows good

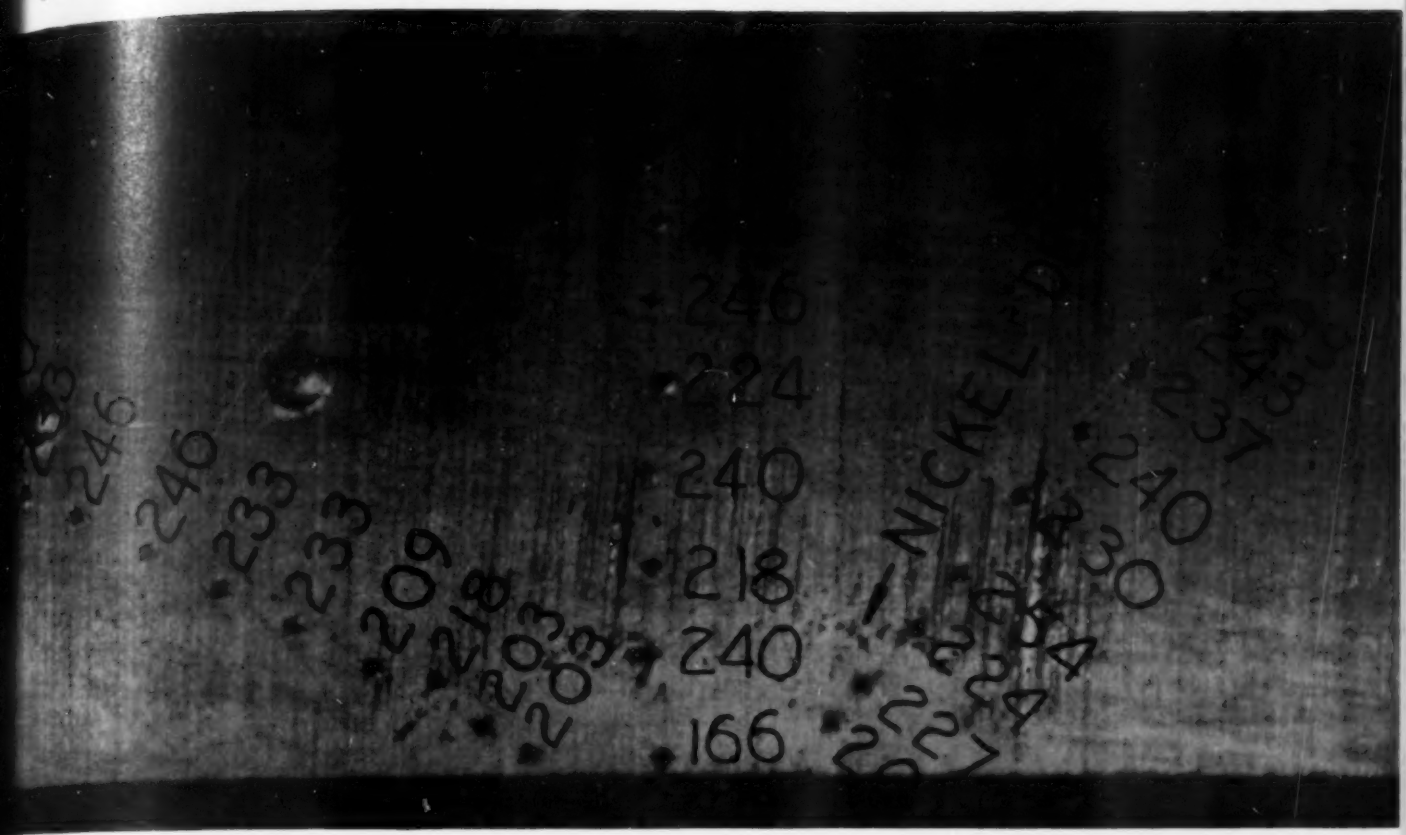


Fig. 20—Hardness transverse taken through welded cylinder bore section.

solubility of nickel and iron at the weld junction. A relatively sharp line of demarcation between the nickel weld and the parent metal is shown. The cast iron structure at the junction seems to be barely affected. Figure 20 shows hardness transverse taken through a welded cylinder bore section with Vickers Brinell 10-kg. load and hardness recorded. Note that the hardness of the nickel deposit was increased from 166 to 203 by peening before the weld was ground. This has no effect on the weld junction.

Tack Welding

This method of welding (also called tack welding) is of great value in the foundry. It is used to repair imperfections and improve the appearance of the castings. Used in this manner it requires no particular skill. The author has several of these transformers placed in the cleaning room of the foundry.

Intermittent Arc Welding. This method is similar to tack welding. In other words, it is not a continuous arc. It does not have the penetration of a straight arc, therefore is used mostly for repair of gasket surfaces, building up the edges of bolt bosses, etc. Its use is very limited due to the lack of penetration. A rod of about 74 per cent or more of nickel and 1/8-in. diameter is used. Little risk of damaging a finished casting is incurred with this method.

Cast Iron Soldering. This method of repair is a valuable aid in the salvaging of finished castings, and generally is used in the repairing of main and cam bearings in cylinder blocks. Spots which do not clean up, sand holes, small blows, and even small shrinks may be repaired and remachined without danger of damaging expensive tools.

Large areas at gasket faces may be repaired quickly. It may also be used on locating pads, where clamping is not too severe, and has a melting point of about 700°F. It is put on with a soldering iron which is at a dull red heat. It will not run and may be built up as high as is necessary. Commercially pure HCl should be used for the acid and surfaces must be clean. It is a time saver and does a much better job than welding when used within its limits.

Plugging. This probably is the greatest money saver in the salvaging of castings. The plug most commonly used is the 1/8-in. pipe plug of annealed cast iron. It has 27 threads per in., allowing a safe repair in metal thickness of 5/32-in. It is not advisable to use larger standard pipe plugs except in metal of greater thickness because these plugs have only 18 threads or less per in.

The author uses a 1/4 pipe plug with 20 threads per in. These cannot be procured on the market. However, the stock can be cast and annealed in the foundry and special tools may be ordered to make and use them. This plug allows a 1/2-in. diameter hole to be repaired in metal wall thickness of 5/32-in.

Conclusion. No attempt has been made to cover completely the subject of welding. X-ray testing and many other subjects relating to welding can be covered much better under their particular headings. However, the author hopes that this article will help to keep many castings out of the scrap pile, with an added assurance of safe and dependable methods of salvage.

HEAT TREATMENT of STEEL CASTINGS

Steel Division

WAR CONDITIONS restricted the activities of the Committee on Heat Treatment of Steel Castings because much of the work being done on steel castings was of a confidential nature, and these restrictions still continue at the time of this report. However, the committee has followed the developments in the field of heat treatment and briefly summarizes them in this report. In addition, it has made a survey to determine the acceptance by the foundry industry of the end-quench test for hardenability.

Widespread use of steel castings under severe wartime conditions made it necessary that optimum properties be obtained in the castings. To obtain these properties it was necessary to liquid quench many types of castings that formerly would have been normalized or annealed. Production of quenched and drawn castings presented several problems, namely:

1. Choosing a steel of chemical composition such that it would harden to the required depth in various sections during quenching.
2. Heating castings uniformly for quenching.
3. Quenching under controlled conditions.

A simple end-quench test, developed to measure the hardenability of wrought steels, was readily adapted to measuring the hardenability of cast steels. As a matter of fact, it was found that cast and wrought steels of like analyses had quite similar hardenabilities. This test enabled the foundrymen to determine rather easily the maximum thickness of section of any particular steel that would harden to the depth required for the particular application at hand. The extent to which the end-quench test was adopted is indicated in the survey at the end of this report.

Having chosen a steel of proper composition by the use of the end-quench test or by other means, it was necessary for best results that each casting made from the steel be heated uniformly to the desired temperature for quenching. This requirement led to the design and installation of heating furnaces especially designed to give uniform heating, as well as to provide improvements in the methods of loading the furnaces.

The most popular type of heating furnace has been either the batch or the semi-continuous furnace because of flexibility in handling a variety of castings. Where production warranted, a continuous heating furnace was used. In one continuous furnace, the castings were hung from rods attached to a monorail above the furnace, which had a narrow slot in the roof through which the rods extended and moved horizontally.

A survey of the end-quench test in the steel foundry industry has been the principal activity of the Committee on Heat Treatment of Steel Castings since 1944. A questionnaire, designed to elicite complete information on the subject, was compiled and sent to steel foundrymen.

Water has been the almost universal medium used for quenching castings. Two quenching techniques have been widely used, namely, timed and intermittent quenching. Timed quenching consists of quenching a casting for a predetermined time before removing it from the quenching bath. Intermittent quenching is accomplished by quenching for a predetermined time, lifting the casting from the bath, allowing the temperature of various sections of the casting to equalize somewhat, and then again quenching the casting.

Quenching Speed

Careful segregation of castings into groups that are to be quenched at one time is necessary if controlled quenching conditions are to be maintained. The speed at which a steel must be quenched to obtain the desired hardness is indicated by the "S" curve for that steel. During the past few years, "S" curves for many of the cast steels have been worked out and have served as useful guides in setting up quenching conditions. Careful attention to the design of castings so as to eliminate abrupt changes in section and sharp corners has materially aided in the production of quenched castings.

Tempering or drawing of the quenched castings has followed more or less conventional practices. Perhaps the chief difference in handling

Presented at a Steel Session of the Fiftieth Annual Meeting, American Foundrymen's Association, at Cleveland, May 8, 1946.

of quenched castings lies in the fact that they should be put into the tempering furnace as soon as possible after quenching.

Several heat-treating practices are being used for wrought steels which, to the committee's knowledge, are not widely used on castings. These are as follows:

1. Flame hardening.
2. Induction hardening.
3. Quenching into molten salt or metal baths.

The first two methods generally are used to produce a hard-surface layer on parts without hardening the entire part. They, undoubtedly, will find increased application to those castings where a hard-wearing surface backed by a tough core is needed.

Quenching into molten salt or molten metal baths is a means of eliminating cracking in parts with nonuniform sections. This method of quenching may be particularly applicable to quite complex castings. Quenching into molten salt or metal baths is also used to effect isothermal transformations so as to produce certain characteristic structures other than those obtained by quenching and drawing. The extent to which this method of quenching may be used on steel castings is not predictable at the moment.

End-Quench Test

Since 1944 the principal object of this committee has been a survey of the use of the end-quench (Jominy) test in the steel foundry industry. A questionnaire campaign was conducted to elicit as complete information as possible on the subject. The questionnaire was quite complete. It asked the purpose of the test, whether applied to research or production control; application as to product specifications; frequency of tests; source of Jominy specimen; comparisons with similar tests on wrought steels; interpretation of results; correlation with actual shop quenching practice; value of tests; and comments on the method as described in the covering ASTM specification.

The questionnaire proved to be premature, particularly as to the details sought, since the method was new to many steel foundrymen and experience was just being gathered. However, the response was gratifying and the results are considered of interest despite the fact that the report covers the situation as it existed over a year ago. Since then there has been a considerable but undetermined increase in the use of the test by steel foundrymen.

Studies looking to establishing hardenability specifications are under way by several groups, and it is probable there will soon be some formal specifications of this kind.

Results of the questionnaire are summarized as shown at right.

Only three report a consistent difference in hardenability between cast and wrought steels of similar chemistry. In view of the new generally accepted equivalence of hardenability, with similar chemistry, of cast and wrought steels, these three reports can be explained as due to limited experience with a new test.

Total questionnaires (to previously screened list)	45
Total replies (82 per cent return)	37
Replies with concrete information (71 per cent return)	32
End quench test used	
In research	18
In research and production	13
In production	1
To study new composition	25
For production control	5
Total tests per month	6 to 250
Percentage of tests on those classes or specifications tested, per cent	1 to 100
Using test to indicate	
Maximum thickness that can be hardened to required depth	11
Performance of individual heats	10
Source of end quench specimen	
Keel block	23
Special specimen	11
Checking grain size	
ASTM method	11
Special method	4
Aim to fully harden maximum casting section	10
Aim to fully harden critical zone of casting as to service conditions	13
Find end quench test helpful in this	14

Personnel of the Committee on Heat Treatment of Steel Castings is as follows: E. R. Young, *Chairman,* Climax Molybdenum Co., Chicago; N. A. Birch, American Brake Shoe Co., Mahwah, N. J.; H. H. Blosjo, Minneapolis Electric Steel Castings Co., Minneapolis; Werner Finster, Reading Steel Casting Div., American Chain & Cable Co., Reading, Pa.; R. A. Gezelius, General Steel Castings Corp., Eddystone, Pa., and C. T. Greenidge, Battelle Memorial Institute, Columbus, Ohio.

INSPECTION---THE BACKBONE OF QUALITY CONTROL

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Director of Research &
Foundry Engineer
Samuel Greenfield Co., Inc.
Buffalo, N. Y.

QUALITY through laboratory control is a phrase that has become the by-word and creed of most nationally advertised products, from toothpastes and canned soups to ingots and castings. However, retrospection on this phrase with respect to all nominal foundry operations, leads one to feel that it should be changed to "Quality Through PLANNED Inspection."

No matter how thoroughly and completely the Metallurgical, Engineering and Production departments organize their personnel and consider all the variable operational factors with standard practice procedures, proper inspection or policing of such practices is absolutely imperative. Otherwise, such procedures are foredoomed to the scrap pile along with a greater number of castings than necessary.

Proper Training

It is a recognized fact that a properly trained and organized Inspection Department is of paramount importance to all successful foundry operations. An Inspection Department organized for PLANNED Inspection is a logical method for improving quality of all products.

PLANNED inspection includes Process inspection; Lay-out inspection; Agenda of Reports from the laboratory relative to chemical analysis, physical analysis, etc.; Nuclear inspection; Non-destructive inspection; Eye or visual inspection, and Destructive inspection.

Wartime quality of castings is now demanded in peacetime production. Therefore, it behooves all astute and wide-awake foundry managements to institute the basic principles involved in Planned Inspection, discussed in this paper.

Layout inspection, and destructive, non-destructive and visual inspection, are well known to the average foundry operating today. Involved in these procedures are dimensional tolerance checks, X-ray inspection, magnaflux inspection, casting sectioning for photomicrographs as well as physical testing and pressure testing.

The value of these various operations has been proved many times during the last several years of specifications more difficult and more exacting than any previously imposed upon the foundry industries. Nuclear inspection is of much value in specialized casting applications where the structure of the material is of prime importance. In Nuclear inspection the X-ray diffraction technique is used for determining the diffraction pattern of the material.

The value of Process inspection and the agenda and correlation of laboratory reports, while of proven value during the war in large production foundries of single casting operation, is now becoming even more necessary to the average jobbing and production foundry.

Process inspection involves complete and thorough inspection of all the variable operating practices and conditions for which standard practice procedures are instituted by the Production, Metallurgical, or Engineering departments. This inspection should be by observation. By observation is meant that it should be the Inspection department's responsibility to observe all operations and practices for which procedures have been issued, but it should not be the Inspection department's prerogative to stop Production or other personnel from any deviation that might be observed from these standard practices.

It should be the Inspection department's sole duty to report any and all discrepancies to those supervisors directly affected by the deviation from standard practice. These reports should be made in writing and distributed to the affected parties so that necessary steps for the correction or discussion of such deviations can be instituted.

Duties Outlined

Thus we find, in many large foundries, a definite outline of duties for the Metallurgical and the Inspection departments. For example, the Metallurgical department is responsible for procedures and methods of melting, sand mixes and special processes such as heat treating and welding. The Inspection department's responsibility is to observe the procedures and methods covered by standard practice bulletins issued by the Metallurgical department.

Similarly, it is the responsibility of the Metallurgical department to issue specifications for raw materials and their physical and/or chemical

This paper was presented at an Inspection of Castings Session of the Fiftieth Annual Meeting, American Foundrymen's Association, at Cleveland, May 7, 1946.

analysis. The Inspection department's responsibility is to observe that only those raw materials approved by the Metallurgical department as used in production.

It has been found through experience that this practice of process inspection and constant correlation and liaison between the Inspection, Metallurgical and Production departments substantially improves the quality of all products produced under such means of control and decidedly reduces internal scrap percentages and customer returns. The extra amount of personnel and work required will more than justify the small extra expense involved.

That the wartime quality of castings, now demanded in peacetime production, can be made profitably only through complete and honest methods, is now self-evident to most foundry managements. Therefore, it behooves all astute and wide-awake foundry managements to institute the basic principles involved in PLANNED Inspection.

DISCUSSION

Chairman: H. W. WARNER, Allis-Chalmers Mfg. Co., Milwaukee.

Co-Chairman: E. G. LEVERENZ, American Steel Foundries, E. Chicago, Ind.

MEMBER: I would like to ask Mr. Gunselman to repeat his PLANNED Inspection, letters referring to inspection types.

MR. GUNSELMAN: "P" is for process inspection; "L" for layout inspection; "A" for agenda, checking physical and any other laboratory reports; "N" for non-destructive inspection; "N" for nuclear inspection; "E" for eye or visual inspection; "D" for destructive testing.

MEMBER: Mr. Gunselman used the expression "quality control." Did you do anything with quality control charts?

MR. GUNSELMAN: There are two definitions of quality control. One is spot control, where you set up charts and pick out particular castings to represent the entire lot of castings. That is not the type of quality control I was referring to. The quality control I was referring to is proven quality and not guess work.

CHAIRMAN WARNER: There are many companies following that chart system throughout, hence, it might be a good idea to have the defects listed. And one advantageous feature about it is that in future work it is well to have a follow-up system.

When new orders are issued the usual procedure is to get the file card on a particular casting and to check the defect that was found on previous orders. This in turn is listed on a production order issued to the shop, in which the foreman can see and guard against any defect before running the complete order.

It is the common practice for the foundryman to take a pattern out of the pattern vault and run the job as it was previously gated and not taking any particular notice of facing necessary to produce the best possible casting. Therefore, the above system warns the foundryman and supervision of possible pitfalls before the castings get into production processes.

HEAT TREATMENT of CAST ALUMINUM ALLOYS

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and
W. E. Sicha
Aluminum Company of America
Cleveland

EFFECTS OF HEAT TREATING aluminum alloys can be followed more readily after examination of the sequence of events occurring during the solidification process. An aid in studying the course of solidification is provided by constitutional diagrams such as have been established for the commercially important aluminum alloy systems[1,2].

These constitutional diagrams are constructed from data obtained by observing the melting and freezing characteristics and the microstructures of a series of alloys in which the relative concentrations of two or more metals are varied. The following discussion of solidification will be based on the partial constitutional diagram of a hypothetical two-component system shown in Fig. 1.

An alloy of the composition indicated by the line *R-S* is completely liquid at the temperature represented by *R*. Solidification starts at a temperature corresponding to point *T*. The first material to solidify does not have the composition of the hypothetical alloy but has a concentration of element *B* indicated by the point *U*. In other words, the first material to solidify contains a far lower concentration of alloying element than is represented by the average composition of the alloy.

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However, as solidification proceeds, the material solidifying increases in concentration of element *B* along line *C-D*. Since this material continues to contain less than the average content of element *B* in the alloy, the concentration of element *B* continues to increase in the liquid along the line *C-E*. For example, at temperature *W* the composition of the material that solidifies is represented by *X*, and the remaining liquid, which has been enriched with element *B*, is of the composition indicated by *Y*. At the temperature *V*, the remainder of the liquid solidifies, and this material has an average composition designated by *E*.

This last material to solidify, known as eutectic, consists of two constituents of the compositions indicated by *D* and *F*. The temperature at which solidification is completed is referred to as the eutectic temperature, and the temperature interval *T-V* is the solidification range of the alloy.

Factors involved in the successful heat treatment of aluminum alloy castings have become of more general interest as a result of the extensive increase in use and production. A thorough understanding of the commercial applications of heat treatments for aluminum alloy castings requires a knowledge of the process fundamentals. It is the purpose of this paper to review these fundamentals and to consider the structural and property changes produced in aluminum alloys by commercial heat treatments.

The resultant solidified metal is made up of two fundamental types of materials. One type forms the continuous background or matrix. This is the material that solidifies throughout the temperature range *T-V*. As solidification progresses in this range, successive layers of freezing material build up on and around the crystals that froze initially.

Dendritic Structure

This freezing material assumes a definite form, similar in appearance to a pine tree. The treelike pattern, known as dendritic structure, is illustrated in Fig. 2. It is evident from the manner in which the dendrite arms form that the concentration of *B* metal is progressively greater from the core to the outer layer. This constituent of varying composition is called a solid solution.

At final solidification, occurring at the temperature indicated by the line *D-E*, the eutectic freezes between and around the dendrite branches. As previously mentioned, the eutectic is made up of intimately mixed solid solution of composition *D* and intermetallic compound of the composition given by the line *G-H*. The latter constituent is called an intermetallic compound largely because of its relatively unvarying composition, and is the other phase or fundamental type of material in the solidified metal.

The properties of both the solid solution and intermetallic compound phases of the solidified metal differ from those of pure *A* metal. Within the solubility limits of *B* in *A*, increasing concentrations of *B* gener-

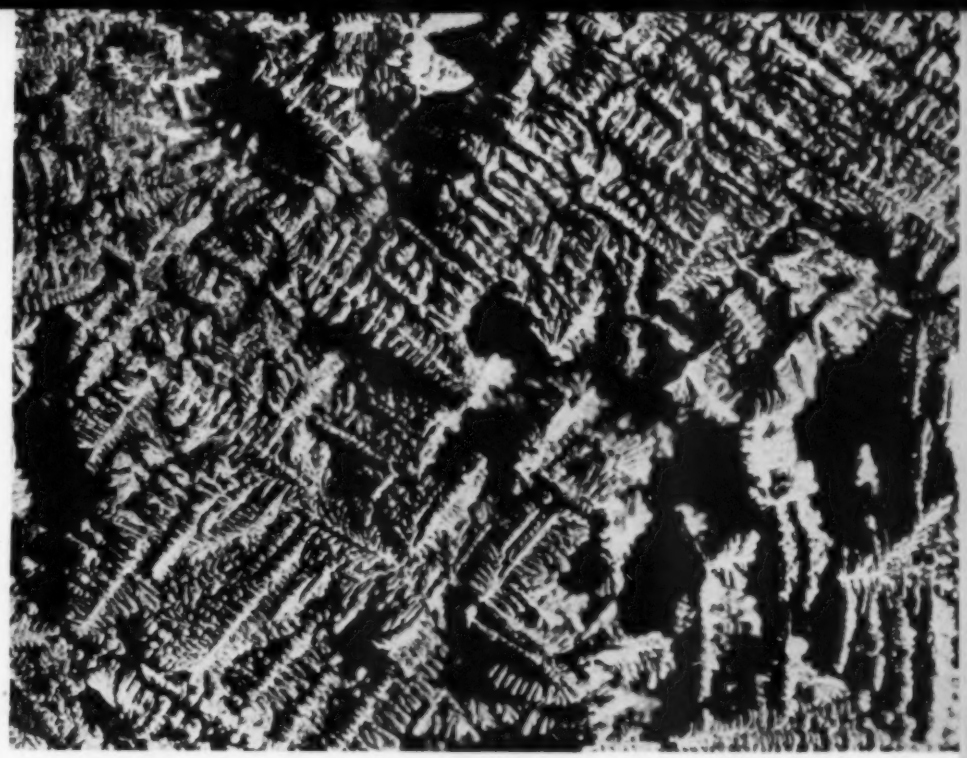


Fig. 2— Dendritic structure developed during metal solidification.

ally produce an alloy of gradually increasing tensile strength and decreasing ductility. Intermetallic compounds usually are hard and brittle. These phases naturally impart their characteristics to the alloy to a degree dependent upon the concentrations present.

Variations in concentration in the solid solution tend to be reduced by diffusion during the solidification process. Also, diffusion is the mechanism by which changes in structure occur as a result of the reduced solid solubility of *B* in *A* with decreased temperature. However, time is an important factor in the process of diffusion, and a rapid rate of solidification will retain concentration gradients which are not in equilibrium at the lower temperatures.

The solid lines in Fig. 1 represent the structures which should prevail under equilibrium conditions or after slow cooling to any specific temperature. With cooling rates such as ordinarily prevail in production of commercial castings, equilibrium is not attained and structural conditions may be more nearly as represented by the dotted lines in Fig. 1. Undercooling, resulting from rapid cooling, tends to lower the temperatures of initial and final solidification. The faster cooling rate results in the formation of relatively smaller volumes of solid solution constituent and larger volumes of intermetallic compound than would occur under equilibrium conditions. This effect is illustrated by the shift of maximum solid solubility from position *D* to *D'*.

Rapid cooling also tends to preserve supersaturated solid solutions and, as is indicated by the relative positions of *I* and *I'*, to retain a larger amount of solid solution. Another point of difference in commercial casting alloys is their normal impurity content. Presence of these impurities introduces one or more additional constituents that have the effect of impeding the normal tendency for adjustment toward equilibrium during solidification.

Heat Treatment. The fact that the solubility of certain elements in aluminum varies with temperature, in the manner indicated by the line *D-I* in Fig. 1, is the basis for heat treatment. The objective of heat treatment is the controlled alteration of properties. Relationship with the foregoing discussion of the solidification process will be evident as the various types of commercial heat treatments are considered.

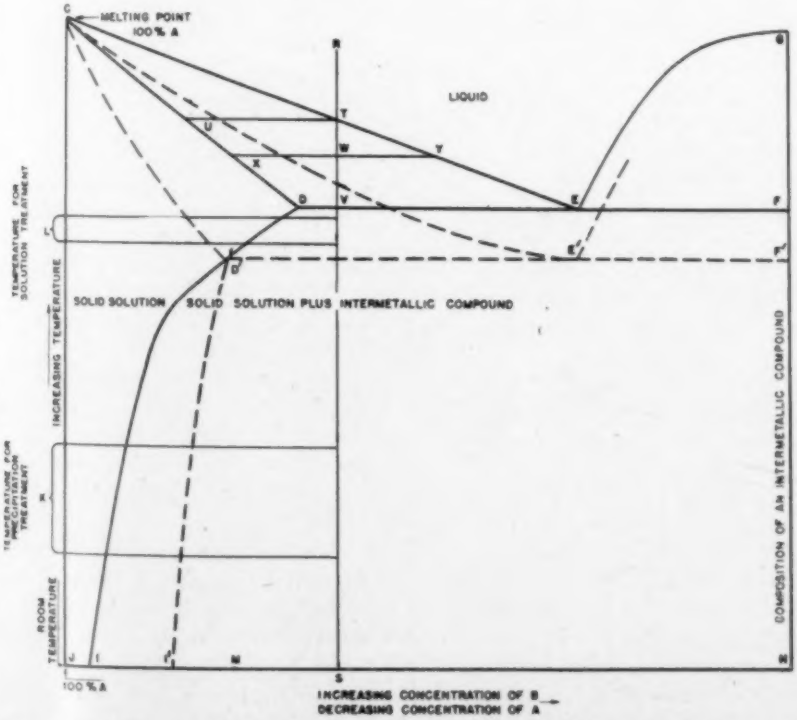


Fig. 1—Partial constitutional diagram of a hypothetical two-component system—course of solidification.

Types of Heat Treatment

Heat treatments applied to aluminum alloy castings can be classified under the following general types:

1. Annealing treatment (intermediate temperature).
2. Solution treatment (high temperature).
3. Artificial aging treatment only (low and intermediate temperatures).
4. Solution treatment followed by low temperature artificial aging treatment.
5. Solution treatment followed by intermediate temperature artificial aging treatment (overaged).

Data on the mechanical properties and uses of heat treated aluminum

casting alloys are generally available[2, 3, 4, 5, 6]. Examples of some of the more generally used commercial heat treatments are given in Table 1.

Casting Requirements. Castings must be of the specified chemical composition and of good quality in order to obtain the expected response to heat treatment. This fact is generally appreciated but it can not be emphasized too strongly. Factors contributing to the production of sound and fine grained aluminum alloy castings are treated in other publications[1, 4, 6].

Furnaces for Heat Treatment. Air-chamber furnaces of proper design are best suited to heat treatment of aluminum alloy castings. Furnaces may be heated by electric heating units, radiant tubes, or, if harmful products of combustion do not enter the heating chamber, by gas or oil burners. Gas furnaces in which the products of combustion are circulated, when suitably designed, can be used satisfactorily in many cases provided that the atmospheres contain a sufficiently high percentage of carbon dioxide and are free of sulphur compounds and excessive moisture[7].

Furnace design should provide a uniform temperature within $\mp$ 5° F. throughout the working zone. Mechanical circulation of the air is essential to accomplish this result and to decrease the heating cycle. Baffles generally are necessary to prevent air being circulated directly from the heating units onto the metal. For electrical resistance heating, units with a low energy imput per unit of area are desirable.

Temperature Control. Furnaces should be equipped with recording pyrometers and automatic temperature control instruments to insure the necessary degree of control. Automatic limit controls which can be set at a temperature slightly higher than the operating temperature are also desirable to provide insurance against damage from accidental overheating through failure of the control equipment. All pyrometric equipment including the thermocouples should be frequently checked against standards.

Furnaces should be surveyed before putting them into service to make certain that the temperatures throughout the working zone are within the prescribed limits of $\pm$ 5° F. This entails checking the temperatures through the heating cycle with several thermocouples placed in different locations in a typical charge of castings.

Annealing Treatment. Castings frequently are annealed for the purpose of relieving the internal stresses that may be induced by contraction during cooling around more or less resistant cores, or by the differential cooling rates of sections of varying thickness. Annealing is also employed to bring about precipitation of alloying constituents from solid solution in which they may have been retained by a relatively rapid rate of cooling in production.

Thus, annealed castings have a structure that is practically insensitive to change at the temperatures ordinarily encountered in elevated temperature operation. They are not subject, therefore, to permanent dimensional changes in service.

Microstructure Changes

Changes in the microstructure resulting from annealing may be visible but usually are not extensive. The microstructures of commercial aluminum casting alloys shown in Figs. 3, 7, 11, 15 and 19 are all representative of annealed material, although most of them are of "as-cast" materials.

The mechanical properties produced by annealing usually are similar to those of the "as-cast" material, although in some alloys tensile strengths slightly lower and elongations slightly higher than the "as-cast" properties result. As would be expected, this type of treatment has limited commercial application.

Solution Treatment. Referring again to the equilibrium diagram of Fig. 1, it will be noted that the line *D-I* indicates the limits of solid solubility of *B* in *A* at different temperatures. The solid solubility increases as the temperature increases, and this change is most marked in the range just below the eutectic melting temperature indicated by line *D-E*. Alloys having such solid solubility-temperature characteristics are susceptible to solution heat treatment.

Aluminum-copper and aluminum-magnesium systems, which are of this type, constitute the basis for commercially important casting alloys. Magnesium and silicon form an intermetallic compound which has solubility characteristics similar to those of copper and magnesium.

Table 1

HEAT TREATMENTS FOR SEVERAL ALUMINUM CASTING ALLOYS

Alloy and Heat Treatment[1]	*Solution Heat Treatment* — *Time, hr.*[2]	*Solution Heat Treatment* — *Temperature, ° F.*[3]	*Solution Heat Treatment* — *Quench*	*Aging Treatment* — *Time, hr.*[4]	*Aging Treatment* — *Temperature, ° F.*[3]
SAND CAST					
122-T61	12	950	Water at 150 to 212° F.	10-14	310
142-T21	—	—	—	2-4	650
142-T77	6	970	Still Air	1-3	650
195-T4	12	960	Water at 150 to 212° F.	—	—
195-T6	12	960		3-5	310
195-T62	12	960		12-16	310
355-T6	12	980		3-5	310
355-T61	12	980		8-10	310
356-T6	12	1000		2-5	310
PERMANENT MOLD					
122-T65	8	950	Water at 150 to 212° F.	7-9	340
A132-T65	8	960		12-16	340
142-T61	4	960		3-5	400
B195-T4	8	950		—	—
B195-T6	8	950		5-7	310
355-T6	8	980		3-5	310
355-T62	8	980		14-18	340
356-T6	8	1000		3-5	310

[1]Aluminum Company of America alloy and heat treatment designation.
[2]Soaking time periods required for average castings after load has reached specified temperature. Time can be decreased or may have to be increased, depending upon particular castings, as demonstrated by experience.
[3]Temperature setting for control instrument. Variation of temperature in furnace should not exceed $\pm$ 5° F.
[4]Exact time required influenced by foundry variables and should be selected on basis of obtaining desired typical hardness values.

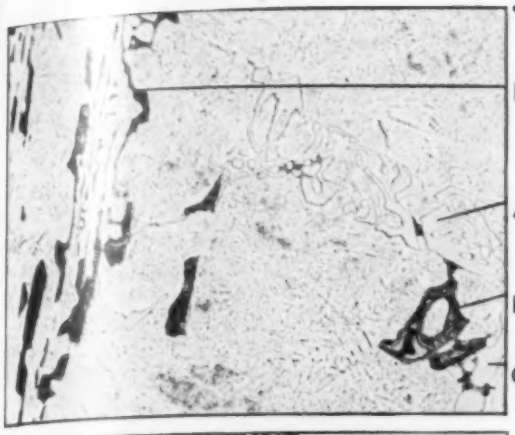


Fig. 3 — Annealed condition. The fine, generally distributed dark material in the matrix is $CuAl_2$ and Mg_2Si precipitate. In the "as-cast" condition, a much smaller quantity of precipitate is present, but in other respects the microstructure is identical with that shown.

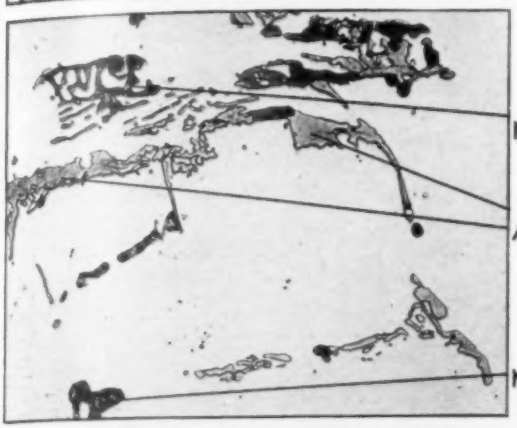


Fig. 4—Normal solution heat treated condition. Solution treatment has dissolved all the $CuAl_2$ and some of the Mg_2Si. The remaining Mg_2Si also has been spheroidized.

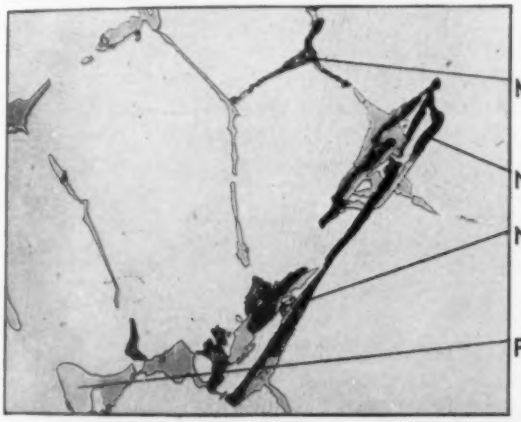


Fig. 5 — Insufficient solution heat treated condition. Some $CuAl_2$ remains undissolved and the Mg_2Si still is angular, indicating incomplete heat treatment.

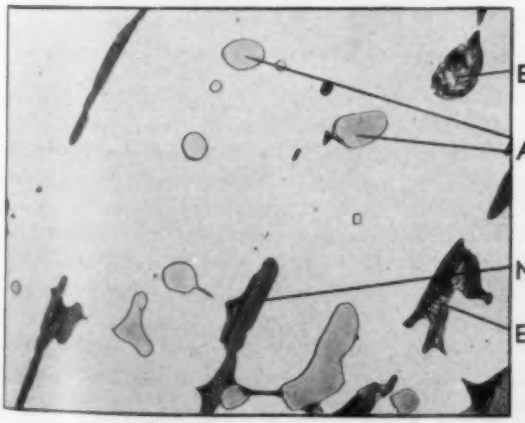


Fig. 6 — Overheated or "burned" condition. Evidence of overheating is shown by eutectic deposits and also agglomeration and spheroidization of the insoluble constituents.

0.5 per cent HF etch 500X

Representative Photomicrographs of No. 142 Aluminum Alloy.

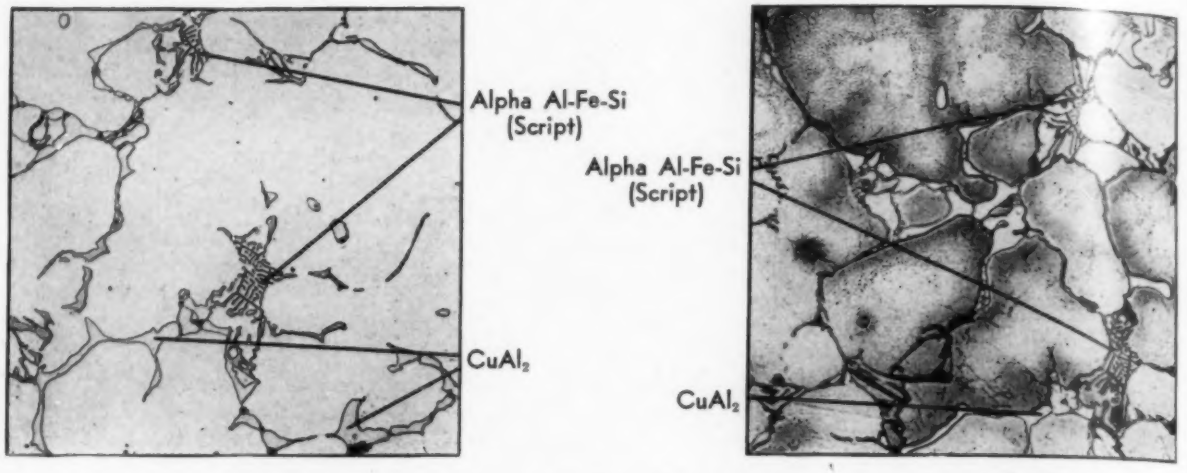


Fig. 7—Representative photomicrographs of No. 195 aluminum alloy. "As-cast" condition. These specimens show a large amount of $CuAl_2$ which is typical of the "as-cast" condition. Note cored structure revealed by mixed acid etch. Left—0.5 per cent HF etch. Right—2.5 per cent HNO_3, 1.5 per cent HCl, 1.0 per cent HF etch. 250X.

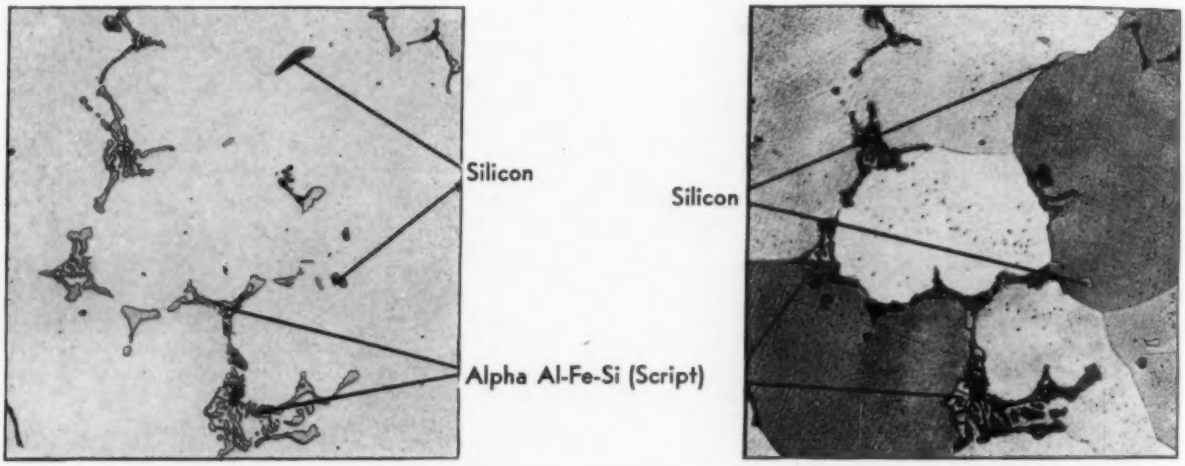


Fig. 8—Representative photomicrographs of No. 195 aluminum alloy. Normal solution heat treated condition. Note that the $CuAl_2$ has been completely dissolved. Also note absence of the cored structure which appears in the "as-cast" condition. Left—0.5 per cent HF etch. Right—2.5 per cent HNO_3, 1.5 per cent HCl, 1.0 per cent HF etch. 250X.

This compound, known as Mg_2Si, forms a pseudobinary system with aluminum, and casting alloys of this system are heat treated commercially. A noteworthy point regarding the compositions of these commercial alloys is that the concentration of soluble element *B* generally is lower than the concentration designated by *D* in Fig. 1, and usually is of the order of composition indicated by *M*, Fig. 1.

In general, the solution treatment temperature for an alloy is limited by the solid solubility line and the eutectic or solidus temperature, and generally falls in the range indicated by *L* in Fig. 1. Since the rate of atomic diffusion of the soluble constituent increases rapidly with increasing temperature, use of the highest safe temperature is desirable.

The maximum temperature is limited by the eutectic melting temperature, and may be governed also by the strength of the alloy at that elevated temperature. Castings should be capable of retaining their shapes, with little or no support, during solution heat treatment. This limitation may necessitate use of a heat treating temperature little above the solid solubility temperature.

Relatively long heating periods are required for solution of the soluble constituents in castings, even with the increased diffusion rate at the heat treating temperature. The time at temperature for attainment of homogeneity is less the shorter the distances through which diffusion must take place. Consequently, solution is more rapid in castings with a small macro grain size and fine dendritic structure. Hence permanent mold castings can be treated in a shorter time than sand castings.

The rate of cooling from the solution heat treating temperature must be rapid in order to preserve at room temperature the solid solution that

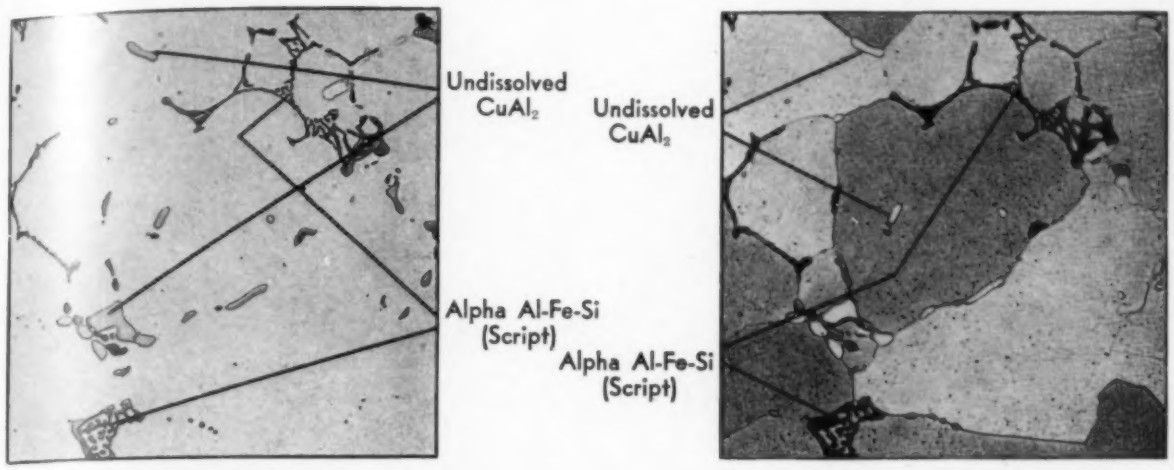


Fig. 9—Photomicrographs of No. 195 aluminum alloy. Insufficient solution heat treated condition. Note large amount of $CuAl_2$ still undissolved. Left—0.5 per cent HF etch. Right—2.5 per cent HNO_3, 1.5 per cent HCl, 1.0 per cent HF (Keller's) etch. 250X.

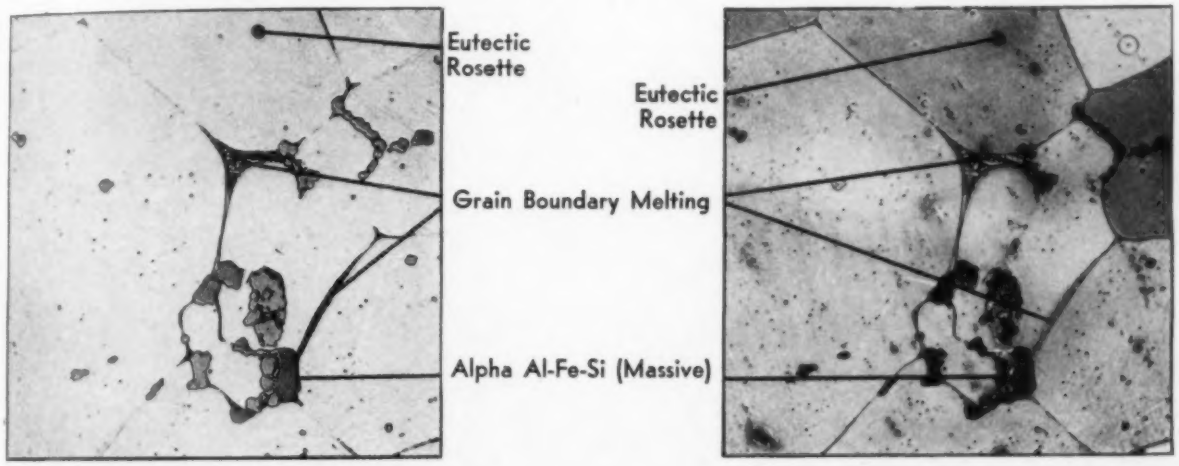


Fig. 10—Photomicrographs of No. 195 aluminum alloy. Overheated or "burned" condition. Overheating is indicated by grain boundary melting of the eutectic material and by presence of the eutectic rosette. The massive Al-Fe-Si constituent may be formed by agglomeration of the script as a result of overheating. Left—0.5 per cent HF etch. Right—2.5 per cent HNO_3, 1.5 per cent HCl, 1 per cent HF etch. 250X.

was in equilibrium at the heat treating temperature. However, an excessively rapid quench can cause extensive distortion through the setting up of excessive internal stresses. Commercial quenching practice is based upon maintaining the solid solution and minimizing the latter conditions.

Solution treatment materially increases the mechanical properties. Some of the physical properties such as density, and thermal and electrical conductivities also are altered by solution heat treatment, but to a degree that is generally unimportant except in special casting applications.

The microstructures of some solution heat treated commercial casting alloys are shown in Figs. 4, 8, 12, 16 and 20. As an aid to the metallographer, Figs. 5, 6, 9, 10, 13, 14, 17, 18, 21 and 22 provide photomicrographs illustrating improper solution treatment.

Precipitation Hardening. A supersaturated structure, such as is produced by solution treatment, tends to adjust to equilibrium conditions. Alloy constituent or compound is precipitated from the solid solution in the course of this adjustment. The rate of precipitation is a function of temperature since diffusion is essential to the progress of precipitation.

Precipitation in a solution treated alloy can be restrained almost indefinitely by keeping the material in liquid air. At such low temperatures, the atomic mobility is too low to permit the diffusion necessary for precipitate formation. The rate of precipitation at room temperature is relatively slow in most commercial aluminum casting alloys, but the rate varies with alloy composition. With increasing temperature, precipitation is progressively accelerated. An indication of the temperature range in which precipitation treatments are performed is shown at *K* in Fig. 1.

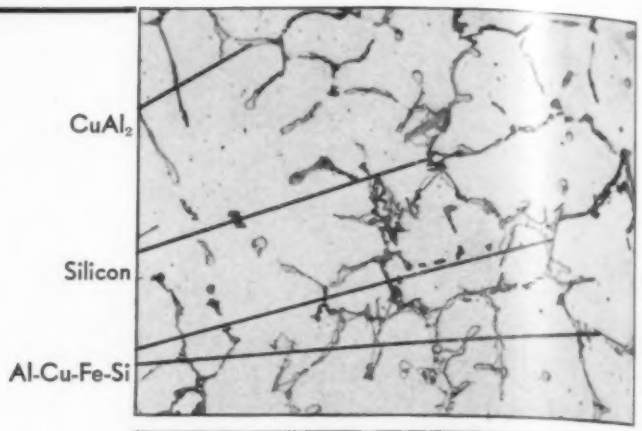


Fig. 11 — "As-cast" condition. Three constituents are readily visible—rounded white $CuAl_2$, angular blue-gray silicon, and needlelike tan Al-Cu-Fe-Si. All form a eutectic concentrated along the dendrite arms and along the grain boundaries.

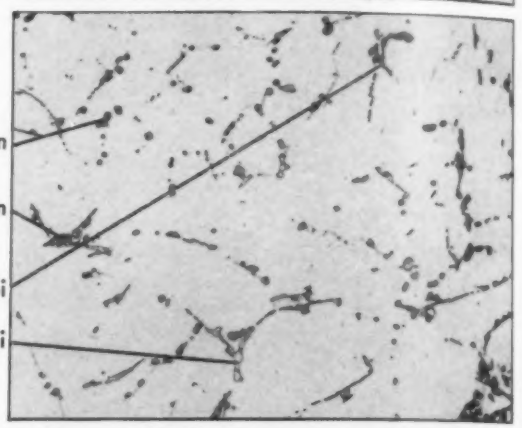


Fig. 12—Normal solution heat treated condition. Solution heat treatment has dissolved most of the $CuAl_2$, rounded and agglomerated the silicon particles and has not changed the Al-Cu-Fe-Si.

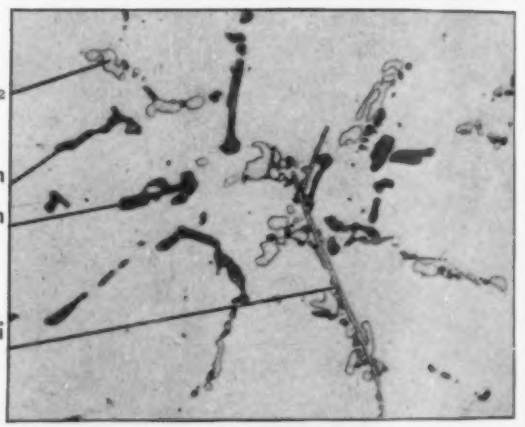


Fig. 13 — Insufficient solution heat treated condition. Solution heat treatment has not been carried on long enough or at a sufficiently high temperature to dissolve enough of the $CuAl_2$ and agglomerate the silicon.

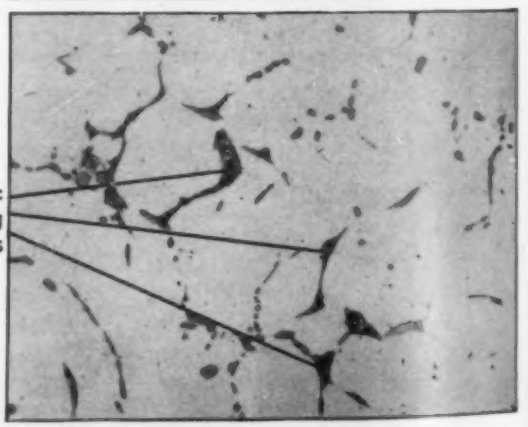


Fig. 14—Overheated or burned condition. Temperature has exceeded the eutectic melting point. Angular deposits are unresolved eutectic which at one time during heat treatment had been molten.

0.5 per cent HF etch 250X

Representative Photomicrographs of No. B195 Aluminum Alloy.

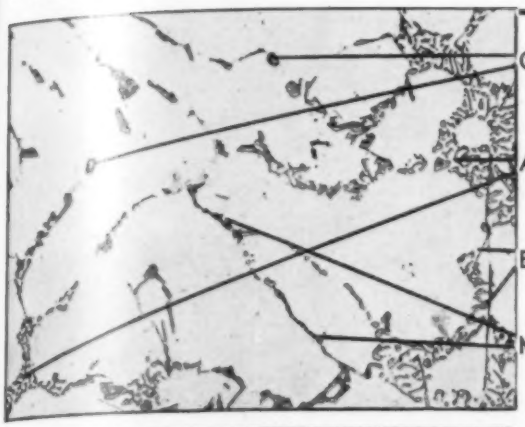


Fig. 15 — "As-cast" condition. The chief constituent of this alloy is the aluminum-silicon eutectic. The precipitation hardening agents are $CuAl_2$ and Mg_2Si.

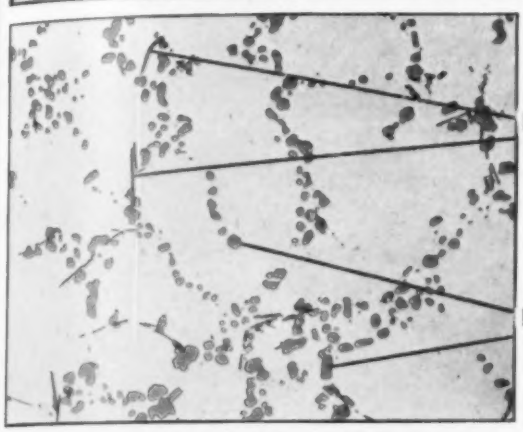


Fig. 16—Normal solution heat treated condition. The silicon particles have become rounded and have agglomerated to a certain extent due to heat treatment. The $CuAl_2$ and Mg_2Si have gone into solution but the beta Al-Fe-Si constituent is unaffected.

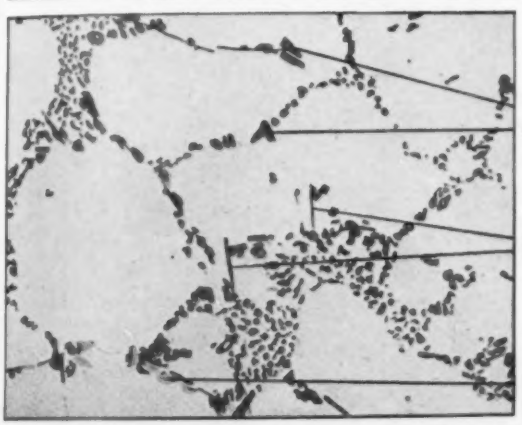


Fig. 17 — Insufficient solution heat treated condition. The solution heat treatment has not been conducted at a high enough temperature or for a sufficiently long time to dissolve all the $CuAl_2$ and round the silicon particles as in Fig. 16.

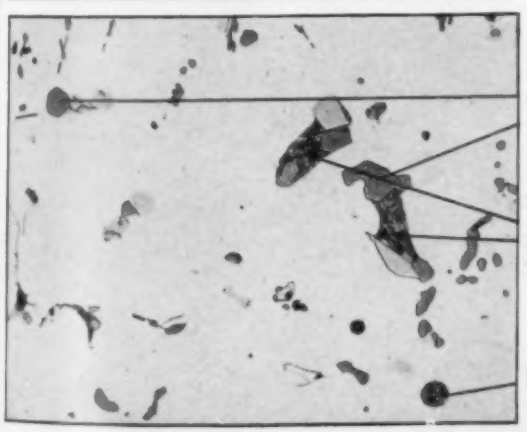


Fig. 18 — Overheated or "burned" condition. Heat treatment has been carried out at too high a temperature. Some material has actually melted or "burned" as shown by the irregular and round (rosette) deposits of eutectic material. The silicon particles have markedly coarsened, which is another sign of overheating.

0.5 per cent HF etch 250X

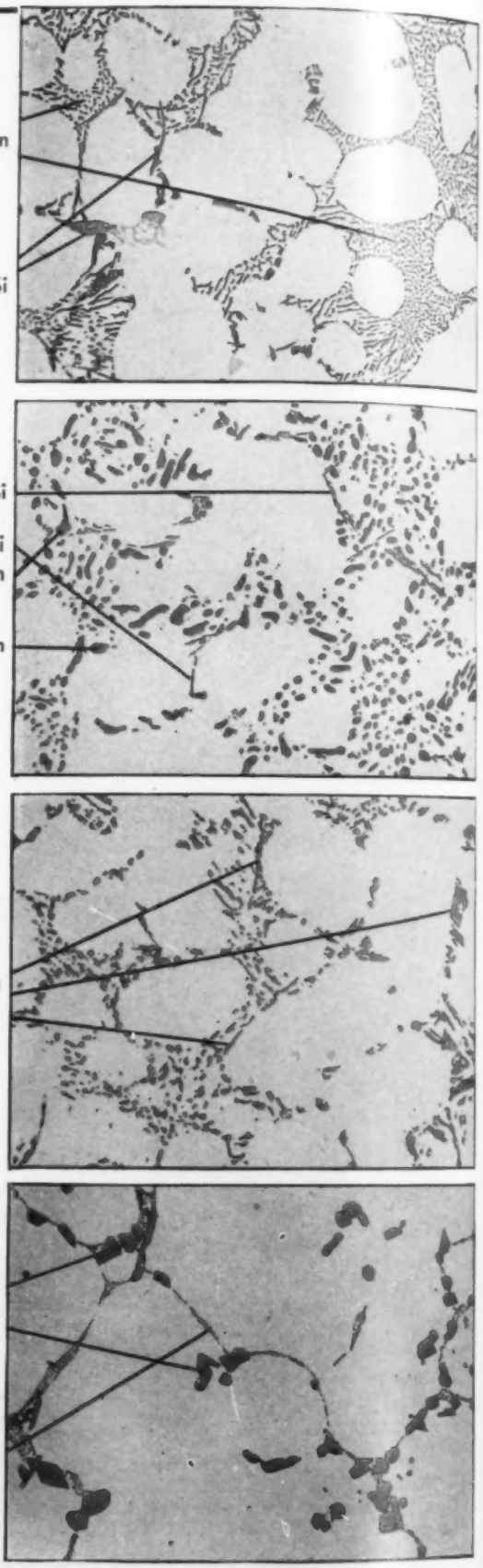
Representative Photomicrographs of No. 355 Aluminum Alloy.

Fig. 19 — "As-cast" condition. The fine aluminum-silicon eutectic is the major constituent of this alloy. Note that the silicon particles in the eutectic appear to be long and narrow for the most part. Any iron present combines with the silicon to form beta Al-Fe-Si.

Fig. 20—Normal solution heat treated condition. Solution heat treatment has rounded and agglomerated the silicon eutectic particles but has not affected the beta Al-Fe-Si constituent.

Fig. 21 — Insufficient solution heat treated condition. Heat treatment has not been carried on long enough or at a sufficiently high temperature, as is indicated by the presence of some silicon particles which have remained long, narrow and angular.

Fig. 22 — Overheated or "burned" condition. Heat treatment has been carried out at too high a temperature. Actual melting or "burning" has occurred, as is shown by the heavy eutectic deposits along the grain boundaries. Note how the silicon particles have enlarged — another sign of overheating.



0.5 per cent HF etch 250X

Representative Photomicrographs of No. 356 Aluminum Alloy.

The structural changes involved in precipitation alter the mechanical and some of the physical properties of the alloy. A typical course of precipitation at a specific temperature involves a relatively rapid increase in hardness to a maximum, followed by a decrease at a moderate rate, and then further decrease at a progressively reducing rate for an extended period of time. In general, the tensile and yield strengths vary directly and the elongation inversely with hardness. The effect of such precipitation on certain physical properties will be discussed later.

Composition and Hardness

Maximum hardness is obtained by heating for long periods of time at slightly elevated temperatures. The specific time and temperature necessary for maximum hardness usually vary with alloy composition. At higher precipitation temperatures, the heating time required to attain "peak" hardness decreases with increases in temperature. However, "peak" hardness values become progressively lower as the precipitation treatment temperature is increased.

In the initial stage, the precipitate formed is too fine for resolution under a microscope. Its presence is inferred from indirect evidence such as change in mechanical properties and etching characteristics. The particles coalesce as the time or temperature is increased, a comparatively small increase in temperature being much more effective than a large increase in time. A treatment that has coalesced the precipitate into particles large enough to be visible under the microscope usually has carried the material beyond "peak" hardness.

Artificial Aging Treatments. In production terminology, precipitation hardening treatments are known as aging treatments, and are further referred to as room temperature aging or artificial aging, depending upon whether they are carried out at room temperature or elevated temperature. Reference was made previously to the three types of artificial aging treatments for aluminum casting alloys that are of commercial importance.

Artificial aging only may be applied to castings in the "as-cast" condition. These treatments are effective because, as previously mentioned, most castings cool rapidly enough in production to retain a considerable degree of supersaturation of solid solution at room temperature. Low temperature artificial aging treatments of this type increase the tensile and yield strengths and the hardness, with some sacrifice in elongation. These changes in mechanical properties afford an improvement in machining characteristics.

Intermediate Temperatures

Aging treatments at intermediate temperatures also may be applied to "as-cast" castings that are to be operated at elevated temperatures and in which even small permanent dimensional changes would be objectionable. Density changes usually accompany structure alterations involved in precipitation. The magnitude of these density or dimensional changes, which varies with alloy composition, is not great but must be taken into consideration in some casting applications. Use of such aging treatments causes growth to occur in the castings and, consequently, prevents permanent dimensional change later during operation at elevated temperatures.

A second type of aging treatment is the low temperature treatment used after a solution heat treatment. In some cases, the aging portion of these treatments is designed to produce maximum hardness and the

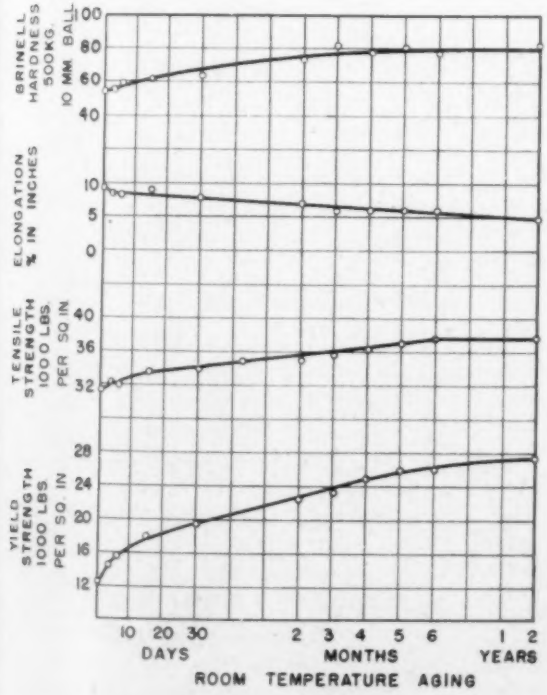


Fig. 23—Effect of room temperature aging on the mechanical properties of solution heat treated sand cast No. 195 aluminum alloy standard ½-in. diameter test specimens.

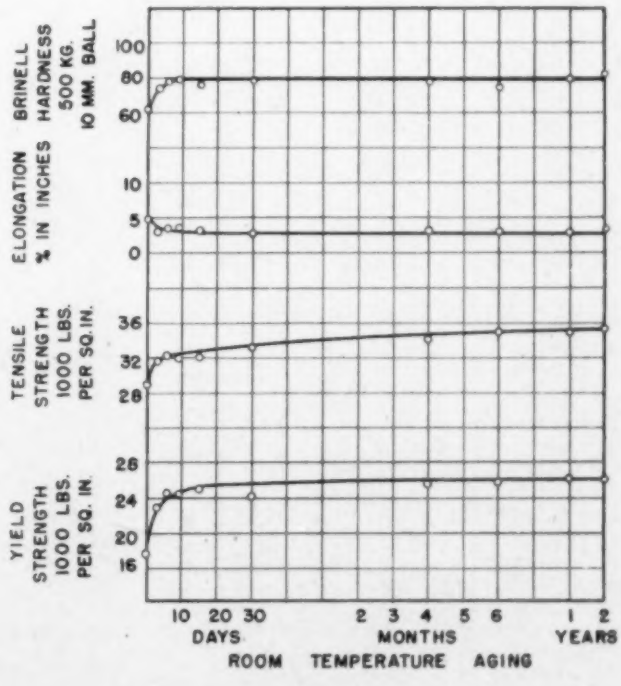


Fig. 24—Effect of room temperature aging on the mechanical properties of solution heat treated sand cast No. 355 aluminum alloy standard ½-in. diameter test specimens.

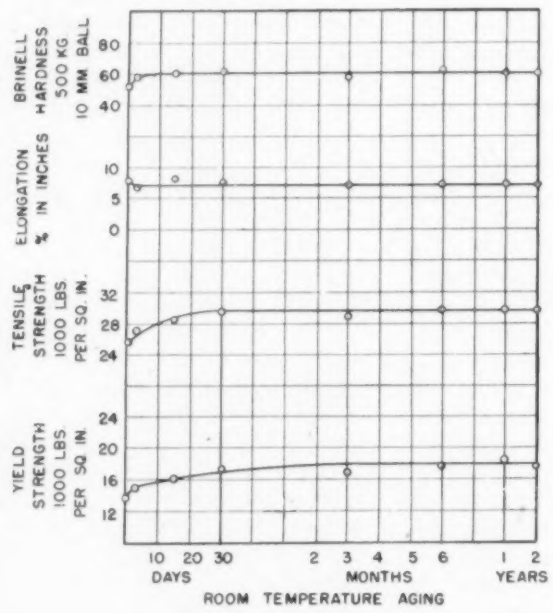


Fig. 25—Effect of room temperature aging on the mechanical properties of solution heat treated sand cast No. 356 aluminum alloy standard ½-in. diameter test specimens.

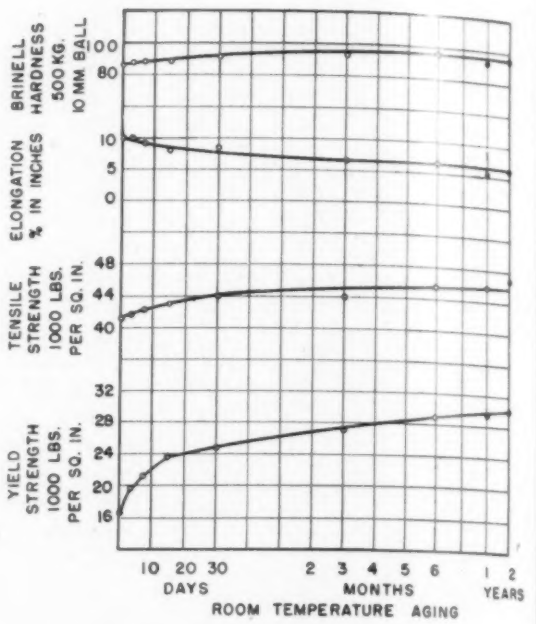


Fig. 26—Effect of room temperature aging on the mechanical properties of solution heat treated permanent mold No. B195 aluminum alloy standard ½-in. diameter test specimens.

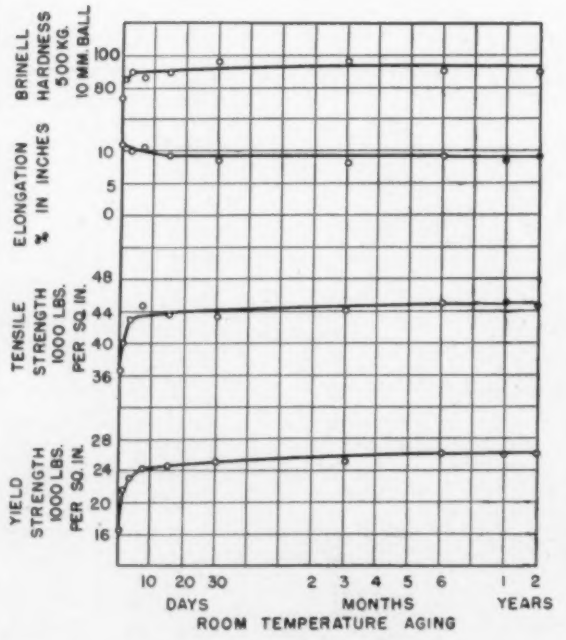


Fig. 27—Effect of room temperature aging on the mechanical properties of solution heat treated permanent mold No. 355 aluminum alloy standard ½-in. diameter test specimens.

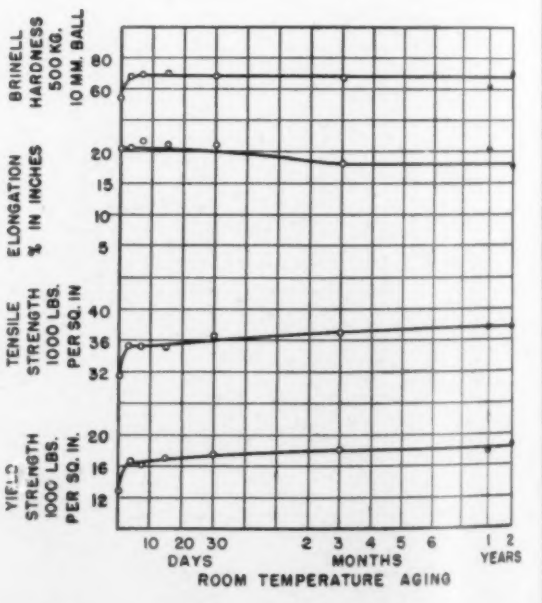


Fig. 28—Effect of room temperature aging on the mechanical properties of solution heat treated permanent mold No. 356 aluminum alloy standard ½-in. diameter test specimens.

attendant higher tensile and yield strengths and lower elongation. Other aging treatments effect a compromise, providing somewhat lower tensile and yield strengths and hardness than are obtained with the foregoing treatments in order to retain higher elongations.

A third type of aging treatment is an intermediate temperature treatment following a solution treatment. The higher aging temperatures that are used produce an overaged material. As the term implies, overaging results in precipitation in excess of that which imparts maximum hardness and tensile strength. The objective of these treatments is to procure substantial freedom from growth in castings that are to be operated at elevated temperatures, and substantial removal of any internal stresses.

Density changes or growth produced by aging was discussed in connection with the first type of aging treatment. Residual internal stresses may be induced during quenching by excessively rapid cooling from the solution treating temperature. Presence of such internal stresses in castings may cause distortion during machining and, in special cases, might reduce the load carrying capacity of the part. These stresses can be minimized by adherence to the proper quenching technique. They can be further reduced by use of a suitable aging treatment.

Corrosion Resistance

Some of the physical properties of aluminum casting alloys are a function of their thermal condition. The resistance to corrosion of wrought products of certain aluminum alloys is influenced by heat treating practices[8].

In general, differences in heat treatment or heat treating practices do not alter the resistance to corrosion of aluminum alloy castings to a degree that is of importance commercially. The thermal and electrical conductiveness of aluminum castings are materially influenced by heat treatment. Conductivity data on commercial casting alloys in different thermal conditions are available[9] for use in considering applications in which these physical properties are important.

Room Temperature Aging. Reference was made, in discussing precipitation treatment, to the fact that an unstable structural condition exists in solution treated aluminum alloy castings. It was also pointed out that the rate of precipitation occurring in adjustment toward equilibrium conditions is comparatively slow at room temperature. The aging that occurs in castings at room temperature usually is most marked in the first few days after quenching from the solution heat treating temperature. It proceeds at a slower rate over a long period of time.

Curves which show the effect of room temperature aging on the mechanical properties of several solution treated commercial aluminum casting alloys are included in Figs. 23 through 28.

Some of the treatments and alloys described in the foregoing are the subject of United States patents held by the company with which the authors are associated.

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3. Cast Metals Handbook, American Foundrymen's Association (1944).

4. Recommended Practices for the Sand Casting of Non-ferrous Alloys, American Foundrymen's Association (1944).

5. *Alcoa Aluminum and Its Alloys,* Aluminum Company of America.

6. *Aluminum Casting Alloys,* Aluminum Company of America.

7. P. T. Stroup, "Atmospheric Control in the Heat Treatment of Aluminum Products," *Controlled Atmospheres,* ASM, p. 207 (1942).

8. F. Keller and R. H. Brown, "The Heat Treatment of 24S and Alclad 24S Alloy Products," Alcoa Technical Paper no. 9.

9. L. W. Kempf, C. S. Smith and C. S. Taylor, "Thermal and Electrical Conductivities of Aluminum Alloys," *Transactions,* AIME, vol. 124, p. 287 (1937).

DISCUSSION

Chairman: Walter Bonsack, National Smelting Co., Cleveland.

Co-Chairman: L. W. Eastwood, Battelle Memorial Institute, Columbus, Ohio.

Member: How much difference is there in mechanical properties when aging the as-cast castings and when annealing them?

Mr. Sicha: This aging or precipitation treatment performed on castings in the as-cast condition has a relatively small effect on the mechanical properties. The effect is so small as to be of relatively little commercial interest from the standpoint of attaining improvement of mechanical properties on sand castings.

It has more application on permanent mold castings and there it is performed largely with the object of changing the properties to a small degree, that is, obtaining some hardening and from that obtaining some improvement in machining characteristics. That is really its principal value.

As far as the annealing treatment is concerned, which is performed at an intermediate temperature, the mechanical properties are generally of the order of the as-cast material, although they are slightly lower in some cases, depending on alloy composition. The yield strengths may be materially reduced by annealing.

Member: I believe Mr. Sicha mentioned the fact that the smaller the macrograins, the more efficient the solution of the "B" components. Will you explain why an increase in physical properties is not obtained when adding boron to the molten metal as a grain refiner?

Mr. Sicha: The question was; with grain refining additions of boron, why is there not a greater gain in mechanical properties attained, in the light of what I said about the desirability of having fine macrograin.

That is probably explainable on the basis of heat treatment provided to the material without the boron addition. With a finer macrograin and with a fine dendritic structure, it is possible, theoretically, and in general practice, to use a shorter solution heat treating time. However, with a material that has somewhat coarser grain structure or coarser dendritic structure than is obtained when a grain refiner is added, excellent mechanical properties frequently can be obtained if sufficient solution time is used.

W. E. Martin[1]: I believe the authors made the statement that treatment at elevated temperatures produces permanent growth.

Mr. Sicha: Aging at certain elevated temperatures can be used to attain permanent growth.

Mr. Martin: I want to mention a minor exception to that statement. The alloys in the aluminum-zinc-magnesium system are not subject to such growth.

Mr. Sicha: Apparently I did not amplify some of my statements sufficiently to make clear, as has been pointed out by Mr. Martin, that growth does not occur in some aluminum alloys as a result of heating.

[1]National Smelting Co., Cleveland.

GRAY IRON CASTINGS
SECTION SENSITIVITY

Henry C. Winte
Worthington Pump & Machinery Corp.
Buffalo, N. Y.

MOST FOUNDRYMEN are familiar with the fact that cast irons of low carbon and silicon contents must be used to obtain high strength, close grained metal in heavy sections. Through experience, they have learned that section size is a variable which affects the tensile strength and hardness of cast irons.

Experimental data published in recent years show that the physical properties of cast iron in test bars are decreased by increasing the section size of the bars. The loss in strength indicates the section sensitivity of the iron. The effect of the total carbon plus ⅓ Si ratio (carbon equivalent) upon the strength and hardness of cast iron as the section size increases has been investigated.

Section Size

In increasing section size, the factor which causes a decrease in the physical properties is the decreased cooling rate. Cooling rates depend upon a number of factors, such as temperature of the iron at pouring, rate of pouring, volume of metal to be cooled, metal surface area for transfer of heat to the sand, conductivity of the mold material, cores, energy releases, number of castings in the mold, amount of sand surrounding the casting, and the position of gates and risers.

The rate at which the metal cools in the mold affects the amount and size of the graphite flakes, the pearlite grain size, and the structural condition of the matrix. With slower cooling, more time is allowed for the graphite to precipitate and grow larger, thus causing a decrease in strength because of the weakness of the graphite. Also, slower cooling allows more pearlite to break down into ferrite, thereby decreasing the strength because ferrite is weaker than pearlite.

Iron Strength

The strength of the iron in a casting will be the same as that of a certain test bar only when the cooling rate of the casting is the same as that of the test bar. The graph in Fig. 1 shows the relation between section size and tensile strength of cast iron. Selection of the proper iron to meet design requirements is of basic importance.

Design engineers may accomplish this by use of the data graphically shown in Fig. 1. For example, a one in. diameter test bar may not have the same cooling rate as a casting having a one in. thick section.

Physical properties of a casting may be predicted from a test bar only when the cooling rate of the test bar is comparable to that of the casting. The relation between section size and tensile strength has been worked out in graphic form, enabling the engineer to select an iron which will meet design requirements of the casting.

Strength of Castings

Mathematical formulas have been developed for predicting the strength of castings, but they are applicable only to simple shapes. According to Bolton*, the volume of molten metal in the casting is proportional to the heat energy to be dissipated. The surface area of the casting is the exposed area from which the heat must be dissipated. Therefore, the cooling rate is directly related to the ratio of the volume in cu. in. to surface area in sq. in. $\frac{V}{A}$. The ratios for a number of bars and plates are shown in Table 1.

It is apparent from the table that a ½ in. diameter test bar will have the same cooling rate as a ¼-in. plate, and a 6 in. diameter test bar the same as a plate 4½ in. thick. In order to predict the strength of a casting it is evidently erroneous to select a test bar having a diameter about the same as that of the casting wall thickness.

In actual practice, casting shapes are seldom simple. In most cases, castings are round or square and contain cores to form the shape of the cavity. When these molds are poured, the core, being almost completely surrounded by metal, will be heated quicky by the molten metal and will retain the heat, since it has no avenue of escape except through the casting as it cools. This process

Presented at a Symposium on Engineering Properties of Gray Iron Session of the Fiftieth Annual Meeting, American Foundrymen's Association, at Cleveland, May 8, 1946.

*John W. Bolton, *Gray Cast Iron*, Penton Publishing Co., Cleveland.

Table 1

RELATION OF COOLING RATE TO VOLUME-AREA RATIO

	Casting Size, in.					$\frac{V}{A}$ Ratio
	Diameter	Square	Thickness	Length	Width	
Bar	½			21		0.122
Bar		½		21		0.123
Plate			¼	12	12	0.120
Bar	1.2			21		0.290
Bar		1.2		21		0.290
Plate			⅝	12	12	0.286
Bar	2			22		0.488
Bar		2		22		0.476
Plate			1⅛	12	12	0.478
Bar	4			18		0.901
Bar		4		18		0.905
Plate			2 9/16	12	12	0.889
Bar	6			18		1.290
Bar		6		18		1.285
Plate			4½	12	12	1.285

greatly retards the cooling rate of the casting.

Three castings, poured of the same cast iron, were selected from actual practice to compare with test bars of various diameters. These castings were pistons having internal cores, but varying in wall thickness and mass. The engineer specified ASTM Class 35 as the cast iron to be used, because an iron having a tensile strength of 35,000 psi. was desired in the finished castings. The foundry estimated that the ASTM Class 35 iron would not give the required tensile strength in the casting and poured from an ASTM Class 45 cast iron.

Figure 2 illustrates the effect of decreased cooling rate on the tensile strength and Brinell hardness of the Class 45 cast iron. By increasing the section size from ½-in. to 6-in. diameter, the tensile strength and Brinell hardness decreased 57 per cent.

Figure 3 illustrates the effect of decreased cooling rate on the graphitic structure. The graphite increases in amount and size with a corresponding decrease in physical properties.

The effect of the decreased cooling rate on the structure of the matrix is shown in Fig. 4. The pearlite grows coarser until finally some free ferrite is precipitated, causing the strength and hardness to decrease.

Piston Specifications

A diesel engine piston, shown in Fig. 5, weighs 980 lb. and five are cast in a dry sand mold. The sections are ½ in. except the wrist pin bosses, ribs and top, which are approximately one in. thick. The piston is cast with ¼-in. finish. A drawing of this casting is shown in Fig. 6. The metal has a tensile strength of 34,100 psi. and Bhn. of 185 at the position shown in Fig. 6.

Although the sections vary from ¾ to 1¼ in., the actual cooling rate is comparable to a 2½ in. diameter test bar. Referring to Fig. 2, a tensile strength of 34,100 psi. is seen to correspond to a cooling rate corresponding with a 2½ in. diameter test bar. This fact is further confirmed by the Brinell hardness. The actual Brinell hardness

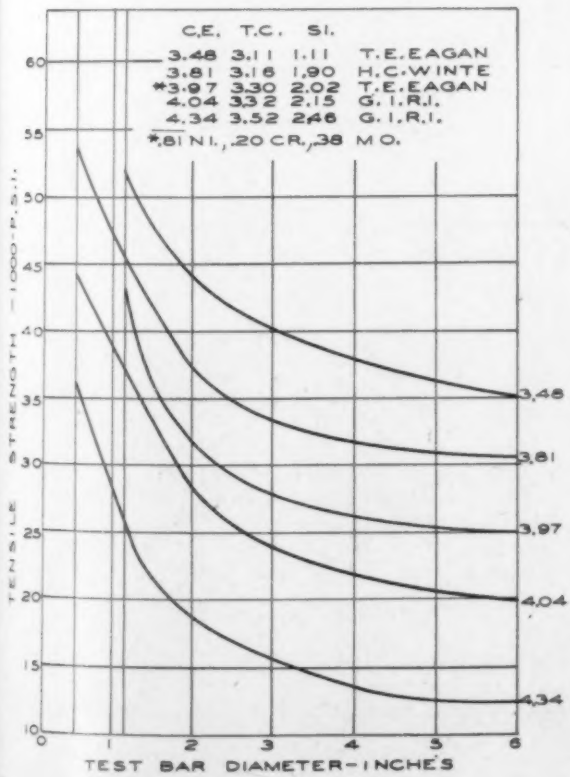
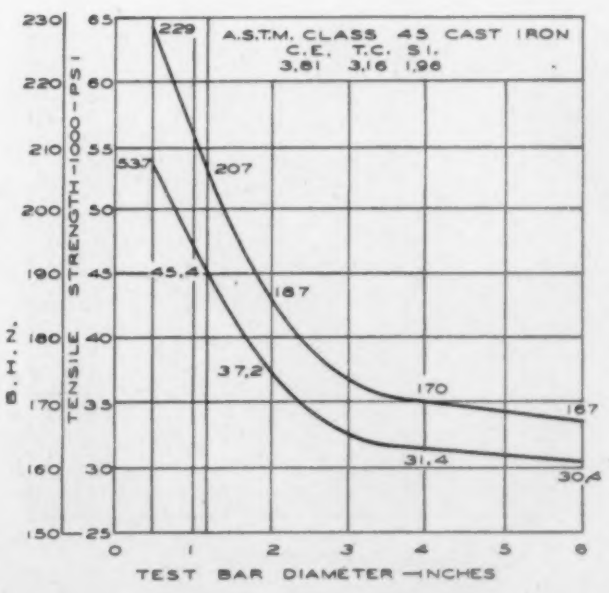
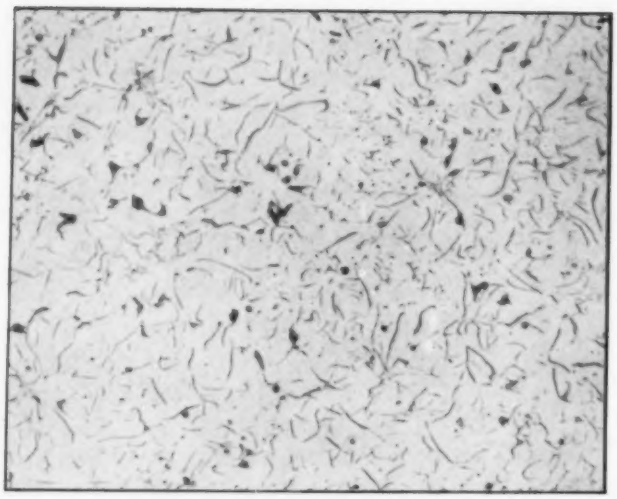


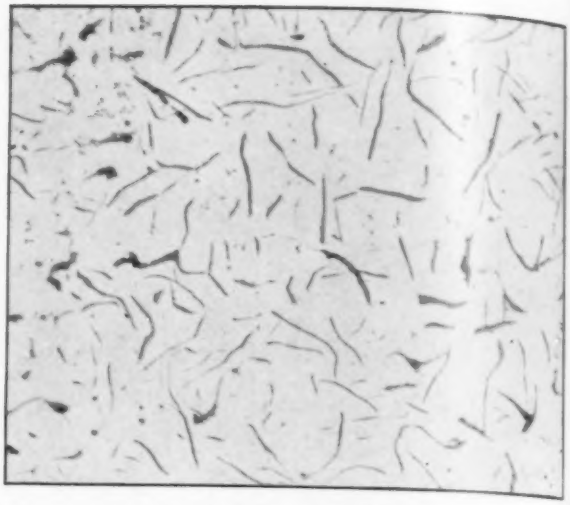
Fig. 1 (left)—Graph showing effect of section size on tensile strength of cast iron in test bars.

Fig. 2 (below)—Effect of section size on tensile strength and hardness of ASTM Class 45 cast iron.

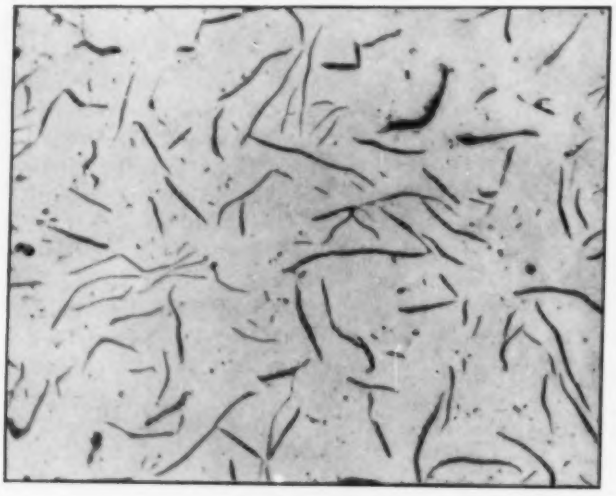




½ in. Diameter Bar



1.2 in. Diameter Bar



2 in. Diameter Bar



4 in. Diameter Bar



6 in. Diameter Bar

Fig. 3—Photomicrographs showing effect of decreased cooling rate upon graphite in cast iron test bars of various diameters. Unetched. 100X.



½ in. Diameter Bar



1.2 in. Diameter Bar



2 in. Diameter Bar



4 in. Diameter Bar

Fig. 4—Photomicrographs showing effect of decreased cooling rate upon pearlitic structure of cast iron test bars of various diameters. Etched. 1000X.



6 in. Diameter Bar

Table 2

CHEMICAL ANALYSES AND PHYSICAL PROPERTIES

Casting	Carbon Equivalent	T.C.	Si	Tensile Section, in.	Tensile Strength, psi.	Bhn.	Casting Weight, lb.
Diesel Piston	3.72	3.00	2.13	¾	34,100	185	980
Ammonia Compressor Piston	3.84	3.18	1.98	7/16	33,475	192	30
Air Compressor Piston	3.71	3.17	1.67	3¼ boss	32,760	179	160
				⅞ boss	30,875	179	

of 185 is close to the indicated hardness of 180 in the test bar.

An ammonia compressor piston, shown in Fig. 7, weighs 30 lb. and four are cast in a green sand mold. The section size varies from 5/16 to ⅝ in., with ⅛-in. finish. A drawing of this piston is shown in Fig. 8. A tensile bar was cut from a 7/16-in. section, as shown in Fig. 8. The tensile strength was 33,475 psi. and Brinell hardness 192.

The actual cooling rate of the casting is comparable to that of a 2½-in. bar since, as illustrated in Fig. 2, the 33,475 psi. tensile strength is comparable to the cooling rate of a 2½ in. diameter test bar. Also, the Brinell hardness of 192 again compares favorably with the indicated figure of 180. Figure 9 is a photomicrograph of the test piece, which also indicates a slower cooling rate than that of a 7/16 in. diameter test bar.

An air compressor piston, shown in Fig. 10, weighs 160 lb. and is cast singly in a green sand mold. The section size varies from ⅝ in. to 3¼ in. at the center boss for the piston rod. This boss is cast solid, as illustrated in Fig. 11. Tensile specimens were taken from the ⅞-in. section, and the 3¼-in. boss, as shown in Fig. 11. One-quarter in. is allowed for finish on the casting.

The ⅝-in. section had a tensile strength of 30,875 psi. and 179 Brinell hardness; the 3¼-in. section had a tensile strength of 32,760 psi. and 179 Brinell hardness. The results are comparable since they are within the limits of error when testing a small-diameter specimen.



Fig. 5 (above)—Diesel engine piston casting.

Fig. 6 (below)—Cross-sectional drawing of deisel engine piston.

The cooling rate of the casting was comparable to that of a 3½ in. diameter test bar for, as shown in Fig. 2, the tensile strength is approximately that of a 3½ in. diameter test bar. This cooling rate is also substantiated by the actual Brinell hardness. The piston had a hardness of 179 as compared with a Brinell hardness of 173 in the 3½ in. diameter bars. The fact that the ⅝-in. section had a slower cooling rate than a test bar of the same size is also indicated in Fig. 12 by the size of the graphite and pearlite structures.

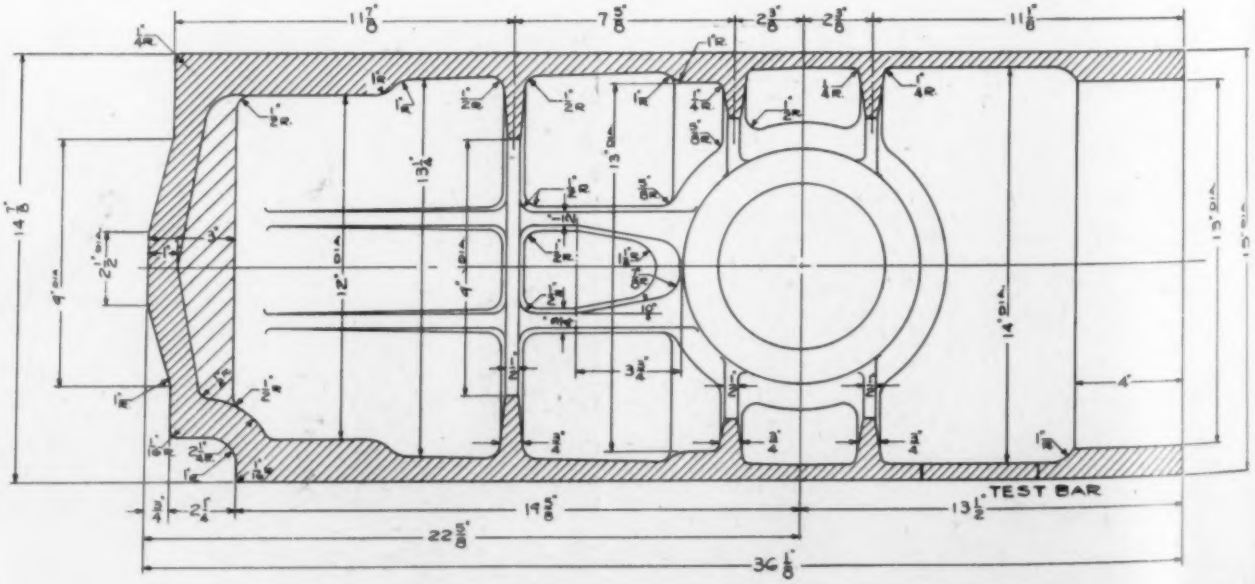
Table 2 shows the chemical analyses and physical properties of each of the castings. Although there is some variation in the chemical analysis, the T.C. plus ⅓ Si (carbon equivalent) is within the specification limits for this iron.

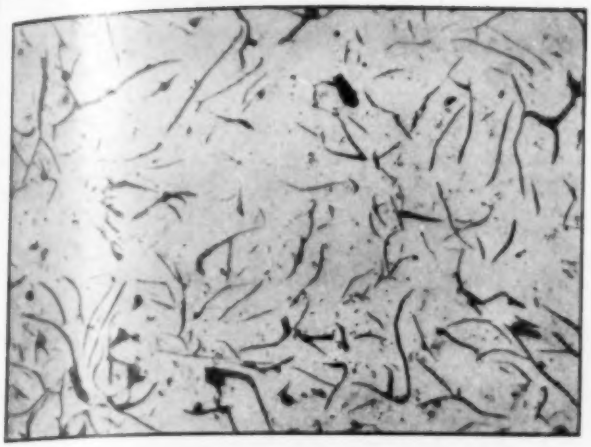
Conclusions

1. A test bar can be used to predict the physical properties of a casting only when the cooling rate of the test bar is comparable to the cooling rate of the casting.

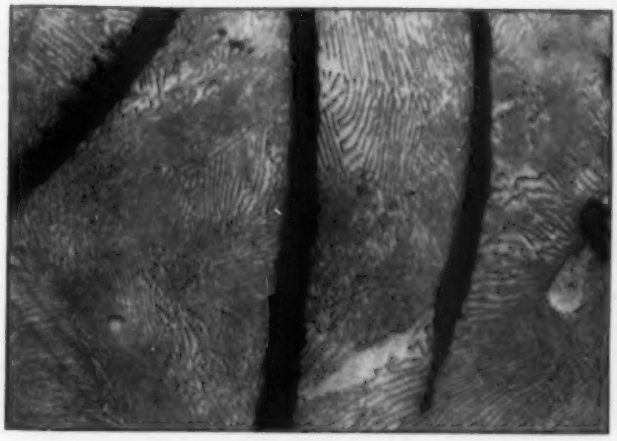
2. To predict the strength of a given casting, the volume-area ratio can be used to compare the cooling rate of a test bar and the casting.

3. The cooling rate of castings is affected by internal cores. Under these conditions, it is not possible





Graphite, unetched. 100X.



Pearlite, etched. 1000X.

Fig. 9—Photomicrographs showing graphite and pearlitic structure of cast iron used for ammonia compressor piston.

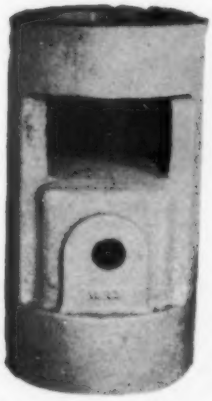
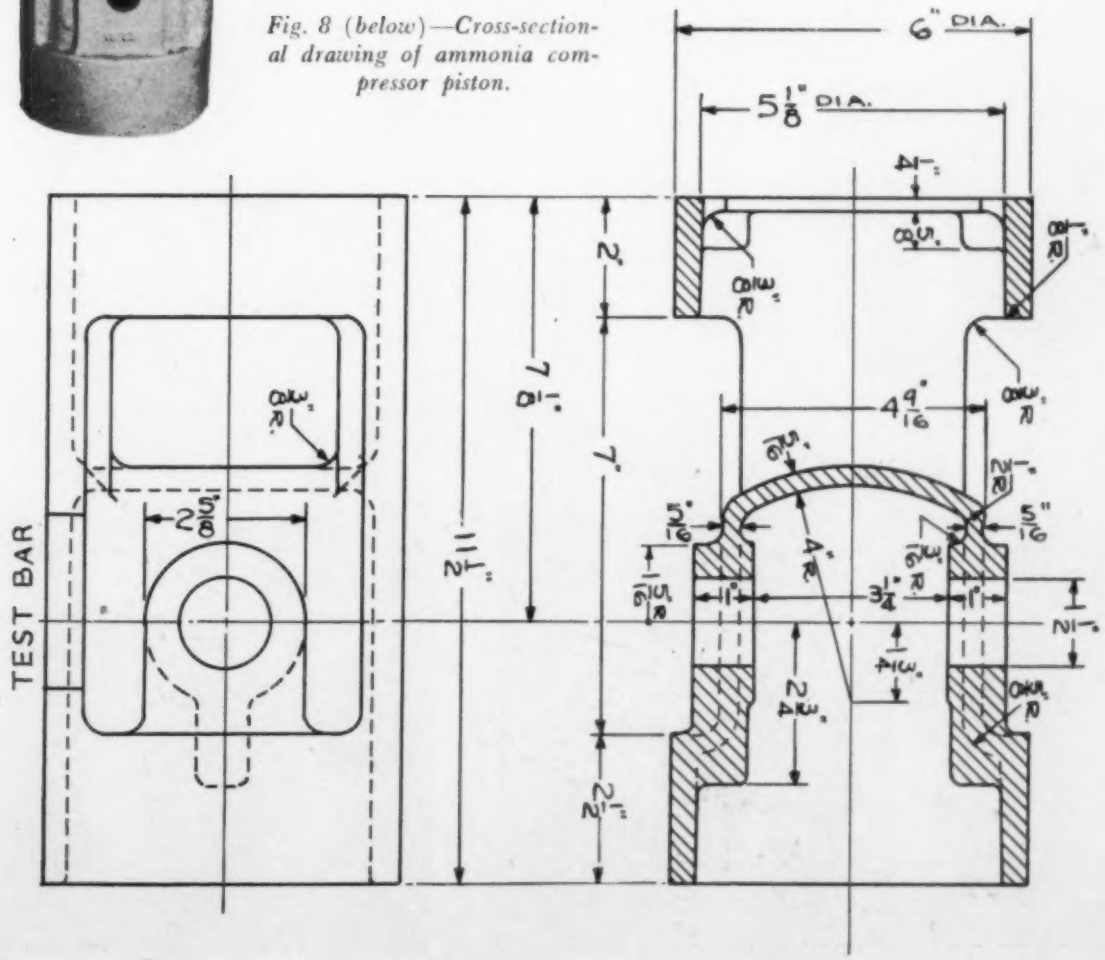


Fig. 7 (left)—Ammonia compressor piston casting.

Fig. 8 (below)—Cross-sectional drawing of ammonia compressor piston.



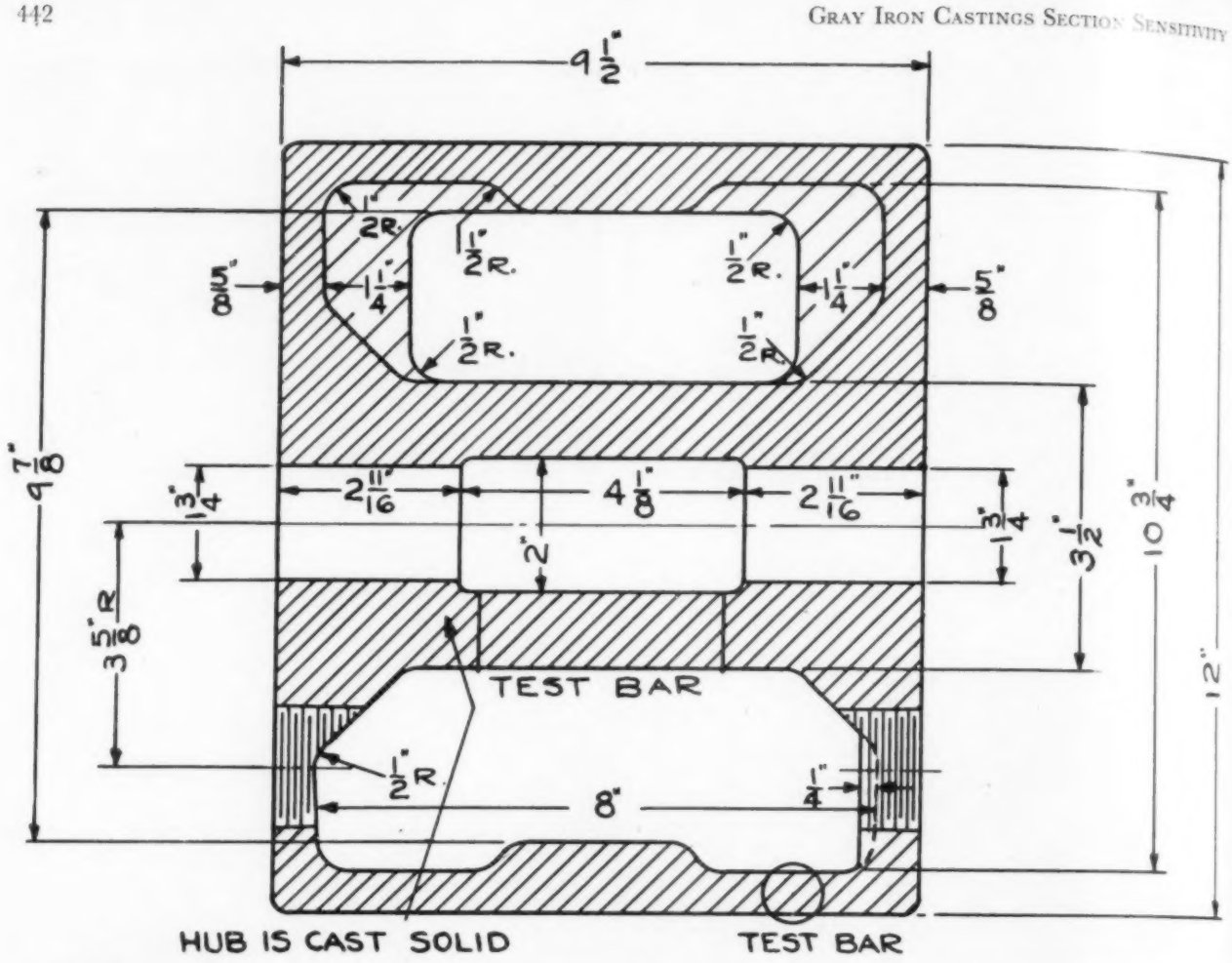


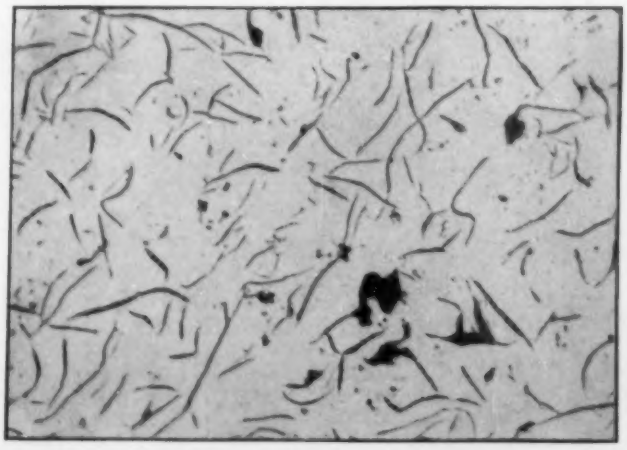
Fig. 11 (above)—Cross-sectional drawing of air compressor piston.



Fig. 10 (left)—Air compressor piston casting.

Fig. 12—Photomicrographs showing graphite and pearlitic structure of cast iron used for air compressor piston.

Graphite, unetched. 100X.



Pearlite, etched. 1000X.



to predict the casting strength. A close estimate is possible by using either a 2 in. or 3 in. diameter test bar, or an appropriate precentage of the strength of the 1.2 in. diameter test bar.

4. Foundries should investigate the effect of section size on each iron poured so as to establish its section sensitivity.

Acknowledgments

Grateful acknowledgment is made to the following men for their welcome suggestions and cooperation in the preparation of this paper: J. C. DeHaven and D. E. Krause, Gray Iron Research Institute, Columbus, Ohio; T. E. Eagan, Cooper-Bessemer Corp., Grove City, Pa.; R. J. Allen, Worthington Pump & Machinery Corp., Harrison, N. J.; R. G. McElwee, Vanadium Corp. of America, Detroit; and Fred J. Walls, International Nickel Co., Detroit.

DISCUSSION

Chairman: R. G. McELWEE, Vanadium Corp. of America, Detroit, Michigan.

Co-chairman: W. E. MAHIN, Armour Research Foundation, Illinois Institute of Technology, Chicago, Illinois.

CHAIRMAN MCELWEE: Data in this paper checks some work that I have seen. I know of cases where arbitrary sizes of test bars have been established by experimentation to fit conditions of a peculiar type of casting so that the test part can be a pilot. I also know of test bars of add dimensions, such as 1½ in., being used because in a particular job which is a long run job, it had been found that it was necessary to establish that type of test bar to fit the casting under consideration. Of course, where you make single castings you do not have the opportunity to do that.

This paper is worth while because it demonstrates that these relationships do have some tendency to follow the laws that you would expect them to and that you can use test bars if you use them judiciously for control methods.

WHAT MANAGEMENT SHOULD KNOW ABOUT TIMESTUDY

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IN ANY SURVEY of timestudy and wage incentive methods, dozens of differences will be found. Some of these inconsistencies result from unlike concepts. Some are caused by negligence.

Either of two groups is at fault according to which way we look at the cause. Timestudy men may be at fault for not insisting upon being allowed to do their work correctly. Management may be to blame for not hiring the right kind of men. However, fixing the responsibility does not solve the problem.

Whether we have an old incentive plan that has become sadly dilapidated or are about to start a new one, there are certain sound fundamentals which must be understood before we can expect to get the best over-all results. Management should understand these fundamentals because it is responsible for the success of the enterprise. Management should know the difference between right and wrong, whether it is hiring timestudy men for the work or using its own men.

One of the first things to be straightened out is the difference between incentive and base wage. These two have been mixed until many people cannot see a distinction. In part, this results from long use of piecework with monetary rates. Also, it comes from the extensive use of the low base rate-loose standard incentive plans. Both types have given consideration to compensation primarily without regard for work measurement.

Much of the mixed thinking carries over from the past wherein incentives of many types have been "figured out" from past performances. These abortions are at the root of most of the negative reactions to wage incentives. They were the chief reasons why we had "rate cutting." There are other reasons, of course, which in the main have to do with poor timestudy. These will be developed later.

But the point is that there is no uniform measure of work done when the incentive base varies between plants. There is no standard definition of a fair day's work. The yardstick has different lengths and the differences in length greatly exceed the variations between the several patternshop shrink rules. Looked at from one side and assuming comparable base rates, the differences are reflected in unlike earnings, and hence, costs. From the other side, the earned rates may be comparable and the differences in work standards will then be shown as differences in base rates.

Here is where we should begin to take stock. It is apparent that two general trends are working on base wages. One is moving all rates upward. The other is making base rates consistent for a company and even for an industry. Both trends disregard differences in standards established for wage incentives.

Sound incentives result in higher take-home wages, lower costs and increased productions. For as small an investment of time and money, these benefits cannot be obtained by any other means now available. It is good business to give serious attention to sound timestudy -wage incentive administration. Management should endeavor to be as soundly practical in this field as it considers itself to be in other parts of business administration. In so doing, it will directly improve its employee standard of living and better the operations of the business as a whole.

Revising Standards

When previously set incentive standards have been made loose to compensate for low base rates, these two trends wipe out some portion of the wage differential. Then, unless the standards have been revised accordingly, the cost includes two payments for the same thing. It includes the cost of the raise and the cost of the extra time in the standard which was allowed previously to offset the low base rate. The extra time is equivalent to an increase in base rate and applies as a percentage multiplier to every increase added to the hourly rate.

So it happens that those plants which set loose standards to make

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up for low base rates will have excessive costs because the wage scale is moving up independently of the time allowances. The higher the wage rates go, the greater will be the distortion.

Normal Output. Many reasons make it obvious that there is little likelihood of base rates being the same universally for the same job. However, there seems to be plenty of evidence pointing toward a growing similarity, as time goes on. Therefore, it appears increasingly important to get uniformity in the task level or work measurement standard. Without uniformity in the measure of a day's work, the correction of inequalities in wage rates will magnify those inequalities and throw them over into greater differences in "take home."

The solution to the problem lies in a common understanding of a normal operator performance. Normal is not average. Average is highly variable and changing. Average is a term which should be crossed out of the timestudy language and particularly omitted from contract clauses purporting to define the basis for wage incentives.

True Measurement

It is necessary to have a "yardstick" which is constant—a measure which does not change from one department or plant to the next. The yardstick should be the same in good times and poor. The reason is that there is but one fair time to perform an operation under a specific set of working conditions.

This can be attained only by having a bench mark or common base of reference. Such it is hoped will result from the work of the National Rating Committee of the Society for the Advancement of Management. It is planned to produce motion-picture films of typical operations at known rates of operator performances. These will be standards because they can be shown at identical projector speeds anywhere in the country, thus establishing a common basis for measuring work performance.

The Big Variable. To some managements and timestudy men effort rating means nothing because they do not know that fair standards cannot be set unless the performances observed are rated in terms of a "normal." They still think in terms of average times, which are unfair and inconsistent. Averages of poor performances produce loose standards. Averages of best operator times result in tight standards, and it is foolish to think that one can select a normal performance without being skilled in effort rating.

Rating

Rating is the appraisal of the actual pace at which the work is being done. Rating is vital to currect timestudies because the actual pace may be anything from 50 per cent to 200 per cent of the normal sought. It goes without saying that herein lies the largest variable in the whole timestudy procedure.

Consequently, there are two benefits to be derived from film standards of known performances. Most important, they can provide the basis for establishing a uniformity in task level so that with like base rates and like performances the "take home" will be practically the same.

Timestudy men can use such films as gauges to measure their ratings. Standardized films can be made to provide training for timestudy men and others in correctly judging the actual performances being studied. Thus work measurement and incentives could be brought to the same standard of expectancy and, as the author sees it, into the same range of uniformity as base wages for like work are approaching.

Beware of Simplification. With the same concept of normal performance, there is another type of mixture which distorts earnings. This is one for which management and accountants are largely to blame. Their thinking is in the right direction. Sometimes, it is in an attempt to meet a laudable need. However, often it is urged upon timestudy men in a mistaken desire to save overhead. The importance of simplifying the incentive plan, without fully recognizing the costs of simplification, is overstressed.

Delays

One of the most important fundamentals to be understood is that fair incentive plans cannot be simple. To be fair, allowance must be made for delays beyond the control of the operator. Delays should not be included in the standards because variations in the actual delays from the averages which might be allowed will directly affect operator earnings.

Separate allowances must also be made for differences between the actual and normal working conditions. Otherwise, the differences will affect take-home pay. Variations in the amount of work to be done and delay time which interferes with production are two complexities which cannot be averaged into the standards if the incentive plan is to be fair. Therefore, it should again be emphasized that incentive plans cannot be simple and, at the same time, be fair.

Piecework is a type of oversimplified incentive plan that is often mentioned as highly desirable because it is so easily understood. Of course, so much a piece is easy to explain because the transaction is exactly like that carried on in ordinary shopping. But so many dollars and cents per piece cannot be fair because it takes in a lot of averages which sometimes work in favor of the operator and at other times have the opposite effect.

Piecework

Then, too, piecework is always being subjected to compromises of many sorts, which means that it soon loses its relation to work done. Besides, every time there is a change in wage rates the piecework prices must be recalculated or adjusted by compound percentages. This latter "simplification" certainly contradicts the initial premise.

Another simplification which resembles piece rates is that worked out in terms of pounds or tons. In foundries and steel mills the basis called "tonnage rates" is often used. These simplified "rates" usually include gross inequities because they are based on averages.

Such simplified incentive applications have the obvious advantage of somewhat easier explanation and computation. But, in addition to being unfair, there are other factors to remember. In the first place, the ease of explanation at the time of introduction will soon be overshadowed by a perpetual stream of explanations of those conditions which adversely affect earnings. In many cases, additional allowances will be made. As a result, the worse-than-average conditions will be paid

for twice. Thus endeth the first lesson—do not use averages.

Oversimplification of an incentive plan may save some clerical costs, to be sure, but that is being penny-wise and pound-foolish. Simplification is effected by using averages for varying conditions and delay times. Averaging these expenses and including them in the standards may save the clerical cost of computation and distribution, but would hide wastes which, if known, would certainly be reduced and salable production substituted. Good, salable pieces can absorb not only the costly items of overhead but also the relatively insignificant cost of a few extra clerks.

Costs of the wastes continue until eliminated, no matter how they are concealed. A sound fundamental of good management requires first of all that the time facts be revealed. This will require more in clerical effort, but the additional cost can be offset by reducing production interferences.

Extra Items

Extra time or work must be paid for if the incentive plan is to be fair. These extra items should be at a minimum when management control is really effective. Under such conditions, there should be no reason for paying "average incentive earnings" for day-work time. However, if management gives way to pressure for this "fringe issue," it should be correctly analyzed as such and tacked onto the base rate.

"Can't Be Done." Suggested by what has gone before is another dreadful pitfall into which many managements tumble. There is no excuse for falling because it can be seen far in advance. It is fundamental that it is not the amount of earnings which cause the troubles, but the differences. It also is true that if incentive is good for some, it is good for all. It is known in advance, then, that incentive must be provided throughout the shop if in one department.

"It can't be done," say many otherwise capable executives. Some have been heard to insist that incentives could not be applied in their businesses, even when they knew of successful application in almost identical shops.

"It can't be done" is not the answer to the highly skilled man on day work who earns less than lower-rated operators on incentive. "It can't be done" is not the answer to the large proportions of indirect people in many plants where incentives have closed up the wage differentials. "It can't be done" will not satisfy the foremen who take home less than the people on incentives whom they supervise.

Incentive Opportunities

It can and must be done in order to provide incentive opportunities for all. And these should not be the "figured out" kind which arbitrarily pass on to the day workers some percentage increase calculated from the earnings of those on incentive. That too is an oversimplified method of granting what amounts simply to a pay increase.

It is easy to fall into these procedures. But they are treacherous. To begin with, there is no assurance of an increase in output commensurate with the greater pay. As a rule, there is not a direct relation between the efforts of indirect and direct workers. In general, the proper ratio of indirect to direct is a changing one, being high at low volumes and low at high volumes.

In the second place, and this is more important, such schemes do not have any record of the work content of the jobs involved. Consequently, there is no logical means for adjusting the incentive payments as the ever-changing methods make it necessary to alter the number of people assigned to the work.

Consequently, either unjustified accusations of "speed up" are aroused, or what is worse, payments of unearned premiums are continued. The author feels that this latter is the more detrimental because the unfairness grows with improvements to the point where the producers complain about having to work to earn their premiums while others can get as much by loafing. When this stage is reached, the whole setup is in jeopardy.

Poor Timestudy Men

It Costs Too Much. Many of these problems apparently cannot be solved because the timestudy men do not know how to measure work which is not routine. Here is another fundamental which management cannot afford to side-step. Management must recognize that poor timestudy men are the most expensive in the long run.

Skilled timestudy men are required to measure nonrepetitive operations. They must know how to make the maximum use of time-studies by the standard data method. This is where the difference lies. Of course, "it can't be done" by the ordinary "rate-setting" methods of direct timestudy. Not only does it cost too much, but direct timestudy cannot provide incentive until after the nonrepetitive job is completed.

If timestudy men cling to the traditions of continuous watch readings, average times, and perfecting the methods before setting the standards, they never will get around to complete coverage by incentives. The answer is to get men big enough to do the job of measuring out-of-the-ordinary types of work.

Bigger men will help in other ways. For example, the right type will not "go along" with some of the crimes perpetrated in the name of incentive. They will prefer to work in other plants. They will have more of a professional attitude toward their work and refuse to do things which may be expedient but incorrect.

Human Relations

Men who can do a good job and stick by their guns will not only solve many of the so-called problems of incentive but will be acceptable for promotion into executive positions. They will be broad enough to understand that sound timestudy work is largely human relations, not stop-watch engineering.

Standards Must Be Changed. Timestudy men who can do a good job will insist also that standards must be kept in line with work done. They cannot be induced to put off the revision of a standard after the method has been changed. They know that the longer it is postponed, the more difficult will be the justification when revised.

There is no escape from the ever present problem of maintaining fair standards. Failures in and of themselves usually do not cause as much trouble as the patchwork remedies applied in other directions in attempting to offset the odious con-

ditions created by the failures to correct for changed conditions. Two wrongs never made one right.

Incorrect and incomplete maintenance soon allows a soundly established incentive plan to become as inconsistent as one set up by poor timestudy. The earnings become irregular and inconsistent. Restriction of production usually sets in. Grievances pile up. Compromises bring on precedents and the spiral is started. When management thinks this has gone too far it may try to stem the tide by saying "No!" Then the cases go to arbitration.

Do Not Arbitrate Standards. Here a shortsighted management may think that arbitration is an appropriate way to solve the problem. Such is not true for several reasons. Arbitration usually ends in a compromise. But that is not measurement. Splitting the difference is not the way to determine incentive standards. Some part of the error continues. The real solution to the problem starts way back at the time methods were changed, when the standard should have been revised in strict conformity.

Correct Procedure

However, having thought that time would effect a cure and eventually having reached an impasse, the proper solution is not arbitration. The only correct procedure is to select a thoroughly competent timestudy man and engage him to determine what the standard should be. In this way the principle that standards should measure the work content of the job is not violated. Also, inconsistencies which "are binding on both parties" are not created.

Explaining Timestudy. There are many other things which management should know about timestudy. Many of these are properly cared for when management picks out high-grade men for timestudy work —those who have a professional attitude toward their jobs. For example, the problems which arise because the organization has not been instructed in the methods of timestudy are recognized and met.

Good timestudy men would not make the mistake of neglecting to inform the supervisors and union stewards. They would know that the industrial relations problems brought on by timestudy and wage incentives can best be overcome by anticipating them with ample explanations. The explanations should not be technical. The shop folks are not interested in techniques. They are more confusing than helpful.

Nevertheless, all questions must be answered satisfactorily. To do this requires that definite policies be established. The rules must be set down. If they are not definite, the explanations will be weak, of necessity, and the "cover up" will be suspected. Then there can be no confidence that the future will be different from the past. Confidence is an essential part of successful incentives. It must be thoroughly established and scrupulously maintained.

As a part of the program of instruction, two parts seem important enough to stress. First, precautions should be taken to see that new people are given the same benefits of understanding that were developed initially. If this is not done, then turnover will dilute the understanding to a point where anything can happen. An ounce of prevention should be given in the way of specific explanations at the time the new employee is inducted. A printed explanation included in the employee's handbook will be helpful.

In addition, the foreman should be obligated as a specific part of his responsibilities to again go over the full explanation of the incentive plan. This will improve the understanding by repetition, even if no other gain is made. But organizationally, it is necessary that the foreman make certain that there are no misunderstandings. This vital step in sound administration is completely overlooked in some organizations, and poorly done in many.

Incentive Plans Not Known

One of the reasons for this is that management has not realized how necessary it is and, therefore, has not made it a requirement. It has been side-stepped because foremen, in general, do not know enough about their incentive plans to make an understandable explanation. This a sorry condition and, until it is corrected, avoidance of industrial relations problems can hardly be expected.

As a stopgap, every foreman should be given extensive training courses in timestudy. And some day, if it is seriously desired to make incentive plans operate smoothly, promotions to foremanships will go only to those capable leaders who have successfully served an apprenticeship in timestudy work.

Why Bother. All of the foregoing makes a big job and requires a lot of hard work. Sound timestudy-wage incentives are no different from anything else worth while. And, experience has shown that there are many elements of business administration on which management spends more time and money without nearly as profitable a return.

Through sound incentives there are higher take-home wages, lower costs and increased productions to be gained. These benefits are not obtainable by any other means now available for as small an investment of time and money. Why not look squarely at the facts? Why not give more serious attention to sound timestudy-wage incentive administration because it is good business to do so? Let us stop dabbling in half-baked, half-hearted incentive schemes. Instead, let us work diligently to be as soundly practical in this field as we believe ourselves to be in other parts of business administration. In so doing, we will directly improve the standard of living of employees and better the operations of the business as a whole. Moreover, we will be doing something besides talk about inflation — something that will add real values to the national dollar.

DISCUSSION

Chairman: R. J. Fisher, Falk Corp., Milwaukee.

Co-Chairman: F. E. Wartgow, American Foundrymen's Association, Chicago.

W. E. George[1]: Mr. Carroll spoke of the existence of only one fair time standard. That time will vary with the incentive opportunity, will it not?

Mr. Carroll: That is correct, but the opportunity, as I tried to indicate in the beginning, seems to have been worked backwards in trying to adjust for a difference in a wage level, from what might ultimately be a more uniform wage level. In other words, there have been a number of plans established where, for example, the man was paid 40 cents an hour and was expected to make 40 cents more, because we wanted him to have an 80-cent take-home to be comparable with similar jobs in the area.

If now we pay him 80 cents for the same job, because they pay 90 cents across the street, then the level for the incentive must be made consistent with it.

Mr. George: The standard in that

[1]Booz, Allen & Hamilton, Chicago.

case would give you $1.80 if you raise the rate to 90 cents. I am speaking of the fact that you can put in the incentive plan to pay off 20 per cent for normal maximum output or you can put it in to pay 30 or 35 per cent.

In terms of Society for the Advancement of Management studies, will the pace you picture represent a pace that is one for the below normal or above normal?

MR. CARROLL: It will be a statistical summary of opinions throughout the country. I have suggested that such will be a 45-unit hour. I do not know what the answer will be. Irrespective of the level established from this, it would still be possible to set up a plan so that you would pay 20 or 30 or 70 per cent over.

MR. GEORGE: I believe that it is easier to establish the top than it is to get it to 45 or 50 or 60 per cent, where you want it.

MR. CARROLL: Yes, and I will agree with you that the top is established by the customer, and if we work back from that, we have got the top on the one hand. If your union establishes the bottom, on the other hand, I would like to suggest that there is a uniformity required, which may be in the nature of 25 per cent premium.

MR. GEORGE: I agree with the 25 per cent, we will say for the moment, but that will be a 64-unit hour, if you expected the employee to earn 80, rather than a 60-unit hour to pay base rate. In other words, in one plant the standard would be 60 units, in the other plant 64 units, if your normal top expectancy is an 80-unit hour.

MR. CARROLL: I do not think there is a top expectancy in a well-operated plan.

MR. GEORGE: I use the words "normal top" figuring there are always some who will exceed any standard.

MR. CARROLL: Yes, but I was endeavoring in this group to close some of the gaps between a 30-unit hour and a 60-unit hour, more than I was between an 80 or a 75-unit hour earning expectancy.

J. A. WESTOVER[3]: I am wondering if it would not be well to say something about expressing standards, either in money or in time. Too many timestudy men have the problem of trying to convince management that they should express their standards in time, and management seems to want to keep it simple, as they term it, by expressing it in money.

MR. CARROLL: You can not allocate overhead on a wage dollar basis and get the right answer. You can not plan production on a piece-work base and get anything that will work. If you attempt it, you must also operate a time basis, and for that reason, and in view of the ever-present tendency to change wage rates every time you rewrite a contract, I am not in favor of piece-work. The reasons for that are several.

First, we immediately mix compensation and measurement, and the timestudy man cannot solve the problems which result. In a great many situations he has no voice whatever in wage rate, because that was negotiated in the contract, and because the intent is the same on the worker's part as it is on yours and mine, to try to get all he can for as little as possible. He puts the pressure on the piece-work price, so we compromise it here and there, to satisfy the complaint.

Next year when we negotiate another contract we re-figure all of the piece-work prices or add on another one of these accounting simplifications, in which we figure from the old base, four changes back with compound percentages, and then wonder why the shop man cannot understand.

But my reasons are much more fundamental than that. Wage incentive, as I see it, is only a means to an end. The only purpose of wage incentive is adequately to reward the man for what he does and to sustain his interests in going along at a pace of operation that you can depend on when you make a quotation.

When you have time standards, wage rates which have been established by the contract, and you know what this group or department has averaged for the last six months, you can predict what the production cost will be almost within nickels.

CHAIRMAN FISHER: You spoke about placing incentive on indirect labor. Could you give us any concrete examples, for instance, how to set incentive on crane operators? On what basis would you establish it?

MR. CARROLL: In many foundry operations the crane man has considerable control over amounts of delay time, and that would be one of the substantial factors that I would consider in placing him on an incentive basis. Under a well established incentive plan, we have a measure of how much delay time there is, and if by a little more diligence on his part he can reduce that a certain per cent, I would be one of the first ones to be willing to reward him for it.

Now, to answer Mr. Fisher's question in a more general way, the way I would approach a question such as that is: What is it we are trying to accomplish? If, for example, in a steel foundry, where the man is using acetylene to burn off the gates, risers, and heads, we wanted to make some cost reductions of value there, I would not forget the fact that the acetylene is worth more than the man. So an incentive plan that would be based on that kind of casting cleaning should have some substantial portion based on the cost of acetylene and some portion based on the man's diligence.

MEMBER: Have you any suggestions on setting the standards on something such as contributory labor, melting department or cupola furnace?

MR. CARROLL: Most everything we find has a time measurement, irrespective of the plan which might be worked out. Take, for example, this crane operator question. I would start with timestudy, and you would say, "Well, this man does 10,000 jobs. Why start with timestudy?" Suppose at the start we decide that there should be four crane men per 100 operators, or what have you, and later we make a change.

Are we not going to be faced with accusations, false though they may be, of rate cutting because later we decide it should be three men, or we change some people in the productive department, which means three crane operators can handle it? When we have no record as to what was in the allowance for four men initially, we have no logical means of changing to something else.

So I would start with the timestudy in every case, and in anything as relatively simply as electric furnace operation, it is almost as clean-cut as the center floor molding.

MEMBER: You mentioned that when an improvement is made on a job or an improvement in the method for doing the job, then the standards should be changed simultaneously. What method should be used to remunerate the employee who suggests that improvement?

MR. CARROLL: I think the remuneration should be through a well-established suggestion plan, and do not assume the operator did not go beyond the second grade and try to get him to trade 100 cents on the dollar for 10 cents.

I do not know what amount it should be, but the most successful plan I know anything about is the Lincoln Electric plan in Cleveland, which has operated successfully with a 50 per cent basis. Whether it should be 50 per cent for one year, 100 per cent for six months, or some other percentage, I do not know.

I think that the suggestion plan, well operated, means adequate reward, adequate publicity, and putting the thing into effect, because the thing that the man is interested in, in the first place, is to see his own handi-work in operation, so that he can point to it with pride.

So there are three phases to successful suggestion operation, as I see it, but in total, I do not think it will solve more than 30 per cent of the problem. You still have a couple of hundred years of incentive background, in which many of us still think that it is the man's invention and the way to pay him is to let him keep on profiting from the old standards. We have got to live that down first, and I think it has to be a good offensive in the way of a well-operated suggestion plan, not one in which a committee of brass hats sits around and decides whether it ought to be $3 or $10.

No, it ought to be the difference in the standards, carried out in so many dollars, or based on, as you may choose, a year's activity, or if you wish to raise the percent of the award, a six months' activity. Some folks have gone so far as to base it on the previous activity instead of the forthcoming activity, in order to cut it off and have it a clean-cut calculation.

MR. GEORGE: Would you pay less than the average earnings? Let us take a dollar-an-hour-man, earning $1.30, and he is put on work that essentially cannot be standardized. We are putting a new model into production, building up from 10 pieces a day to 500 over three weeks. This is not exactly a trial period; it is getting the equipment ready, getting the

[3]Westover Engineers, Milwaukee, Wis.

job into high production.

MR. CARROLL: In order to answer your question I will go along with your assumption that you cannot put it on standard, but I do not agree. Let us take that assumption.

MR. GEORGE: You might have two or three temporary standards at different stages in the three weeks. Would you advocate using temporary standards?

MR. CARROLL: Not temporary standards, because temporary has a strange way of becoming permanent, especially if the contract reads that at the end of 90 days a temporary standard shall become permanent. I would not use a temporary standard under any circumstance. It has had a distasteful flavor for a great many years, because we in timestudy and management use the word "temporary" so we can keep a string on the standard and pull it back if we do not like the outcome.

I would set what I would call a plus standard. If you have standard data and a well-equipped timestudy department, especially if you purchased equipment on the basis that it was going to save you so much money which had been predetermined, then I would compensate the man with an extra time allowance, the same as if the casting got hard or the sand got wet or any other extraneous condition spoiled the casting.

The expense part of the standard would always remain separate from the basic standard. However, assuming that you cannot put the man on standard, as you state is the case, I urge the use in that and similar cases of a carefully thought out, proper job evaluation base rate, for that kind of skill.

MR. GEORGE: Suppose you had a properly thought out base rate in the example of a dollar an hour.

MR. CARROLL: But you would not have it for that kind of experimental work.

MR. GEORGE: Essentially it is not experimental work. It is carrying the job on. You are changing the job every day, improving the equipment, getting new fixtures in the production line. I agree with you, if the standard were the plus value, but many companies do not do that. They figure, "We are going to run anywhere from 100 to 500 and up it more and more each day until we get up to 500." In that period, to me the dollar is the valuated base rate. The man has been earning $1.30. If you call it an experimental period and choose to give him an evaluation of 115 per cent of base, you might do that. I wondered if you advocated that.

MR. CARROLL: I advocate that because I am deeply set against paying the average earned rate.

MR. GEORGE: So am I, but I wondered if you would rather see it go back to the dollar or go back to what we will call an experimental value stated in the contract, 110 or 115 per cent of normal valuation.

MR. CARROLL: A fixed amount is more fair because if you and I are working together on such a job and, because of your skill or energy you make $1.35 and I have been working at only $1.25, to guarantee each of us our past earned rate does not solve the problem. One of the difficulties arises because usually we pick the better earners to carry on this kind of work.

MR. GEORGE: I tried to avoid that by giving you just the production line changing from the 1946 model to 1947. We merely picked the old line and 17 workers on the line, so we did not deliberately pick the best people.

MR. CARROLL: To pay past earnings is another illustration of my initial premise that we are mixing together compensation and measurement. There is no relationship whatever between the output that is turned out during that period and any guarantee you might make. So why do we not call it day rate, and why do we not pay a rate for the job which is acceptable, or which, if it happens to involve an extra skill, as sometimes it does, the employee might have been earning $1.30 but this job might be worth $1.50, pay him $1.50.

MR. GEORGE: Essentially I am talking about the dollar job, and I am questioning whether he deserves more than the dollar when you have to take him off $1.30 incentive work temporarily and run it as a day-work job, maybe two days or ten days.

MR. CARROLL: The answer is, "No," because you have kept insisting that there is no extra skill involved. We know in advance he has given us 50 per cent performance or a 30-unit hour on day work, so there is no deserving of any more than $1. That is not the question.

MR. GEORGE: That is the question because many companies figure the man deserves to make more merely because he has been earning more. They lose sight of the value of the job, and see what the man has had as an income.

MR. CARROLL: That is right. That is only an easy remedy for what they term to be a difficult situation. It is a way of avoiding grievances.

C. S. HUMPHREY[3]: From your remarks, it occurred to me that there was a greater difference between a wage incentive plan and piece work than I had realized, possibly more than some of the other men have realized. I wonder if you would elaborate on the difference?

MR. CARROLL: Do not misunderstand me. Both of them are incentive plans, and there are a lot of incentive plans that are expressed neither in time nor money. An incentive plan is anything which is brought into effect which will induce the employee to do something he did not do before or would not do otherwise.

This suggestion award method is an incentive plan. That happens to be financial, but this idea of training programs, so that people can move up in the organization, is one of the primary desires. Most of us think wages are the primary desire, but all of your best industrial psychologists will tell you that wages come way down on the list. The desire to get ahead, the opportunities offered for advancement, training programs, and all that kind of thing are incentive plans called non-financial.

The only difference between the two plans that you asked about is that in one of them the standard is expressed in time, and in the other it is expressed in dollars and cents.

Unfortunately, prior to Taylor's time and in roughly 30 per cent of the country today, we still continue the piecework plan where we offer to pay 15 cents a dozen or $3.50 a ton for turning out so much production.

Essentially that was time, in the first instance, because they decided that the employee could do so many units an hour, and then wanted him to earn 80 cents, so they figure backwards, and that is the way they get the piece-work price.

There is a lot of good piece-work that was established from timestudy, and the quarrel that I have is not with piecework as such, but that we have gotten away from expressing our standards in time, which will be universal. There are still 60 minutes in the hour so that time is not affected by economic conditions. All of industry operates on a time basis, not a money basis.

The instant the customer agrees to give us the order, in the shop we are working on a time basis. So I have reduced consideration of overhead to the simple explanation that overhead is a rental charge which the customer pays for using our facilities for a certain length of time. All of industry is geared to do a lot in the way of controls that cannot be worked out with money as a base.

Suppose, for example, the molders get $1.25, and the workers at the shakeout get 80 cents per hour, and both rates have been used to set up piece-work prices. Production cannot be planned because there are two different sets of figures mixed. You would have to use a conversion factor, which is the conversion equivalent of using a time factor.

So that wage incentive, as I see it, is a way of continuing to get the man's interest in turning out a certain number of pieces with regularity. It seems to me much more convenient to pay on a time basis as we keep shifting contract wages around, and if we use time, then we can have business controls of one kind or another without going through all of the problems of changing over into money.

We can allocate overhead correctly, and we can plan production correctly. We can predict a lot of things in the line of improvements which bring men to a conference like this, where they can see all of the better ways of doing things. Then it is a question of whether or not we get our money back. All of those things are much more easily done with time as a basic measure.

MEMBER: In the jobbing type of operation, is it practical to write operation sheets to keep standards up-to-date?

MR. CARROLL: In my opinion, no. Operation sheets are a very decided advantage, and they partially answer the question with regard to how you know whether or not there has been a change

[3] C. S. Humphrey Co., Moline, Ill.

in the method.

For example, if a foreman were alert and understood the timestudy business, he would know that you cannot expect an operator to make the anticipated take-home unless at least he starts with the methods on which the standard was set. So if he had an instruction card he would know that much to start with. But you run into the law of diminishing returns, where the advantages may not outweigh the disadvantages of the costs of instruction cards.

In the jobbing foundries where you make two of this and six of that, and may never make them again, I would suggest that you might find some simplification, but by and large you could not have what I understand instruction card to be because you could not afford it.

MEMBER: I did not quite understand your answer to the question on experimental jobs. If you have an experimental job, I feel you cannot accurately determine the number of expense units to be added to that job, and you further feel that that job in order for it to be done properly would be necessarily given to one of your more experienced men. That man is used to earning, let us say 72 or 78 units, however, and you know he is a conscientious worker, and you get good supervision.

I am interested in knowing for control sheet purposes whether you would enter that as day work, where you convert your unit hour to 40, or break him off at 60 or 72 or 78, the unit hour he is accustomed to.

MR. CARROLL: I would break it off at 30. Maybe it should be 15, because if we do not know how many expense units there are in the total, I would certainly be on the conservative side, because if you carried your control sheet over into perpetual inventories and other things, I would rather anticipate the shrinkage than try to write it off later.

MEMBER: That is accurate for control, in other words.

MR. CARROLL: I wish you would not use the word "accurate." There is no such thing as an accurate standard, as you know.

MEMBER: How are you going to remunerate the man? How would he be paid? Suppose he just refuses to do the job if he does not get his double rate?

MR. CARROLL: Of course, that is where I suggested you might have a standardized specially prepared evaluation rate, for that and many other things like it, if you admit you cannot set standards on it.

I will give you a case in point. A plant introduced a new model and they were in a tremendous hurry to get it on the market because their competitor was introducing something new and was going to get it on the market first.

In order to prepare the operation sheets, preliminary to setting standards from data, they borrowed some highly skilled tool makers who had been earning some substantial premiums, to write those operations sheets.

They borrowed them partially because they needed an extra force, but primarily because they had the skill to write the methods. The question arose, How would you compensate those men? They had been making, shall we say, $1.25 base rate and 35 or 40 cents more premium.

The question was answered in terms of a specially prepared job evaluation rate, and the management said, "But you cannot do that. That is more than we pay the timestudy men."

My answer was, "But the timestudy men cannot do it."

It is not, as I am trying to answer your question, related to what the man has been making at all. It might be 10 cents an hour more or 20 cents less. It is what you would set up as a proper job evaluation rate for the kind of skill that you are asking that individual to display, fully recognizing that he is probably going to be, no matter how conscientious, operating at a day-work speed. The day-work speed is not his fault. Almost first, last and always, it is management's fault that he operates at the 30 rate hour, because we do not get the material to him.

MEMBER: Does a money standard give you more productivity than a unit or time standard?

MR. CARROLL: I do not think that question can be answered. If the two standards represent the same value and if the operators understood the incentive plan with its complications, as I explained it, I do not think there would be any difference. I can answer your question the other way, but you might not care for the answer.

If we were introducing an incentive plan to replace the piece-work plan, I think I could assure you of at least 30 per cent increase in production, my reason being that you would not be introducing a new plan if you were satisfied with the old.

MEMBER: What would you guarantee for an incentive plan over a day-rate plan?

MR. CARROLL: In production or cost?

MEMBER: Production.

MR. CARROLL: Fifty to 150 per cent. But management has to do the work. It is not the man in the shop that refuses to turn out more than 30 units an hour. It is because we do not get the material to him and get it to him in such shape that he can work on it.

You can take timestudy into a day-work shop and get a 75-unit hour, which is 150 per cent over a 30-unit hour, but only by segregating management's errors and getting management to reduce them. You cannot eliminate them, but you can reduce them. I say you cannot; you can, but you should not because it will cost more to eliminate them than you would save by so doing, but you can reduce them.

You can do a better job of planning. You can make the stuff come more nearly right. You can have the flasks so they fit and have them delivered. You can do all those things to enable the producer to keep on working. Most of those folks are just as anxious to do a good job as we are, and they do not do any more than they do, because we will not let them.

FRANK WARTGOW: A small packing plant was working strictly day work, and had always done so. They were faced with the possibility of cutting back from 48 to 42 hr. a week. Management did not desire to cut wages.

They called in consulting industrial engineers. The requirements that they set up were these: They did not care how much their employees earned over and above, but they wanted to be sure that they would not earn less for 42 hr. than they did for 48 hr.

After the survey we estimated the efficiency of the group was 35 per cent. Our superior was reluctant to submit that figure to the packing plant management, so we increased it to 45 per cent, knowing we were on the safe side according to management's specifications.

On my last visit to this plant prior to joining the A.F.A. staff, the operations were earning a 25 per cent bonus. They had increased their output 257 per cent.

MEMBER: Is there a difference between wage rate and compensation?

MR. CARROLL: I think of compensation in terms of take-home. I think of wage rate in terms of the base pay, the hourly guaranteed rate, many of which are negotiated or worked out through wage surveys or what not.

The wage rate is what the man should get for every hour he is present. I think he should get that regardless of the job we put him on. If he is a 90-cent man and we put him on a 60-cent job, I personally think we should keep on paying him 90 cents, because we agreed to utilize his skills at a certain rate, and it is our fault if we do not keep him working on jobs for which he is skilled. He gets that as his assured rate for every hour present, excluding transfers.

The base pay plan is supplemented by adding the incentive earnings, and the total is what I term compensation.

MR CARROLL *(author's closure)*: Concluding, may I suggest that we keep four things in mind.

1. Incentive standards should be built from standard time data to gain consistency of opportunity, fairness in revision, and maximum coverage.

2. Standards should be revised simultaneously with method change to avoid the difficulties involved with trying later to eliminate the excessive unearned premiums.

3. Incentive plans should be revised to eliminate from the standard times all allowances which may have been made to offset base rates that were lower than going rates.

4. Incentive plans should be extended to measure work before compromise brings on the payment of incentive monies for non-incentive performance.

REPORT OF THE SUBCOMMITTEE

GRAY IRON FOUNDRY SANDS

The Subcommittee on Physical Properties of Gray Iron Foundry Sands at Elevated Temperatures, Foundry Sand Research Project of A.F.A., adopted the procedure of first devising pattern and testing equipment, followed by actual foundry work. Laboratory research work under way should yield test methods of practical value.

Subcommittee on Physical Properties of Gray Iron Foundry Sands at Elevated Temperatures

THE FIRST MEETING of this subcommittee was held Oct. 30, 1944, at Detroit. The duties of this subcommittee were outlined as the formulation of test methods that will be required by the gray iron and malleable iron foundries to measure the elevated temperature physical properties of iron molding materials and to set up performance specifications of the necessary testing machines.

The subcommittee agreed to first attack the problem of developing a test method whereby the shakeout effort required to remove cores from castings could be measured by a quick laboratory test. The first step in the program was to devise two sets of pattern equipment which could be used in the foundry to produce first a test casting with a relatively light metal thickness as compared to the thickness of the core. Details of this test casting are given in Fig. 1. This casting is to be poured with gray iron at a measured temperature using core mixtures of various compositions to secure a wide range of mechanical work to remove the core from the casting. This work is to be expressed in time of jolting required to remove the core from the casting.

The second test pattern setup,

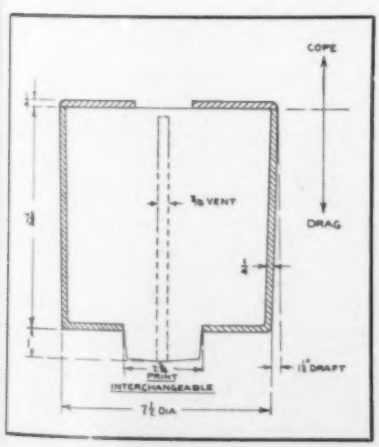


Fig. 1—Test casting for evaluating ease of core knockout of different core mixtures.

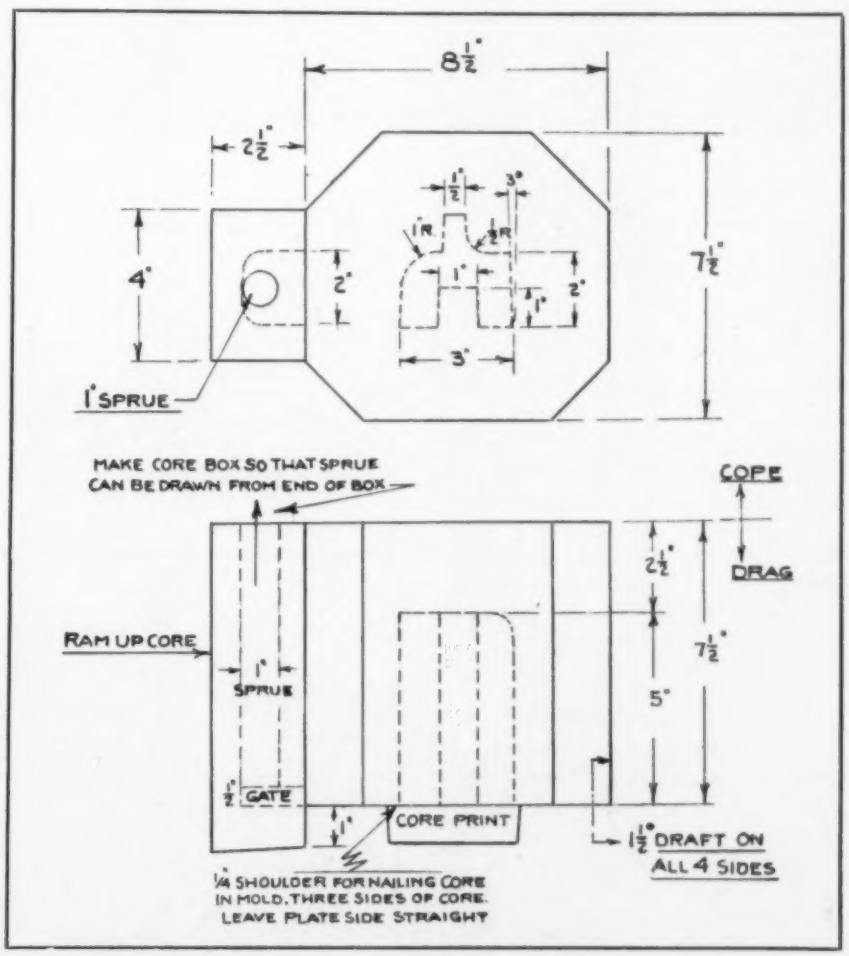
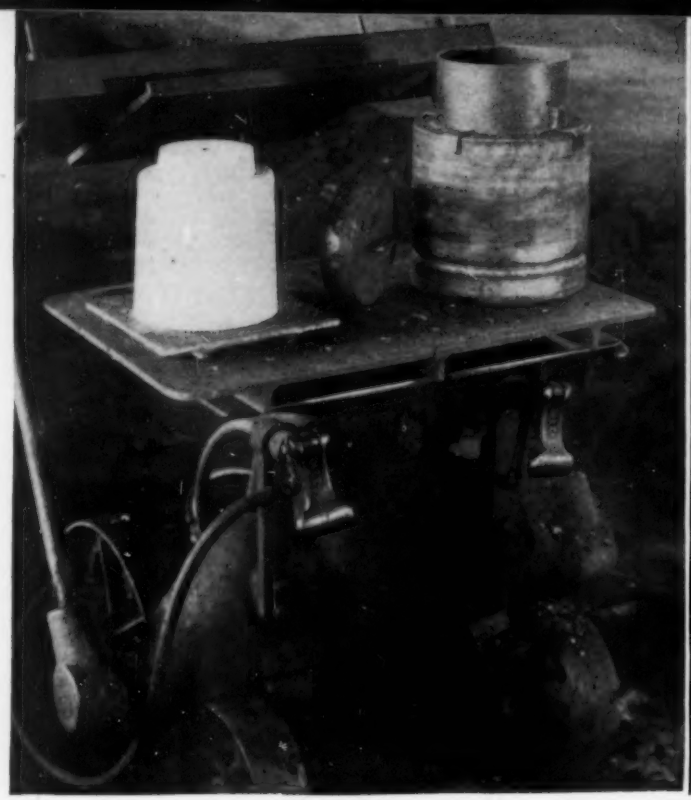


Fig. 2—Test casting for measuring the degree of penetration, quantity of veining, and quality of surface.



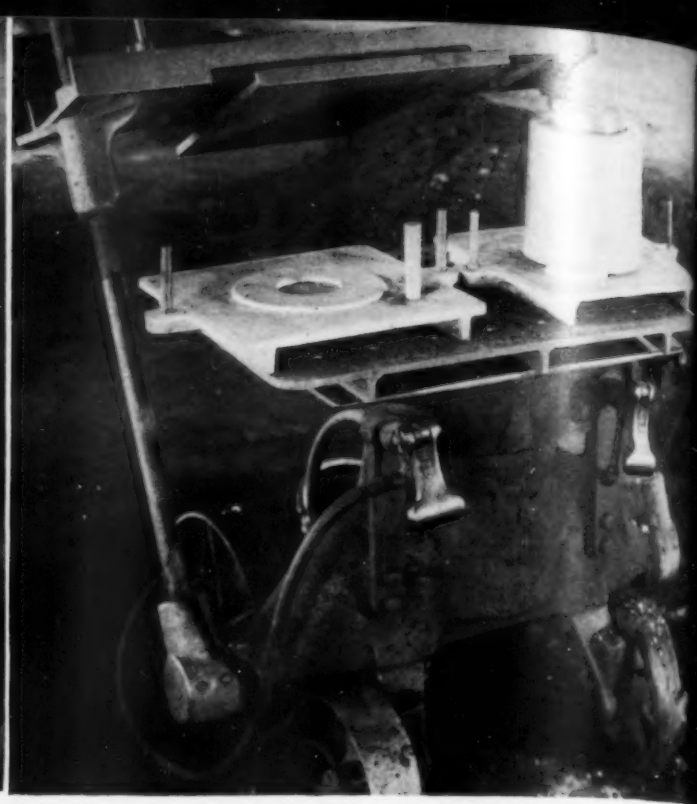


Fig. 3—(above) Core knockout pattern equipment for making the knockout test core. Fig. 4—(right) Pattern equipment used to make the mold for core knockout test casting.

as detailed in Fig. 2, is to be used to measure the degree of penetration and veining, also the quality of cored surface. The metal, at a controlled temperature, is to be cast against cores of different composition. From these foundry test castings, these various core mixtures are to be graded as to the ease of shakeout, degree of penetration, quantity of veining and quality of surface.

Laboratory Evaluation

After this program, these same core mixtures are to be tested in the laboratory, endeavoring to devise test methods that will evaluate these cores in the same order of ease of shakeout, degree of penetration, quantity of veining and quality of surface.

At the present time, all of the pattern and core equipment has been constructed. The core equipment for the core knockout effort test is shown in Fig. 3, while the molding and pattern equipment is shown in Fig. 4. Two penetration test cores are shown in Fig. 5 as used for studying veining, surface finish, and penetration. Associate Professor W. A. Spindler and John Grennan of the University of Michigan, Ann Arbor, where the foundry tests will be made, have cast a number of trial castings from this equipment.

The question as to the size of a laboratory test furnace that should be used in this problem has been raised. The majority of the dilatometer furnaces used are equipped with a furnace with a 2-in. inside diameter hearth. Two larger dilatometer furnaces which are available for this investigation have hearths with an 8-in. inside diameter, having an available inside diameter of 4½ in. with an 11-in. length.

The heat transfer into a 1⅛x2-in. sand specimen is shown in Fig. 6 for these two sizes of dilatometer furnace. It may be noted that there is very little difference in the rate of heat transfer. The sand specimen reaches a furnace test temperature

Table 1

TEST CORE MIXTURES AND PROPERTIES

Mixture	*Material*	*Percentage*	*Moisture in Test, per cent*	*Flowability*	*Green Compression Strength, psi.*	*Core Permeability*	*Tensile Strength, psi.*
No. 1	Michigan City sand	99	2.9	88	0.25	184	140.5
	Oil	1.0					
	Water	3.6					
No. 2	Michigan City sand	70	3.9	91	0.4	125	185
	Michigan bank sand	28.3					
	Cereal binder	0.5					
	Water	4.5					
	Oil	1.2					
No. 3	Michigan City sand	87	5.1	86	2.0	95	164
	Silica flour	10					
	Western bentonite	0.5					
	Cereal binder	1.0					
	Water	5.4					
	Oil	1.5					
No. 4	Hayville gravel	49.0	7.45	70.5	4.1	56.4	15.5
	Heap sand	49.0					
	Pitch compound	2.0					
	Water	7.5					
No. 5	Michigan City sand	15.0	7.6	68	9.25	66.1	17.5
	Heap sand	81.5					
	Cereal binder	0.5					
	Resin	3.0					
	Water	8.5					

of 2500° F. at the same time of 8 min. for either furnace.

The rate of heat transfer is slightly faster for the large furnace in the first 15 sec. where the moisture is converted to steam. This faster rate of heat transfer is distinguishable but not of sufficient practical magnitude as this sand specimen temperature increases up to 2000° F.

The subcommittee's method of first securing practical foundry evaluation and then proceeding with the laboratory research work should yield test methods that will be of real practical value.

The second meeting of this subcommittee was held at the foundry of University of Michigan from June 4 to 7, inclusive. At this meeting the core knockout effort was measured and degree of penetration was studied by actual foundry casting tests.

Preparation of Core Sands

All ingredients of core sand mixtures were weighed carefully and mixed in a size 0 muller-type mixer. Dummy batches were prepared before test batches were mixed to condition the surfaces of the mixer. The compositions of core mixtures for the five different core sand mixtures employed are given below, as are the order of adding ingredients and length of mixing time.

Mix No. 1—Ingredients: Lake sand and oil added. Mixed 4 min. Water added and mixed 6 min.

Mix No. 2—Ingredients: Lake sand, bank sand, and cereal binder added. Mixed dry 2 min. Water added and mixed 4 min. Oil added and mixed 4 min.

Mix No. 3—Ingredients: Lake sand, silica flour, western bentonite and cereal binder added. Mixed dry 2 min. Water added and mixed 4 min. Oil added and mixed 4 min.

Mix No. 4—Ingredients: Hayville gravel, heap sand and pitch compound added. Mixed dry 2 min. Water added and mixed 6 min.

Mix No. 5—Ingredients: Lake sand, heap sand, cereal binder and resin added. Mixed dry 2 min. Water added and mixed 6 min.

Coremaking. Six shakeout test cores were made of each core mix. The cores were rammed with 25 jolts on jolt machine with a 3-in. drop and with a 18.5-lb. weight resting on top of the sand. Four block penetration cores were made of each core mix, employing ten jolts with a 7-lb. weight resting on top of sand and ten additional rams with ten drops of 7-lb. weight falling 6½ in.

For laboratory tests, the following test cores were prepared for each mix in accordance with A.F.A. specifications or approved methods:

6—Tensile cores.
300—1⅛x2 in. dilatometer test cores.
3—Core permeability cores.
3—Confined expansion cores.
3—Hot gas pressure cores.

Core Baking. All cores were fired in a foundry core oven at time and temperature as shown in Table 1.

Molding. All molds were made with well prepared heap sand which tested:

Green Permeability.....36.0
Green Compression, psi..... 7.1
Moisture, per cent..... 9.2

All molds were rammed with 25 jolts with a drop of 3 in. Aluminum flasks and bottom boards were used with cast iron weights clamped in place. All the sand molds were numbered.

Metal. The molten metal was obtained from an indirect-arc, electric, rocking type furnace. Metal analysis

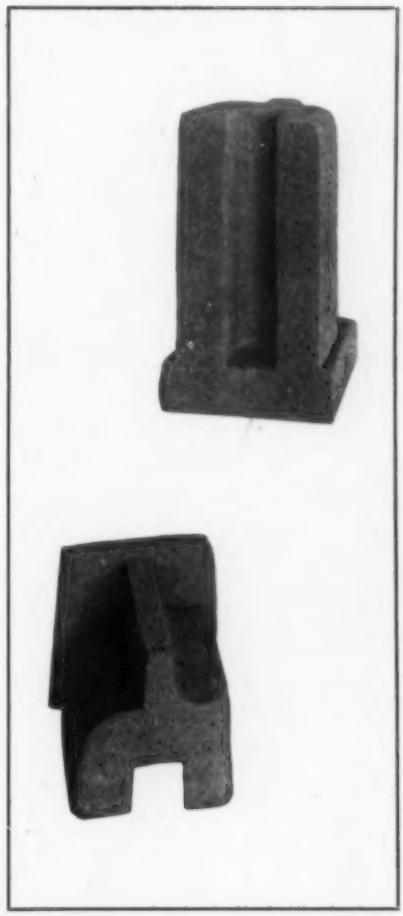


Fig. 5—Type of core used in penetration study.

Table 1

Test Core Mixtures and Properties

—Baking Time, hr.— Foundry Core, hr.	Test Core, hr.	— Average — Top 100's	Bottom 100's	Average Pouring Temp., °F.	Casting in Mold, hr.	Casting Cooling Time, hr.	Core Knock-out, sec.	Knock-out Sand Temp., °F.	Order of Finish
5	3	65	76	2740	1	4.5	5.6	112	4
5	3	68	90	2692	1	4.5	44	106	1
5	3	83	90.5	2746	1	4.5	315	138	3
5	3	81	78	2650	1	4.5	479	103	5
4	3	40	60	2765	1	4.5	18.6	137	2

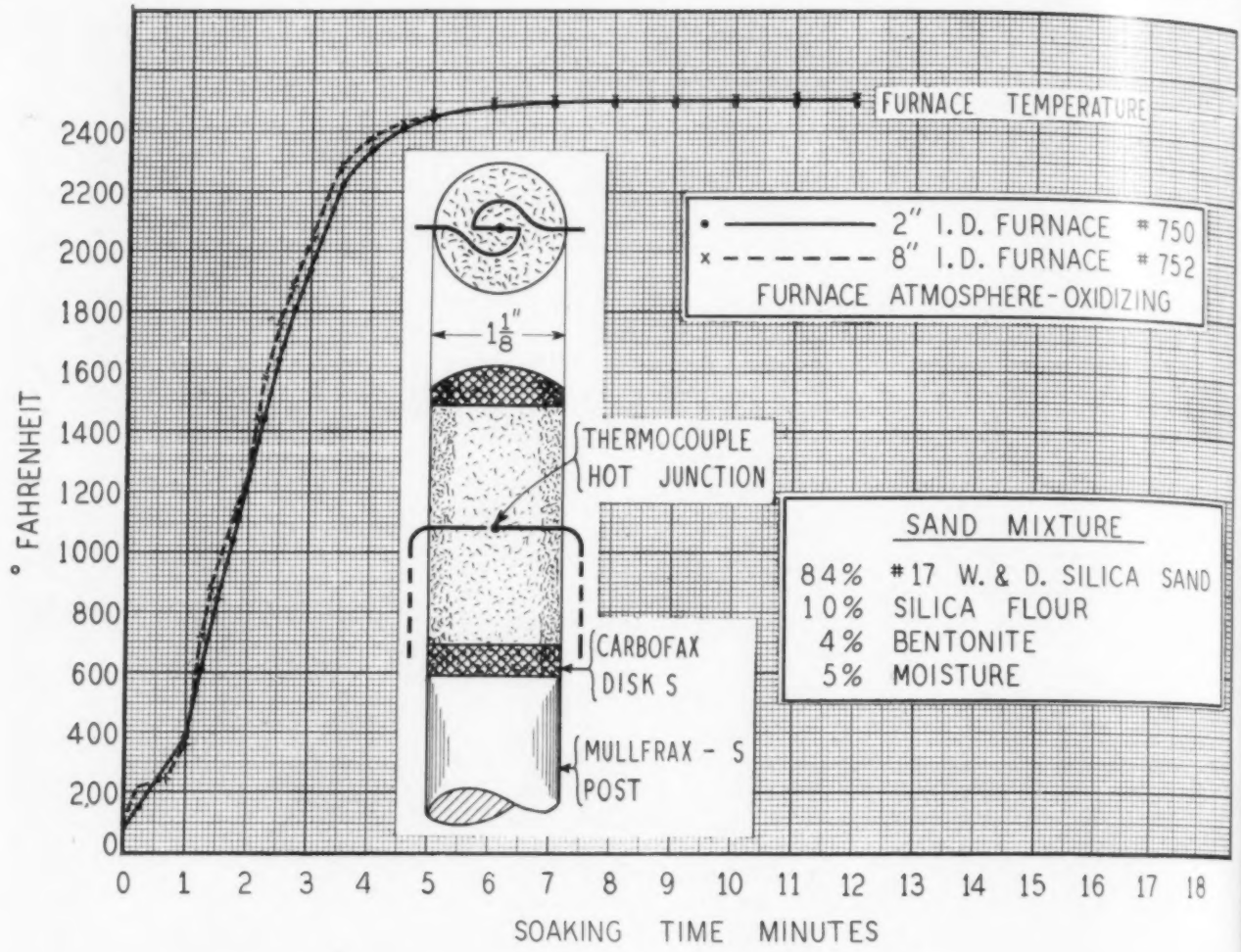


Fig. 6—Heat transfer into a 1⅛x2-in. sand specimen when inserted into two different sized furnaces.

of the mixes are shown in Table 2.

Metal was tapped from furnace at 2800° F. ±50° F., using an optical pyrometer for temperature measurements. The molds were poured with a 150-lb. double-shank ladle and metal temperature readings were taken. See Table 1 for average pouring temperatures.

The penetration test castings shown in Fig. 7 were made on the first day, and no more of these castings were made since no penetration was experienced with the electric furnace iron of the composition used. Committee members have experienced penetration with this test pattern when using cupola metal. A report will shortly be made covering this phase of the work.

Casting Cooling Time. After the core knockout test molds were

Table 2

METAL ANALYSIS

	Mix Number				
Component	*1*	*2*	*3*	*4*	*5*
Silicon, per cent	2.29	2.26	2.16	2.40	2.36
Total Carbon, per cent	3.11	3.12	3.02	3.20	3.16
Combined Carbon, per cent	.....	.....	.....	.....	.....
Manganese, per cent	0.54	0.63	0.53	0.61	0.61
Phosphorus, per cent	0.269	0.273	0.318	0.311	0.276
Sulphur, per cent	0.095	0.101	0.101	0.082	0.090

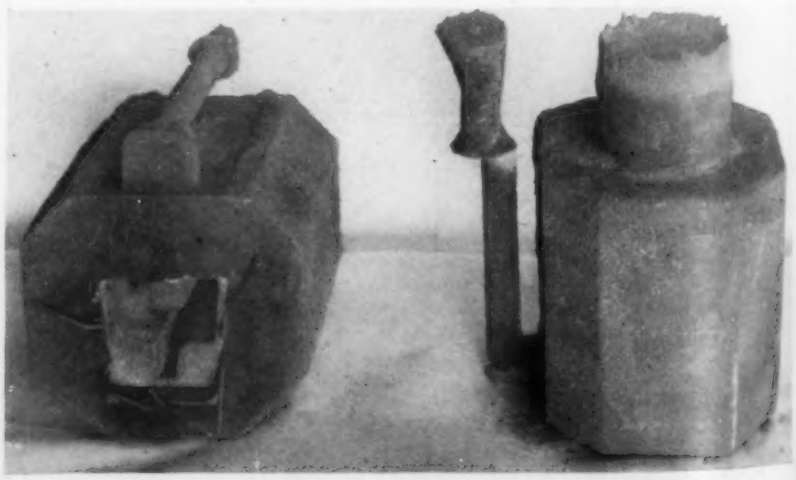


Fig. 7—Penetration and veining test castings made from electric furnace iron.

poured. They were allowed to cool for a period of one hour before the castings were removed, and then the castings were allowed to cool for a period of 4½ hr.

Core Knockout. To measure the energy required to remove the cores from the castings, a casting holding fixture was bolted to a jolt machine with a 3-in. ram drop, as illustrated in Fig. 8. The time in seconds of jolting required to remove the core from the test casting was used as the index of core removal energy. This method of test worked satisfactorily. The knockout time is shown in Table 1. A very good spread of knockout time was secured. This gives a very good start for correlation of knockout time with high temperature test method research work. The core knockout time for the five cores under test is shown in Fig. 9.

Laboratory tests at elevated temperatures are now under way on core specimens made at the time the casting cores were made.

All members present at the June meeting agreed to make additional foundry penetration tests. Test castings will be made by Messrs. Satz, McMurray and Olsen. The problem of penetration, while extremely interesting, is difficult to reproduce.

It was agreed that correlation between foundry knockout effort and laboratory tests will form our major problem, carrying along the research on foundry tests on penetration as a minor problem.

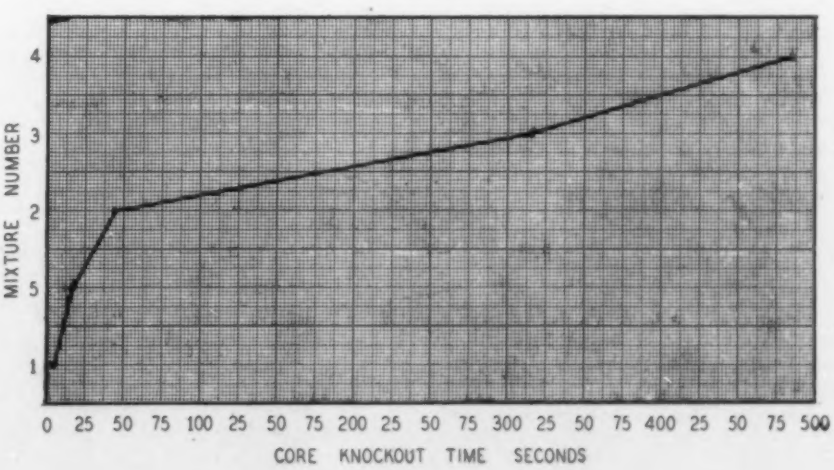
Subcommittee Personnel

Personnel of the Subcommittee on Physical Properties of Gray Iron Foundry Sands at Elevated Temperatures at the time this report was prepared was as follows: H. W. Dietert, chairman; Day E. Cutler; J. Schumacher; J. A. Gitzen; Wm. A. Spindler; H. G. McMurry; Leon B. Thomas; John Grennan; D. C. Williams; E. W. Olson; George F. Watson; E. C. Zirzow; and Arnold Satz.



Fig. 8—Core knockout equipment as developed for measuring work required to remove core from test casting. Committee members conducted tests at University of Michigan foundry.

Fig. 9—(below) Spread of knockout time for removal of the five types of cores under consideration in these tests.



RAMMED REFRACTORIES for ELECTRIC FURNACES

Robert H. Zoller
Zoller Castings Co.
Bettsville, Ohio

BRICK AND MONOLITHIC refractories in steel melting are familiar subjects among foundrymen. The aptitude of brick or monolithic lining to the job is of prime importance to the operator and the overall foundry operation.

Selection of the proper type of refractory should correct refractory problems and cut operating costs. Of course, no one refractory can be made to completely correct all problems, but the job can be studied and the right type of refractory applied to that job. The thought must be that an improved refractory can be developed for any job. Use of a refractory should be discontinued when it is known to be giving poor service and hindering production.

Monolithic Linings

Brick refractory used today in electric furnaces or ladles may be replaced in almost all cases with monolithic linings. However, brick cannot be replaced by rammed linings in cases of arches or suspended sections where the refractory must support its own weight. Some work has been done on the development of a refractory mixture for ramming furnace roofs.

In electric furnaces, where high temperatures are encountered, silica brick are used. Silica brick have been used for many years, and with sufficient care can be prepared for service without spalling or cracking. However, time for heating is required or deficiency in the brick will be encountered. Temperature change is not responsible for all unsatisfactory performances of refractory brick. A silica brick, within reason, has the same expansion as the brick next to it.

Expansion

However, laying of the brick may not at all times be such that expansion will be the same in all directions. The principle reasons for this are: (1) the first row of bricks not started correctly; (2) bricks not of true size. The first reason can be corrected by sufficient care and training. In many cases furnaces have metal penetration in brick seams, entailing much extra work for preparation of a level footing for the first course. As to brick sizes, refractory manufacturers have made much progress toward producing brick of uniform size.

A picture of the electric furnace lining has now been presented. Difficulties due to expansions, brick joints or seams, and leveling of bottom of side wall can be eliminated by using a rammed refractory. A clean, irregular surface is all that is required for the foundation of a side wall of a monolithic-lined furnace. For good practice, only a liquid bonding agent is required on this surface.

If for some reason a lining is becoming thin in a section, due to burning of the arc or from corrosion, the side wall can be increased in thickness at this section. A furnace might, under certain conditions, require only partial relining. This repair job would not be at all difficult as no leveling is required.

After the section has been thoroughly cleaned and washed with a liquid bond or a dry bond mixed with water to make a wash, the refractory is rammed into place. If the surface of old refractory has been properly cleaned and bonded, metal penetration will not occur. The expansion will be the same in all directions, and proper density obtained with skilled ramming.

Ramming Mixture Coarseness

Rammed electric furnace bottoms are successful if a proper refractory mixture is worked out for the job. It has been found that ramming mixtures should have a high per cent of coarseness, and it is the author's opinion that coarseness should be increased almost to the point of causing trouble in the ramming operation. This gives greater fusion depth and, if a high-temperature ganister is used, it will also raise the fusion temperature of the complete ramming mixture.

It must be remembered that proper skill should be exercised in all ramming operations. A bottom

Selecting a refractory for the job; developing routine methods for refractory installation and preparing the unit for service; drying and preheating are important factors in lengthening refractory life and decreasing foundry operating costs.

Presented at a Refractories Session of the Fiftieth Annual Meeting, American Foundrymen's Association, at Cleveland, May 7, 1946.

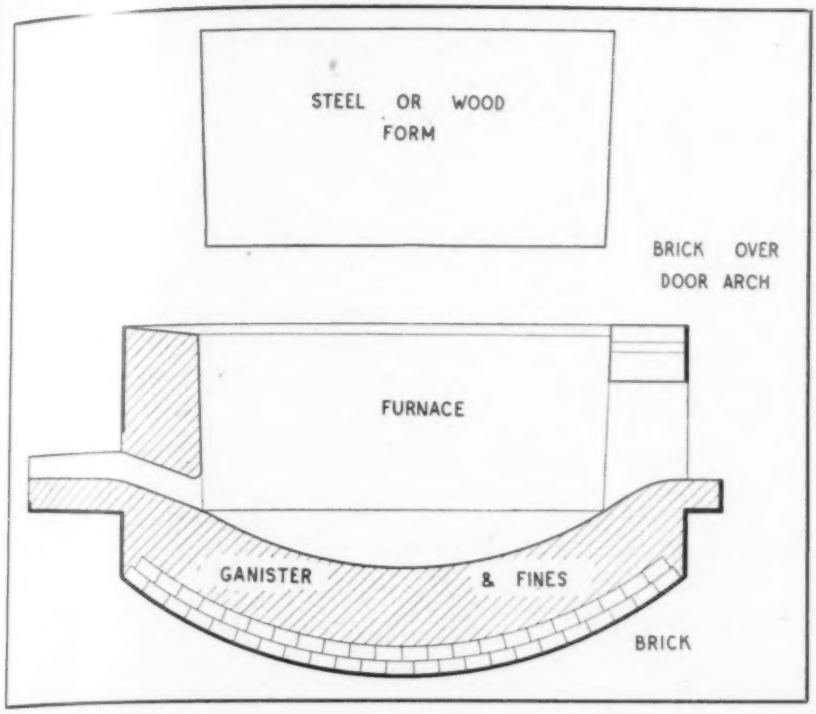


Fig. 1—Method of form construction for furnace side wall rammed refractory.

should have as great a density as possible. The density obtained by a good ramming job will be cushioned and bound by the high percentage of coarseness in the refractory mixture, as the coarse material has less expansion than the fines. Upon heating to high temperatures, this cushion and binding are definitely needed or sections will break loose, thus taking away refractory life from the rammed unit.

Ladles of different sizes can be lined with the same mixture or one similar to that used in the electric furnace side wall. Usually, it will be found that standardizing refractory mixtures will be of great assistance in keeping them consistent. In some small ladles with refractory bridges, trouble may be encountered and the ramming mixture might have to be changed accordingly.

Refractory Heating

Development of a highly bonded refractory, or one with a degree of plasticity well under top fusion temperature will not solve this problem, as slagging of the refractory will be encountered. Proper heating of the refractory is important and will be discussed in a later paragraph.

In refractory selection, the versatility of the monolithic refractory and its aptitude for the job should be considered. In most cases, special shapes or cutting of bricks are out of the question. Forms are made for the furnace side wall, as shown in Fig. 1, and for a teapot ladle, as shown in Fig. 2. The forms are constructed to give the contour desired in a unit.

Ramming Operation

In many cases, for convenience, the form may be made in sections, either loose or joined together for adjustments in the ramming operation. In extreme cases, where certain sections of a furnace or ladle may require a patch, the form may be adjusted, blocked or wedged in such a manner that a new portion can be rammed in any part of the unit.

In working out a good refractory for steel furnaces and ladles, three main points should be considered: fusion temperature and sintering range, fines or coarseness, and density of the ram. These points should be studied carefully. In steel furnaces, a refractory with a high fusion temperature is desired. It is not at all a simple matter to properly bond together a refractory mixture with a high fusion temperature.

It is definitely desirable to use a good grade of high-temperature ganister in the mixture. If advisable, a sized and washed ganister may be employed at no additional cost as compared with a good grade of ganister. Here is found a definite reason for improvement because of the certainty that a consistent amount of coarseness or fines is added to the mixtures.

This point cannot be minimized because it is important to use the high-temperature ganister in as great proportion as possible in the refractory mixture. Consequently, if properly studied and applied, both fusion temperature and coarseness or fines may be used to their greatest advantage.

Coarseness Proportions

However, it is not advisable to use the high proportions of coarseness in a refractory mixture for a furnace side wall or ladle as that used in a furnace bottom. The reason for this is that a side wall cannot be properly bonded for heating. The third point, density of ramming, is obtained by proper proportions of fines and skill in ramming.

Here again proper proportions of of coarseness and fines play an important part. A consistent result necessitates: (1) a consistent refractory mixture; (2) the ramming or handling of that refractory mixture should be the same at all times.

A furnace or ladle should always have a refractory drying period to

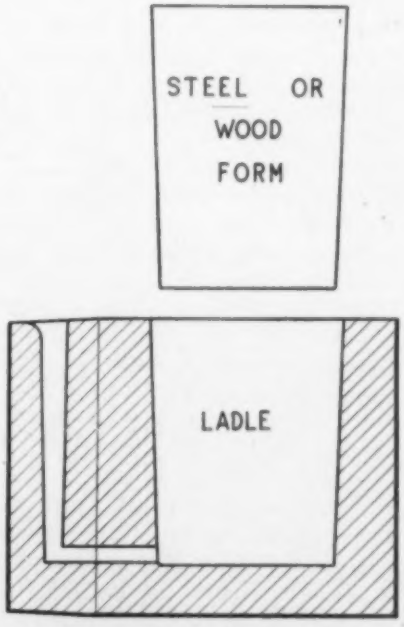


Fig. 2—Form construction for teapot ladle rammed refractory.

obtain good results. This fact is admitted in most cases, but few ladle departments have done much about it. Ladles may be put into operation with only partial drying of the refractory and fair results obtained when operating a makeshift set-up.

Properly Dried Ladles

If an insufficient dried ladle is put into service holding high-temperature steel continuously during 24-hr. days, it has been proved that the results will not be satisfactory. Usually, records will show 75 per cent, or less, of the efficiency realized from a properly dried ladle. The time required for drying ladles is about 2 to 3 hr. per in. if dried in an oven at 350° F. The moisture added for the bonding qualities and ramming is definitely undesirable after the ramming and should be totally eliminated, except for the low percentage of combined water in the mixture, before the unit is heated for use.

Furnace and ladle shells can be easily prepared for drying. Small ladles, such as one-man shank ladles of 1 or 1½ in. refractory thickness, need only an oven dry. The furnace and larger ladle shells should be prepared by drilling or burning ⅜-in. holes on 12 to 18-in. centers. This will allow the moisture to escape freely in all directions.

With the proper drying of a ¼-ton ladle with skimming bridges, and pouring continuously, it was found that with continuous pouring a 15-ton average could be maintained through these ladles, as compared to the 3 to 5-ton average obtained with improperly dried and heated ladles.

With reference to ladle drying time, the author wishes to emphasize that this time period is from the time ramming is completed to the time the ladle is to receive metal. The ladle is properly dried, then heated to temperature for the molten steel by the ladle cover and heater shown in Fig. 3 (for large ladles), and in this heating the refractory should remain intact.

Spalling or breaking away of sections may be caused by applying heat too drastically or by a direct flame against a portion of the refractory itself. The heating of any ladle is of prime importance and cannot be taken lightly. If the refractory is too low in bonding, which it will be for steel, in most cases, it will be necessary to add a liquid bond to the surface. This will penetrate into the refractory about one in. and will hold the refractory from spalling through this period. Many liquid bonds on the market will answer the purpose.

Proportions of the ingredients should be consistent at all times if the refractory is to be kept at high efficiency. The work area of the refractory department can be so arranged that all work can be standardized for the mixing operation. Bins for the ingredients, with proper measuring tools, should be arranged conveniently. Measuring cans should be provided for the moisture or any liquid bond that might be added.

Moisture, coarseness, fines, and bond can be added accurately. It is always a good practice to place all dry ingredients in the mixer and mix thoroughly in the dry state before adding moisture. Usually, a ramming mixture will be satisfactory for ramming if the moisture is held at between 5 and 6 per cent. However, to obtain maximum ramming qualities in some refractory mixtures may necessitate varying the moisture content out of this range.

Dry Refractory

Drying and heating for use of the electric furnace does not require any great skill, but a furnace should definitely have a dry refractory before starting. The bottom can be burned in slowly with coke after an overnight or its equivalent period for predrying with burners or wood. This is a simple operation. The furnace is completely lined, with roof and electrodes in place.

A large wheelbarrow of coke for each ton of furnace melting capacity is leveled off over the bottom of the furnace. The electrodes are then lowered to the coke with power on and rheostats adjusted to low fire, with low tap. A short period of fire will induce the coke to burn in the center. At short intervals of firing, the furnace may be either dried out further while being heated for use or, if desired, the roof and side walls may be glaze fired.

For many who have not used coke for this purpose, it is interesting to know that coke will make exceptionally good contact and can be heated to a white heat if desired. If the bottom is to be fused on the surface, electrodes may be placed as a letter "T," making a contact between the three electrodes. Then high-tap firing at intervals will fuse the surface, as the arc may be directed in any direction. In a bottom-fusing operation, the furnace must be preheated to a red heat.

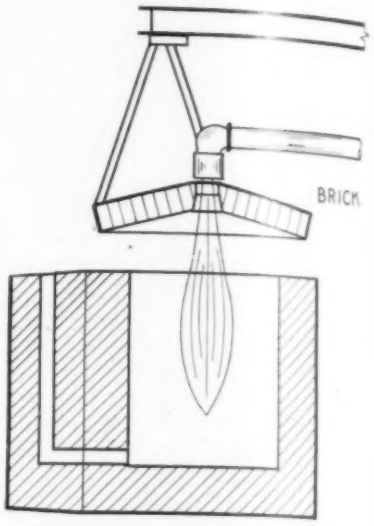


Fig. 3—Ladle cover and heater.

If the side wall is also rammed, it should be dried before heating with coke, but need not be heated to a red heat such as recommended for fusing bottoms.

Standardization of refractory mixtures and maintaining routine practices have always been necessary in making a good refractory mixture. The importance of methods set up in the refractory department, and care of the refractory in the unit being repaired or rebuilt, cannot be minimized.

A method which may work in one plant, of course, does not always work to a good advantage in another plant, but that is no reason for condemning the refractory. Different operation necessitates different handling of metal; consequently, the same results may not be realized and the refractory must be changed or altered for the operation in which it is to be used.

DISCUSSION

Chairman: C. E. BALES, Ironton Fire Brick Co., Ironton, Ohio.

Co-Chairman: C. S. REED, Chicago Retort & Fire Brick Co., Chicago.

CO-CHAIRMAN REED: What type of mixing equipment do you recommend?

MR. ZOLLER: I have seen a concrete mixer, regular mullers, and high speed mullers used. One refractory ladle lining mix that was used had a life of 54-5/10 heats per ladle. That is continuous pour-

ing of manganese steel castings 24 hours a day. This lining was mixed in a cement mixer. A high speed muller was then used and the lining life was increased to 85 and some tenths heats per ladle. Then we took the same high speed muller and by adding 35 per cent coarse material of ⅛ to ¼-in. high-fusion washed ganister to the mix; increased the lining life to 109 heats.

[1]Technical and Research Director, Steel Founders' Society, Cleveland.

CHAIRMAN BALES: Mr. Briggs of the Steel Founders' Society, did you observe any foundries where ramming refractories were in use when you were in Germany a short time ago?

C. W. BRIGGS[1]: Yes, I saw many furnaces rammed with monolithic linings. However, nearly all the furnaces were basic lined. The few electric furnaces using an acid lining were lined with silica brick.

I should like to make the following comments concerning the paper by Mr. Zoller. Rammed silicious linings for acid electric furnaces have been used for a number of years in a fashion similar to that described by the author. It was regretable that Mr. Zoller could not have added to this information by detailing the procedure employed, such as; the time of mixing, the type and quantity of bond materials, sizing classification, moisture content, etc., instead of continually referring to the proper mixing, the proper degree of coarseness, and to other proper conditions.

ENGINEERING PROPERTIES OF HEAT TREATED CAST IRONS

J. S. Vanick
International Nickel Co., Inc.
New York

GRAY CAST IRON as defined[1] embraces a product with an "as cast" strength range of 20,000 to 80,000 psi., a hardness of 120 to 350 Brinell, and a machinability that varies as the strength, hardness and structure indicates. This material is essentially the product of all gray iron foundries. It occasionally dips into the output of malleable iron foundries which deliver machinable gray iron castings without applying the malleabilizing anneal.

Methods of manufacture, or alloying processes, or similar foundry practices do not exclude castings from exhibiting changes in properties in accordance with the fundamental laws governing the heat treatment of cast iron. By resorting to the simpler analogy that cast iron is essentially a steel interspersed with graphite flakes, it is obvious that the laws governing heat treatment of steel apply in general to those governing cast iron. Cast iron may be hardened, softened, strengthened, and to a moderate extent, toughened by appropriate heat treatment.

Cast Iron vs. Cast Steel

There is one important difference between castings of cast iron and steel. Practically all steel castings must be heat treated to break up a weak and brittle, coarsely crystalline structure that forms during solidification. Cast iron, in contrast, precipitates graphite during solidification.

This precipitation of graphite obscures and apparently obstructs the progress of crystallization such as occurs in steels, so that the amount and disposition of this graphite formed at the birth of the casting largely determines the toughness and the strength characteristics, while the composition and rate of cooling establish the final strength and toughness. This is a fundamental reason why cast iron, properly alloyed, develops excellent properties in the "as cast" form, whereas cast steel, especially if alloyed, should be heat treated to develop best results.

Unless the foundryman delivers heat treated castings to the prescribed physical property specifications, it is important for the heat treater to know all about the composition of his material to achieve the best results. This is especially important for the heat treatment of production lots of castings where a close adherence to specifications is necessary so that a standardized heat treatment may be depended upon to yield uniformly desirable results. The most common heat treatment procedures in practice today are shown in Fig. 1.

Stress Relief Annealing. Stress relief annealing is defined as the process of reducing internal stresses in a casting by heating it to a temperature below the transformation range and holding for a proper time at that temperature, followed by slow cooling. When properly applied, the treatment does not alter the general engineering properties of the material appreciably.

The object of this paper is to simplify the description of heat treating processes, and contrast "as cast" with "heat treated" in as simple a comparison as possible. The heat treatment of cast iron is comprehensively dealt with in the "Cast Metals Handbook" and "Alloy Cast Irons," published by A.F.A.

It serves mainly to apply a controlled, slow, uniform rate of cooling to castings which by their design, shape, section, composition and method of molding and gating, have cooled in the mold at irregular rates. For such castings, heating to 900-1150° F. serves to relieve any local concentration of stress. In more common cases, the treatment is applied to achieve dimensional stability where accuracy in dimensions is a required property.

Castings should be heated slowly (usually at a rate not exceeding 100° F. per hour). Large castings of irregular shape and section must be handled carefully to make sure that the heat penetrates uniformly into heavy sections and box-shaped pockets. Holding time at the annealing temperature need not exceed one hour per inch of section, unless a lower temperature, in the 800-900° F. range is used, where an additional 5 to 6 hours may be added to compensate for the slower rate of stress relief, as indicated in Table 1.

Most castings are used in the "as cast" condition. Foundrymen have applied practices which reduce internal stresses to a negligible factor. Designers have cooperated by providing uniformity and symmetry in sections. Many large castings of heavy section cool slowly and uniformly in the molds and do not retain enough residual stress to require an anneal.

The frames, beds and parts of high precision machinery may be stress relief annealed to be assured of permanence in accuracy. Large, irregular shaped castings which are heated in service and must retain dimensional accuracy are usually

This paper was presented at a Gray Iron Session of the Fiftieth Annual Convention of the American Foundrymen's Association, May 10, 1946, at Cleveland.

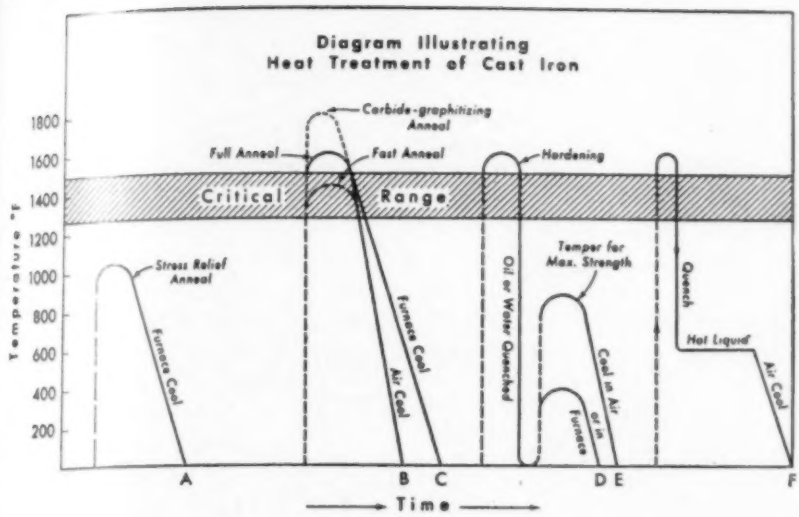


Fig. 1—Heat treatment of cast iron.

stress relief annealed. Steam turbine housings are a typical illustration.

Auto brake drums and clutch plates which become heated in service must be free from stresses or be stress relief annealed; otherwise, eccentricity may occur. Engine castings such as cylinder heads, blocks, liners, pistons, et cetera, may require annealing, depending upon design and foundry practice. In general, unnecessary stress relief annealing can do no harm, but it will add to the cost of a casting.

Residual Stresses

In some castings, a residual stress is a necessary factor to successful operation. Piston rings are a typical example. Foundrymen and engineers in general have not learned how to introduce and control favorably disposed residual stresses which could be employed to improve the subsequent performance of the equipment. This remains an attractive field for future research.

Annealing. Annealing has been described as a process involving heating and cooling usually applied to induce softening. The engineering properties of cast iron are altered by annealing to the extent that the strength may be decreased 10 to 30 per cent; the hardness decreased 30 to 150 units on the Brinell scale, and the structure may contain more graphite and ferrite.

Improvements may occur to the extent of a small increase in toughness, better thermal conductivity, greater dimensional stability, and an increase in electrical conductivity with associated electrical properties; also improved enamelability in the case of enameled ware.

Softening for Machinability

Production castings which are to be hardened by quenching and tempering are often annealed prior to machining. Manufacturers equipped with heat treating facilities frequently find it economical to apply a softening anneal to castings such as cylinder liners which after machining are quenched and tempered to achieve a high hardness.

The desirable pattern of the graphite which has formed during the solidification of the casting is not altered appreciably by the anneal. The low temperature anneal (heating within the critical range) is preferred in volume production because of its time saving advantages. Annealing treatments are illustrated in Fig. 1, and summarized in Table 2, covering Annealing Treatments.

Annealing treatments "A" or "B" (Table 2) are most commonly employed. The "A" treatment requires close control, but it does not weaken the castings excessively and does not tie up furnace equipment to the same extent as "B" or "C." Certain automotive castings such as pistons may be annealed to achieve an aggregate improvement in control over uniformity in the machine shop as well as dimensional stability under elevated temperatures.

The "B" treatment softens and weakens the iron. The heating time may be prolonged until practically all of the carbon has been precipitated as graphite—and a hardness of 120 to 140 Brinell is achieved. A maximum heating rate for small castings would be about 200° F. per hour, a sojourn at the annealing temperature not exceeding 1 hour per inch of section, and a cooling rate down to 500° F. of less than 100° F. per hour. Certain kinds of hardware, typewriter parts, instrument housings and electrical equipment are typical castings.

Many thin castings produced by permanent (metal) molds may be processed through the "B" or "C" treatments to achieve toughness or machinability. Centrifugally chill cast pipe and bushings are examples. The high temperature "C" treatment requires heating to "malleabilizing" temperatures to produce machinability in castings containing "chilled iron" structures, or "white iron" in their surfaces or edges.

"Graphitizing" the Chill

Heating at 1700-1800° F. (for 1 to 3 hours plus 1 hour per inch of section) will usually break down the chill, provided chill stabilizing elements such as sulphur, chronium, molybdenum, vanadium, et cetera, are not present in excess. The anneal "graphitizes the chill" and restores machinability.

The cooling rate may be timed to (1) conserve the strength by air cooling from 1700° F. to 1000° F.; then furnace cooling at a rate not exceeding 100° F. per hour to eliminate residual stress, or (2) cool from 1700° F. in the furnace or at a rate of less than 100° F. per hour to obtain a maximum softening comparable to treatment "B" above.

Table 1

SUMMARY OF STRESS RELIEF ANNEAL TREATMENTS

Heating Rate Per Hr.-Max.	*Stress Relief*	*Temperatures*	*Holding Time per Inch of Thickness*	*Maximum Cooling Rate*
200° F.	Slow, Low Temp. Anneal....	800- 900° F.	1 Hr. + 5*	100° F./Hr.
200° F.	Typical for Soft, High Silicon Irons	900-1050° F.	1 — 2 Hrs.	100° F./Hr.
200° F.	Typical for High Strength Irons	1050-1150° F.	½ — 1 Hr.	100° F./Hr.

*Viz—for 4-in. section; total time = 9 hr.

Heat resisting castings such as engine exhaust manifolds may be alloyed to possess a high resistance to heat. Then they may be annealed as described, for machining, while their alloy content remains to deliver substantial improvement in subsequent performance.

In other cases this heat treatment is mainly used as a reclamation measure upon castings whose composition is out of balance in terms of susceptibility to chill.

Annealing is rarely employed in regular practice because foundrymen are aware of the need to maintain machinability in castings and are skilled in the methods for achieving it. Soft irons can be made with Brinell hardness as low as 140, and used without annealing where strength properties are unimportant and machinability is a dominant factor.

Quenching and Tempering. Quenching is described as a process of rapid cooling from an elevated temperature by contact with liquids, gases or solids. In the quenched condition, castings are relatively brittle and weak (Fig. 2), and for most conditions of service it is necessary to toughen them by a reheating process called "tempering."

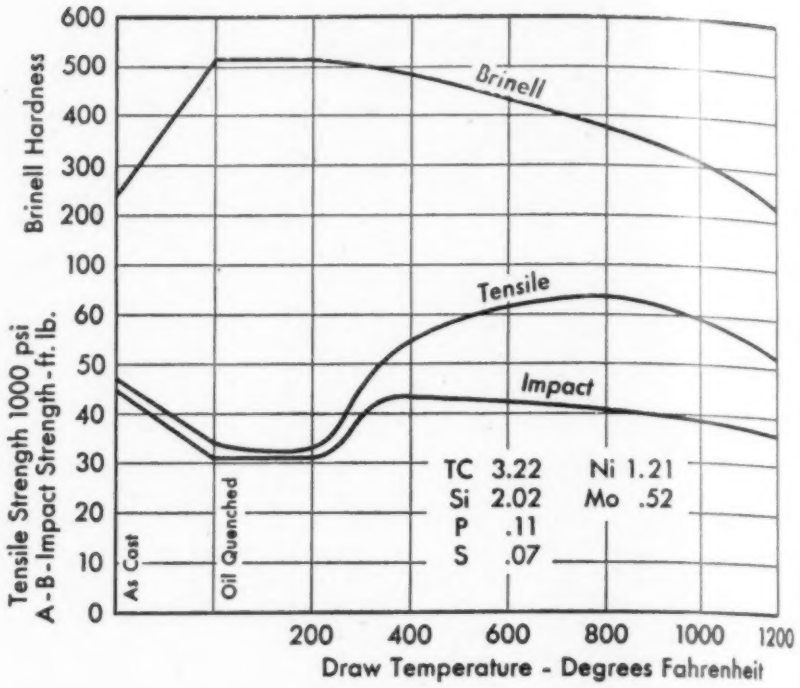


Fig. 2—Change in properties with drawing temperature for a nickel-molybdenum cast iron.

High Hardness

Tempering is described as the process of reheating a hardened cast iron to a temperature below the transformation temperature range, followed by any desired rate of cooling. The engineering properties obtainable are summarized in Table 3, and it is important to point out that the high strength is developed by the high temperature tempering, while the high hardness is developed by the low temperature tempering (Fig. 2).

Most of the discussions on quenching and tempering feature the strength properties that are developed, while practical operations upon castings are mainly directed toward obtaining a high hardness for resistance to wear and abrasion. One reason for this situation rests in the fact that high strengths can be produced by alloying or other means in castings without quenching treatments and with greater security in the case of complicated shapes. High hardness can usually be obtained only by heat treatments involving quenching. Descriptions in more detail follow.

When cast irons are quenched from above the critical range they acquire an increase in hardness of from 100 to 300 Brinell units. A soft, weak iron high in graphite content will show a lesser response to this hardening treatment than will a composition that is stronger and harder "as cast."

Thus, a soft iron of relatively high carbon content may be hardened from a level of 175 Brinell to approximately 325 Brinell by applying a quenching treatment, while a high strength, lower carbon type may move up in hardness from 220 Brinell to 450 Brinell. With the help of suitable alloys these hardnesses can be increased an adidtional 100 units to top levels of 500-550 Brinell.

Table 2

ANNEALING TREATMENTS

	A Low Temp.	*B High Temp.*	*C Chill Elimination*
° F.	1300-1400	1500-1600	1700-1800
Holding Time—Hours per in. of section	1 — 3	½ — 3	1 — 5
Cooling	Slow	Slow	Slow

The common quenching media are air, oil and water.

The usual quenching medium for cast iron is an oil bath because it provides a faster cooling rate than is obtainable in air, and a slow enough rate to avoid temperature stresses and gradients of a magnitude that might cause cracking. Most of the castings that are heat treated to develop high hardness are quenched in oil from above the critical temperature range.

Holding Time

The usual procedure consists of heating the castings to approximately 1550-1600° F. Preheating in a separate furnace to about 1100° F. reduces the risk of cracking due to temperature differentials and cuts down the time of heating at the higher temperature, thus reducing scaling. This procedure is rarely used, and the more common practice consists of heating from a relatively low temperature to the quenching temperature in the same furnace.

Castings need to be held only long enough to acquire a uniform temperature, at which time they should be quenched. Tempering should preferably follow immediately after quenching, and for best practice the casting may be at a temperature of

Table 3

APPROXIMATE TENSILE STRENGTH AND HARDNESS OF HEAT TREATED VS. "AS CAST" IRONS

Tensile Strength		Hardness	
"As Cast"	Heat Treated for High Strength	"As Cast"	Heat Treated for High Hardness
65,000	85,000	280	530
55,000	75,000	250	500
40,000	60,000	220	480
30,000	45,000	190	430
20,000	30,000	165	380

200-300° F. at the time it enters the tempering furnace. Castings that are heat treated for high strength usually possess a higher hardness after heat treatment than in their "as cast" condition.

Since a casting is expected to possess about the same degree of toughness that it had in the "as cast" condition, it is necessary to acquire this toughness in quenched castings by tempering after quenching. Figure 2 shows that by tempering in the range of 350-500°F. the hardness is not appreciably changed from the quenched condition, while the strength and toughness have been improved to approach their "as cast" levels.

High Abrasion Resistance

With the alloyed irons it may be possible to heat to as high as 750° F. and retain a high level of hardness, accompanied by a further improvement in strength over the "as cast" product. Castings requiring a high resistance to abrasion which is met by a suitable microstructure at a hardness usually above 400 Brinell, are tempered after quenching by reheating them in the low temperature range of 300-800° F. (Fig. 1).

The time at tempering usually does not exceed one hour per inch of section after the casting has acquired the desired temperature. Castings may be cooled in the furnace after tempering, thus eliminating any risk of acquiring additional internal stress due to irregular cooling. However, air cooling from the low temperatures to which castings are heated on tempering, rarely produces a cooling rate that is rapid enough to cause distortion or difficulties due to internal stress.

Engine cylinder liners are the most common type of casting processed through the quenching and tempering heat treatment. Others are cams, gears, rollers, dies, polishing and grinding equipment, hardware, and numerous small machinery parts.

The cooling rate in air quenching is usually so slow that only very thin castings are successfully treated by this means, and then only if alloys which retard the transformations are present in adequate amounts. Heavy sectioned castings such as dies may similarly be air cooled to acquire some hardening, and because of the slow rate of heat dissipation, require much larger proportions of alloys in their composition.

Water Quenching

Quenching in water imparts a cooling rate to cast iron which is too drastic in most cases for the castings to successfully survive the stresses raised in their surfaces by the difference in temperature between surface and underlying metal and the limited toughness of the base metal itself. Therefore, water quenching is rarely practiced.

There are special heat treatments such as flame hardening and induction hardening where the skin of a casting is locally heated to a depth usually not exceeding 0.125 in. and quenched by a water spray immediately following the heating operation. Most of these castings are not subsequently tempered, and the successful use of this hardened surface is made possible through the shallow zone that has been treated and the absorption of the accompanying stresses by the tougher, unheated, underlying metal.

The local application of heat by the flame hardening or induction hardening processes involves close attention to mechanical details and conditions of surface in order that the heat flow into the casting may be uniformly controlled. The procedure is limited to applications which possess contours, dimensions, and hardness requirements that are satisfied by a local and shallow hardening of accessible surfaces capable of absorbing the stresses accompanying rapid heating and cooling.

Apparently the properties of the water quenched zone are similar to those prevailing in quenched material prior to tempering. Typical castings which are heat treated in this manner are the wearing surfaces of lathe beds, certain gears and cams, rollers, plungers, cylinder liners, etc. These surface hardened castings are ordinarily used under conditions of loading in which surface pressures are relatively light and the need for a hard, strong underbody does not exist.

High Strength

It is possible to add 10,000-25,000 psi. additional to the strength of a cast iron that has been quenched in oil and tempered (Fig. 2). Tempering after quenching to produce a high strength consists of reheating the quenched castings to a temperature of 600-1000° F. Most cast irons acquire their maximum strength with this tempering treatment which leaves them some 20 to 100 Brinell units harder than they were in the "as cast" condition.

If the tempering treatment after quenching is carried out to restore the "as cast" hardness, it is usually found that the "as cast" strength is restored and no improvement from heat treating is obtained. Consequently, the machinability of these irons that have been heat treated for high strength is inferior to their machinability in the "as cast" condition. The advantages of the treatment rest in the possibility that machining operations may be performed on the softer "as cast" product, after which the stronger and harder castings can be completed by minor machining operations.

Little Change in Impact Strength

The impact strength of a cast iron is not appreciably changed by heat treatment, except in the poorer grades of iron where a low strength and a low combined carbon content exist in the "as cast" condition. In general, the impact resistance is less after quenching and tempering than existed in the original "as cast" condition. Any expectation of an increased impact resistance should be carefully investigated. The modulus of elasticity is similarly unchanged or not improved by the quenching and tempering type of heat treatment.

A special type of heat treatment, limited in its application to castings that can be quenched in a hot salt

or lead bath, has been used where a combination of hardness and toughness produces a desirable wear resisting structure. This is called a "hot quenching treatment," which treatment with the above explanation adequately defines the process.

In practice, castings of suitable composition are quenched from above the critical range into a bath at a temperature of 450-800° F. and held for sufficient time to obtain the desired change in structure. Usually 2 to 5 hours in the quenching bath is sufficient, but longer intervals have been tested out experimentally. The resulting castings possess a hardness in the range of 280-380 Brinell, and for their hardness level they posses a higher degree of toughness than occurs in castings that have been cold quenched and tempered to the same hardness value.

Toughness tests as measured by the deflection or A.B.[2] impact test indicate that these irons are much tougher than companion pieces that have been hardened by quenching and tempering in the usual way. Bartholomew[3] reports that at the same hardness as quenched and tempered castings, the hot quenched castings are able to carry twice the load for the same degree of wear, or, conversely, carry the same load two or three times as long. The operation has been limited to cams, gears and similar heat treatable castings whose service requires of them a high resistance to wear.

References

[1] American Society for Testing Materials, *Standards* of 1942, Part 1, p. 1090.

[2] Arbitration bar specimens 1.125 in. dia. by 8 in. long broken on 6 in. span in the Charpy test, or arbitration bar specimens 1.125 in. diam. by 3 in. above supports in the Izod test.

[3] Bartholomew, E. L., "Gray Cast Iron," *The Iron Age*, Aug. 1, 1940, pp. 52-54.

DISCUSSION

Chairman: H. BORNSTEIN, Deere & Co., Moline, Ill.

Co-Chairman: V. A. CROSBY, Climax Molybdenum Co., Detroit.

H. W. LOWNIE, JR.[1] *(written discussion):* The author has presented a general summary of the various heat-treating practices applied to cast iron, but has given only very limited and general information on the engineering properties of these irons as indicated by the title of the paper.

In discussing stress-relief annealing, it is important to emphasize that the *rate* of cooling from the annealing temperature is important only as it affects the *uniformity* of temperature within a particular casting. Castings of uniform section thickness may be effectively stress relieved at surprisingly rapid cooling rates *if* attention is given to the arrangement of the castings within the furnace and to forced circulation of the furnace atmosphere. Careless piling of these same castings into a poorly designed furnace, or into a furnace lacking in forced circulation, would require extremely slow cooling to obtain effective stress relief.

To be effective, stress-relief annealing should be conducted under careful supervision and with effective control. Poor stress-relieving practices actually increase residual stresses. If the application of a casting justifies the expense of stress-relief annealing, the application then also justifies a determination of the efficiency of the practice being used.

An experience of the writer with a small electric motor showed the pronounced effect that heat treatment may have upon the wear resistance of gray cast iron. The motor frame was of light section, about 5 in. long, 2 in. in diameter, and contained an integrally cast cylindrical brake ring at one end. Stellite brake shoes mounted upon the shaft were used to stop the motor quickly by friction upon the cast iron brake ring. The application required that the brake mechanism pass a service test involving 20,000 stops. With the motor frames in the as-cast condition, the Stellite shoes cut completely through the cast iron ring in less than 5,000 operations. Foundry attempts to increase the wear resistance of the iron were unsuccessful as the machinability had to be kept high to permit drilling and tapping of a number of small holes.

Several assembled motors were taken from stock and disassembled so that the frames could be heat treated. Tapped holes were plugged and the ends of the frames containing the brake ring were heated by immersion in a salt bath and then water quenched. The frames were tempered at a low temperature. Distortion during these operations was so low that the internal components of the motors were reassembled into the hardened frames and ran perfectly. Each of the motors with hardened brake rings completed 75,000 operations and, in each case, the Stellite shoes failed when the cast iron brake ring showed only slight wear.

MR. VANICK *(answer to Mr. Lownie's written discussion):* Mr. Lownie's experiences with heat treating cast iron are worthy contributions. Emphasis on the need to maintain uniformity in cooling during the stress-relief anneal deserves attention.

The theoretical premise that if every portion (internal as well as external) of a casting cools uniformly, there will be no locked up stress, may be approached in small, thin castings where thermal conductivity becomes a negligible factor; but in almost all commercial practices, the elimination of a temperature gradient in the casting can be best accomplished by a furnace cooling rate that does not exceed by more than 50 to 100° F. the cooling rate of the thickest section or greatest heat-retaining section of the casting.

The description of the water quenched and tempered motor brake rings adds another successful application of the heat treatment of cast iron to parts that require good machinability "as cast" and hardness for resistance to wear when assembled.

J. H. SCHAUM[2]: We are conducting stress-relief experiments at the Naval Research Laboratory with the aid of a relaxation machine which shows the rate at which stresses are removed at different temperature levels. In this work it has been found that the first hour of stress relief removes the greatest percentage of the stresses and any time after that is almost wasted because of the asymptotic shape of the time-stress curve.

Low temperatures such as 800° F. and 900° F., mentioned by Mr. Vanick are quite often ineffective. For example, one of the plain carbon gray irons tested at our laboratory showed a stress reduction of only 4,000 psi. (from 15,500 psi. to 11,500 psi.) in one hour at 900° F. Seventeen hours were required to relieve an additional 4,000 psi. of stress. Such a long time at temperature would be uneconomical. This same iron, when heated to 1050° F., had the stress reduced from 12,000 psi. to 2,000 psi. in 1 hr. For most applications 2,000 psi. residual stress would be of little consequence, thus eliminating any need for continuing the heat treatment for the 24 hr. required to remove this last 2,000 psi. of stress.

MR. MCELWEE: We should recognize that if we get a dendritic pattern in an iron that is originally cast we can not expect to improve that structure. In other words, the as-cast structure must be good.

CO-CHAIRMAN CROSBY: In other words, he should remelt it.

[1] Battelle Memorial Institute, Columbus, Ohio.
[2] Naval Research Laboratory, Washington, D. C.

Effect of Certain Elements on Mechanical Properties of 1% Cu, 5% Si, and 0.5% Mg Aluminum Sand Casting Alloy

An investigation of the effect of certain elements on the mechanical properties of an aluminum sand casting alloy with a composition of 1 per cent copper, 5 per cent silicon, and 0.5 per cent magnesium.

R. A. Quadt

American Smelting & Refining Co.
Barber, N. J.

THIS INVESTIGATION was conducted on a group of aluminum sand casting alloys with a nominal composition of 1.0-1.5 per cent copper, 4.5-5.5 per cent silicon, and 0.4-0.6 per cent magnesium. It is described as A.F.A. Alloy B-41[1]. Specifications by other agencies include QQ-A-601, class 10; SAE 322; ASTM B26-44T (Alloy SC 21); AN-QQ-A-376 Am.2; and AMS 4210.

For convenience, this alloy will be referred to in this paper as F150.5 in accordance with a system of nomenclature in which aluminum casting alloys are defined according to their copper, silicon, and magnesium content to the nearest integer in the indicated sequence.

Magnesium

Because of the effectiveness of small amounts of magnesium in these alloys, this element is indicated in tenths of a per cent. The actual alloys used are representative commercial composition within the F150.5 specifications. The mechanical properties specified for this alloy are listed in Table 1.

As will be subsequently proven, the solution treated alloy is unstable and begins to age at room temperature immediately after quenching. After a few weeks the properties of the quenched alloy approach those of the alloy aged 3-5 hr. at 310° F. (Table 1).

This hardening process is accompanied by dimensional changes and, for this reason, the as-quenched alloy is seldom used commercially and an artificial aging treatment similar to that of the 3-5 hr. at 310° F. (Table 1) is utilized to obtain properties which are essentially stable upon subsequent room-temperature aging. This is the most commonly used heat treated temper for this alloy.

Heat treatments similar to the solution treated, quenched, and aged 12 hr. at 310° F. (Table 1) are occasionally used when high strength properties are required for specific applications. No specifications exist for the alloy in the as-cast condition although excellent properties are obtained. The as-cast alloy may be stabilized and stress relieved by heating to a temperature of 450° F. for 5 hr. and air cooling.

Various similar stress relief treatments are used commercially, but the work described in this paper was limited to this one treatment when stabilized properties were studied.

F150.5 is a comparatively new aluminum alloy which has been used in large tonnages in the past few years, particularly in air craft castings. This widespread acceptance of the alloy in aluminum foundries stems from its desirable combination of chemical and physical properties.

It has excellent foundry characteristics in that it is extremely fluid at low casting temperatures, produces pressure tight castings, casts readily into thin sections, has low solidification shrinkage, as compared to other aluminum casting alloys, and is essentially free from hot shortness.

Corrosion Resistant

It machines well, especially after heat treatment, and has high resistance to corrosion in saline atmospheres. It is an excellent general purpose alloy in that a foundry may use it for all types of castings since a wide range of mechanical properties can be obtained by using the alloy in the as-cast condition or in various heat treated tempers.

Melting and Casting Procedure. All melts were made in No. 45 clay-graphite crucibles in a coke-fired,

Table 1

ALUMINUM ALLOY MECHANICAL PROPERTIES

Heat Treatment	Properties: Tensile Strength (min.), psi.	Avg. Yield Strength* (0.2% offset), psi.	Elongation (min.), per cent in 2 in.
Solution Treated and Quenched	27,000	20,000	4.0
Solution Treated, Quenched, and Aged 3-5 hr. at 310° F.	32,000	25,000	2.0
Solution Treated, Quenched, and Aged 12 hr. at 310° F.	36,000	33,000	0.5

*Yield strengths almost invariably are listed as "average" or "typical" in specifications of this and other aluminum casting alloys.

Presented at an Aluminum and Magnesium Session of the Fiftieth Annual Meeting, American Foundrymen's Association, at Cleveland, May 7, 1946.

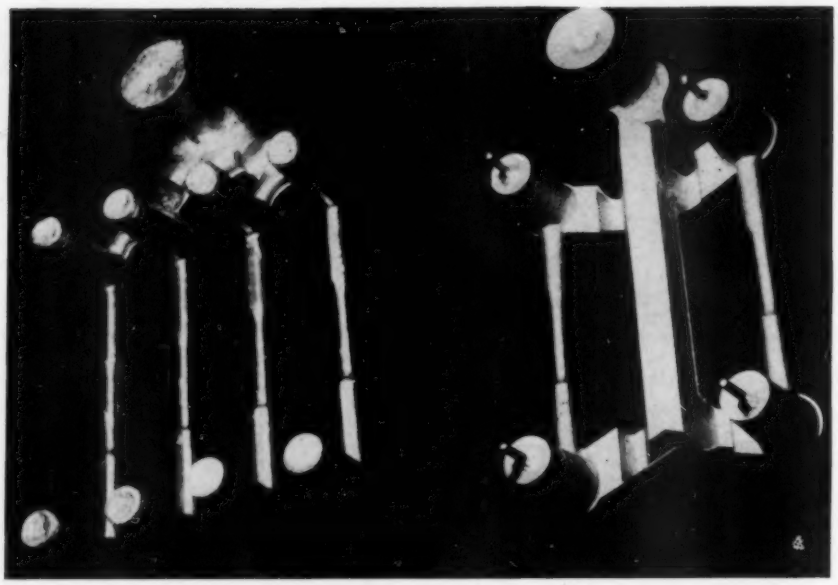


Fig. 1—Types of castings utilized to obtain cast to size tensile specimens. About 1/7 actual size.

forced draft, pit furnace. About 30-45 min. was required to melt 50 lb. of metal. Ingots or pieces of test bar castings were continually added to the melt through the top cover of the furnace during melt down. This guaranteed that no moisture was present on the metal by the time it was submerged in the previously melted alloy.

Surface hydrates were assumed to be negligible due to short intervals of storage and to favorable storage environment. When all of the metal was in the crucible and the hardener or alloying element, if any, completely dissolved, the metal was stirred with a one in. diameter graphite rod. The crucible was then removed from the furnace. Typical metal temperatures at this time ranged from 1400 to 1600° F. A few ounces of sodium silico fluoride was gently worked into the top dross to assist in its removal.

Metal Cooled

The metal was then permitted to cool quietly to the desired pouring temperature, nominally 1220° F. unless otherwise specified. No degassing operation was performed and no grain refiners were added to any of the heats reported. Temperatures were determined with a protected chromel-alumel thermocouple in conjunction with a portable Leeds and Northrup potentiometer.

The metal was cast into green sand molds to produce standard ½-in. diameter cast-to-size tensile specimens. Two types of castings were used and are shown in Fig. 1. No difference in mechanical properties could be detected in the specimens from either type of casting. About 75 per cent of the bars cast were obtained from the 4-bar casting since twice as many bars per lb. of aluminum are produced and fewer instances of inclusions experienced.

Specimens Poured

Cast-to-size impact specimens were also poured from most of the heats. The casting is shown in Fig. 2 and produces twenty-six 10x10 mm. Charpy specimens. Half of the specimens are unnotched and the rest have a standard V-notch cast into the drag. The specimens were tested unmachined on a modified Charpy impact machine of 16 ft.-lb. capacity. The production of impact specimens by this method has been discussed elsewhere[2].

Heat Treatment Practice. Solution heat treatments were conducted in an electrically heated, circulating air furnace. Bars were nominally soaked 12 to 15 hr. at 980° F. ± 5° F. and quenched in boiling water.

To standardize the procedure, bars were aged for one day at room temperature before being tested for solution heat treated properties or before being artificially aged at elevated temperatures. The stress relief treatment given as-cast specimens consisted of heating for 5 hr. at 450° F. and air cooling.

All low temperature treatments were carried out in electrically heated, constant temperature fused salt baths. The temperature was automatically controlled to within ± 2° on each of three baths set for 310° F., 400° F., and 450° F.

Properties listed for the as-cast temper were obtained from one to 4 days after casting unless otherwise noted.

Alloying Procedure. Since magnesium, nickel, and zinc readily dissolve in aluminum they were added as pure metals. Iron and manganese were added as hardeners, which were made by melting a few lb. of the alloy under investigation and adding the desired quantity of these elements. By this procedure dilution of the initial composition was minimized.

Chemical determinations were made on the final test bars, not on the intermediate hardener. Whenever an alloy was remelted, 0.03 per cent magnesium was added to make up for the amount lost in the process. Occasional chemical determinations were made to check the magnesium content of the final alloy since this element has a critical effect on the properties obtained.

Effect of Pouring Temperature and Overheating. An alloy of the

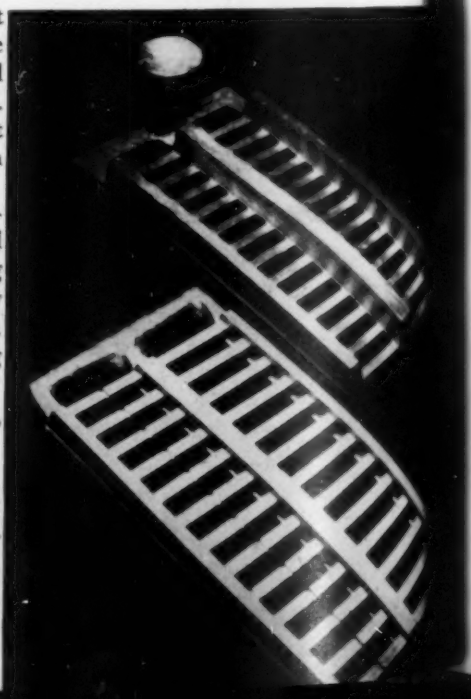


Fig. 2—Casting for producing cast to size 10x10—mm. Charpy impact specimens. Top—Cope view. Bottom—Drag view.

composition shown in Table 2 was used to determine the effect of casting temperature on some properties in the as-cast or "green" temper.

The alloy was melted and heated to a temperature of about 1700° F. After removal from the furnace and skimming, a series of castings was poured at various temperatures. The melt was permitted to cool quietly in the crucible between pours.

That the pouring temperature is of considerable significance on the properties of a casting is clearly shown in Fig. 3. It is also indicated that overheating the alloy does not affect the final properties if the metal is permitted to cool to the proper casting temperature.

This absence of any effect on the final properties due to previously overheating the metal has been checked on most other common aluminum casting alloys with the same result.

Some data to this effect have been reported elsewhere[3]. This does not concur with the widely promulgated assertion that overheating (in excess of 1500° F.) aluminum casting alloys favors coarsening of the final grain with a resultant deleterious effect on properties[4,5,6]. This result has never been experienced by the author.

Grain Size

In the absence of an active grain refiner, many tests on various aluminum alloys have indicated that grain size and mechanical properties of test bars are largely determined by the final casting temperature and are little if at all related to the thermal history of the melt. This may not be true of high-purity alloys since all tests were conducted on commercial alloys with analyses in the commercial range.

For such commercial alloys the grain size of tensile bars can be controlled by pouring temperature alone and superheating is not a factor. Should titanium be present and still retain grain refining properties, overheating, or even holding at moderate temperatures, should be avoided since these procedures reduce the effect of the titanium and a coarser grained casting will result.

However, it is important to note that this grain coarsening is due to the effect of temperature upon the titanium constituent and not upon the aluminum.

Hydrogen Free

It is of utmost importance to recognize, however, that these experiments were conducted in furnace atmospheres which were essentially free of hydrogen compounds, hence typically sound castings resulted.

In a furnace atmosphere containing combustion products from gas, oil, or coal, overheating should be kept to a minimum since the solubility of the hydrogen obtained from the compounds present increases with the temperature of molten aluminum, and excessive porosity may result which will cause inferior properties. These experiments do prove that overheating, other factors being equal, is not deleterious.

Another test which indicates the desirability for low pouring temperatures was performed on the alloy composition shown in Table 3.

It was melted in the described manner, cast at 1220° F. and the specimens heat treated and tested. Upon completion of these tests the alloy was remelted and recast at 1300° F. Similar mechanical tests were conducted on the alloy. The results of these tests are plotted in Fig. 4. Ultimate strength has been markedly reduced, ductility has been slightly impaired, while yield strength is unaffected. Thus a pouring temperature of about 100° F. above the optimum detracts from the quality of the casting.

At least a part of the reason for the inferior properties accruing from a high pouring temperature is the resulting increased grain size. Fig. 5 portrays the difference in crystal structure between tensile specimens cast at 1220° F. and at 1300° F.

It will be noted that Fig. 5 indicates a greater difference in grain size between the two pouring temperatures than would be predicted from the curve given in Fig. 3. The latter data were obtained from two-bar castings while the former resulted from the four-bar type. The metal for the rear risers of the four-bar casting passes through the tensile section and heats the sand.

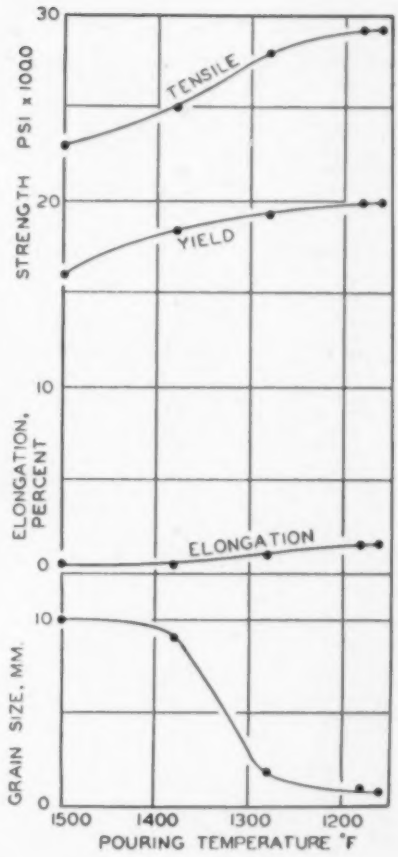


Fig. 3—Effect of pouring temperature on as-cast properties. Specimens poured from the same crucible.

Retards Freezing Rate

This retards the freezing rate of the bars appreciably, and at elevated temperatures a coarser grain results. Utilization of this type of casting displaces the grain size curve shown

Table 2

ALLOY COMPOSITION

Component	*Per Cent*
Copper	1.68
Silicon	5.13
Magnesium	0.58
Manganese	0.42
Iron	0.74
Nickel	0.04
Zinc	0.33
Chromium	0.005
Titanium	0.005
Aluminum	Balance

Table 3

ALLOY COMPOSITION

Component	*Per Cent*
Copper	1.23
Silicon	5.21
Magnesium	0.46
Manganese	0.43
Iron	0.54
Nickel	0.02
Zinc	0.35
Chromium	0.10
Titanium	0.13
Aluminum	Balance

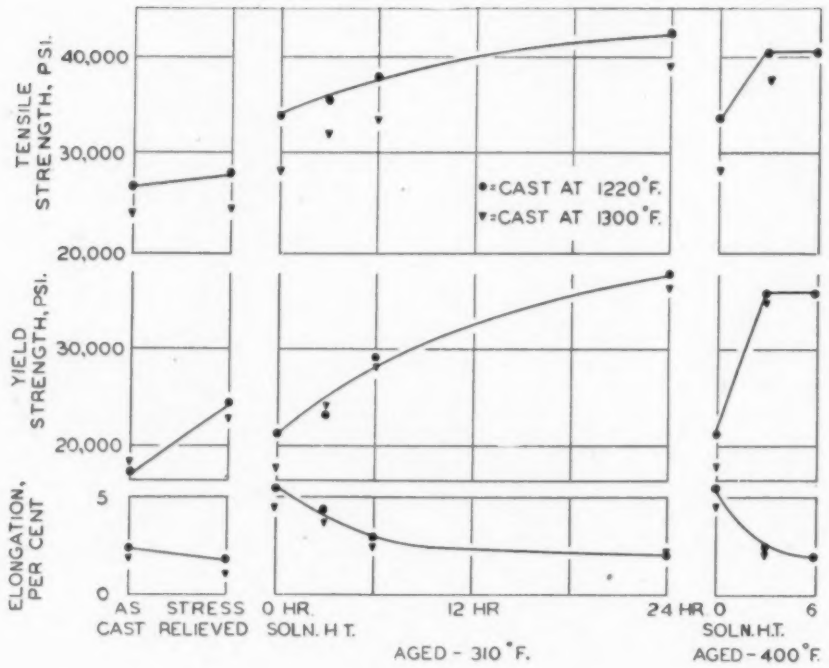


Fig. 4—Effect of pouring temperature on some mechanical properties.

in Fig. 3 to the right. Occasionally tensile bars from this casting will show a grain size gradient ranging from coarse at the sprue end to relatively fine at the rear risers.

In the two-bar casting, on the contrary, two streams of metal meet in the center of the reduced section and the freezing rate must be maximum at this point. A grain size gradient has never been observed in bars from this type of casting.

Effect of Magnesium. F150.5 alloy responds to heat treatment largely because of the magnesium and silicon in the composition. Of these two elements the magnesium content is most critical in determining the alloy's mechanical properties, whether heat treated or not. In a test to prove this fact, some properties of the alloy shown in Table 4 were determined.

Alloy Remelted

It was then remelted and the magnesium removed. The properties were redetermined on the resulting alloy, which then contained 0.002 per cent magnesium, and are compared in Fig. 6. It is seen that the initial alloy reacts to heat treatment while the same alloy without magnesium shows essentially no response to heat treatment.

Both sets of properties shown in Fig. 6 are lower than would normally be expected because the data were obtained for another purpose from test bars cast at 1350° F., an excessively high temperature for pouring test bars from this alloy. The data reflect the effect of the magnesium content, notwithstanding.

Aging Treatments

On another alloy of the composition shown in Table 5 the effect of magnesium was further investigated. After mechanical properties were obtained on this alloy it was remelted and magnesium added, which raised this element to 0.56 per cent. The effect of various aging treatments on solution treated specimens of the two alloys is plotted in Fig. 7.

Marked instability of the as-quenched alloy on standing at room temperature is indicated. A large amount of aging occurs the first day, followed by further aging at a reduced rate. The hardening process

Table 4

ALLOY COMPOSITION

Component	*Per Cent*
Copper	1.68
Silicon	5.13
Magnesium	0.53
Manganese	0.34
Iron	0.79
Nickel	0.00
Zinc	0.30
Chromium	0.005
Titanium	0.005
Aluminum	Balance

is largely completed in a few weeks. It is clear that the magnesium content is critical in this alloy.

Addition Made

Since the element is slowly lost on remelting and fluxing, an addition must be made to maintain correct composition. In general, increasing magnesium contents in this alloy produce higher tensile and yield strengths combined with reduced ductility for any particular temper.

Effect of Nickel. Chemical specifications variously require that nickel be limited to 0.05 or to 0.10 per cent in this alloy. Figs. 8, 9 and 10 show the effect of increasing nickel on the F150.5 alloy shown in Table 6. These data indicate that nickel as high as 0.3 per cent has essentially no effect on the alloy. More than this amount generally impairs most properties.

A summary of the effects of high nickel is as follows: The strength is reduced, tensile and yield strengths approach each other, and serious reductions in elongation and impact strength occur. As a concession to clarity the impact strengths in these and in some subsequent plots are averaged for the tempers being considered.

The nickel can be detected in the microstructure even when only 0.1 per cent is present. It is apparently almost insoluble in the solid alloy.



Fig. 5—Axial tensile bar sections, 2/3 size. Etchant, 1 per cent HF. Top—Grain structure of casting poured at 1220° F. Bottom—Grain structure of casting poured at 1300° F.

Table 5

ALLOY COMPOSITION

Component	Per Cent
Copper	1.58
Silicon	5.22
Magnesium	0.18
Manganese	0.50
Iron	0.67
Nickel	0.04
Zinc	0.28
Chromium	0.10
Titanium	0.01
Aluminum	Balance

It is found only in the eutectic structure, probably as the compound $NiAl_3$ (Fig. 11). Heat treatment has no effect upon its structure or appearance.

Effect of Zinc. Zinc in F150.5 alloy is limited to maximums ranging from 0.03 to 0.30 per cent, depending upon the chemical specification consulted. The curves in Figs. 12, 13, and 14 indicate the effect of increasing zinc on some properties of the initial alloy shown in Table 7.

It is obvious that no deleterious effect accrues from zinc additions until more than one per cent is present in the alloy. The microstructure reveals no change in the form or occurrence of the usual constituents as the zinc content is increased to 2.5 per cent. This is not surprising since relatively large amounts of zinc are soluble in the solid alloy at room temperature even under conditions of thermal equilibrium.

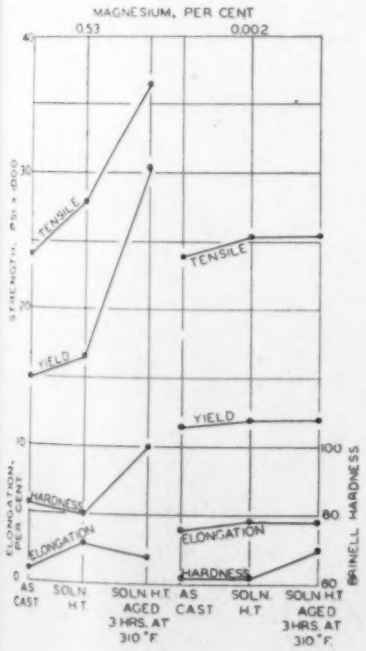


Fig. 6—Effect of magnesium on mechanical properties and heat treatability.

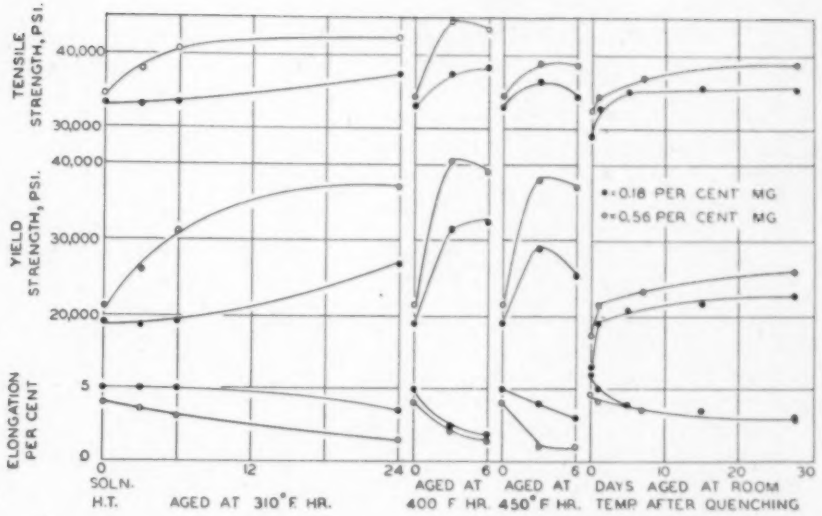


Fig. 7—Effect of magnesium on some heat treated mechanical properties.

Effect of Manganese. This element is limited in amounts varying from 0.03 to 0.6 per cent in this alloy by the existing chemical specifications. Figures 15, 16, 17 and 18 summarize the resulting properties obtained upon adding manganese to the alloy of the composition shown in Table 8.

Up to one per cent of manganese generally improves the mechanical properties of F150.5 alloy in the tempers most commonly used. It does, however, accelerate the rate of overaging upon subjecting the solution treated alloy to temperatures of 400° F. or over, although this tendency is not pronounced with less than one per cent of manganese.

Microstructure Changed

Two changes were noted in the microstructure as a result of the manganese additions. The structure of the initial alloy included needles of beta (Fe-Si), utilizing the nomenclature of Phillips and Varley[9].

Addition of 0.3 per cent of manganese eliminated these needles or plates, and the secondary alpha (Fe-Si) phase was apparent in its usual "Chinese script" habit. The structural change and its effect will be more completely discussed later. The alloy containing 1.65 per cent of manganese revealed masses of alpha (Mn-Si) instead of the previous script, and is shown in Fig. 19.

Effect of Iron. It is generally required that iron be under 0.5 or 0.6 per cent in sand castings of F150.5 alloy. The effect of iron was investigated by adding it in various increments to the initial alloy shown in Table 9.

Results of these tests are plotted in Figs. 20, 21, 22 and 23. Generally, iron under one per cent has no effect on the properties measured. As iron is increased over this amount it has only slight influence on properties except upon aging at the relatively high temperature of 450° F. It decidely influences the rate of overaging at this temperature, increasing it to a marked degree.

Effect of Iron

However, iron as high as 2.4 per cent has only an insignificant effect on the properties measured in all the other tempers. Curiously enough, this high iron appears to retard the rate of aging to a slight degree upon precipitation hardening at 310° F.

At the same time, it has no effect whatever when solution treated and aged at 400° F. and, in fact, appears to slightly improve the strength of

Table 6

ALLOY COMPOSITION

Component	Per Cent
Copper	1.35
Silicon	5.52
Magnesium	0.55
Manganese	0.01
Iron	0.34
Nickel	0.01
Zinc	0.04
Chromium	0.10
Titanium	0.05
Aluminum	Balance

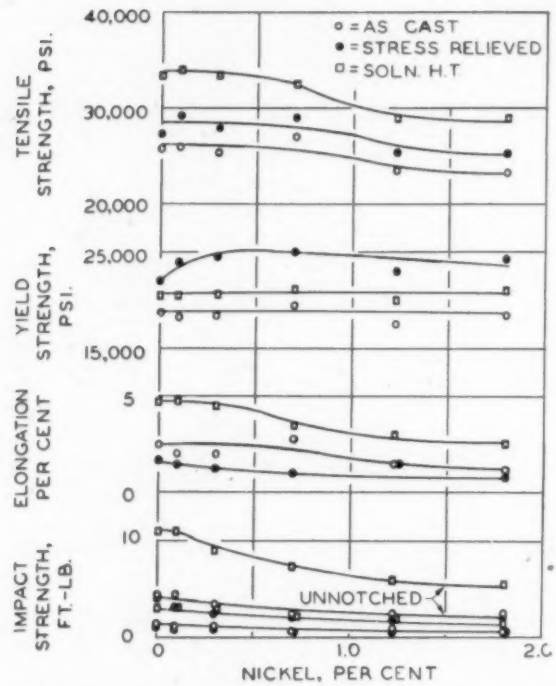


Fig. 8—Effect of nickel on some properties in the as-cast, stress relieved, and solution heat treated tempers.

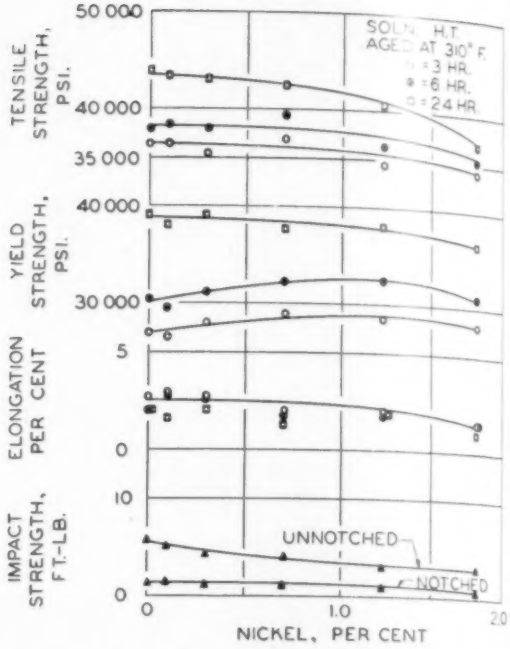


Fig. 9—Effect of nickel upon solution treating and aging at 310° F.

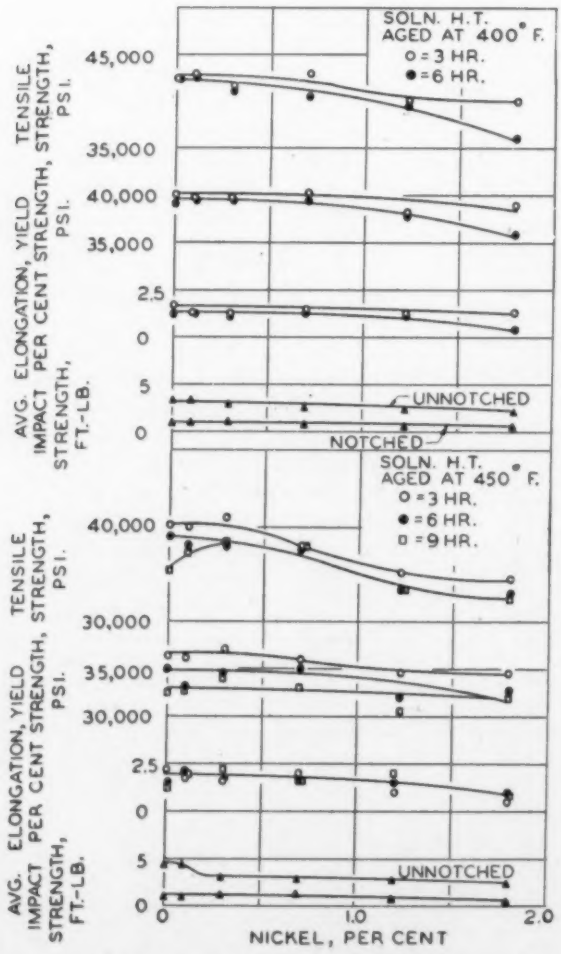


Fig. 10—Effect of nickel upon solution treating and aging at 400° F. and 450° F.



Fig. 11—As-cast structure of an F150.5 alloy with 1.8 per cent nickel. $NiAl_3$ (black), eutectic (mottled), silicon (bright), dendritic alpha (speckled). Etchant, 1 per cent HF. 300X.

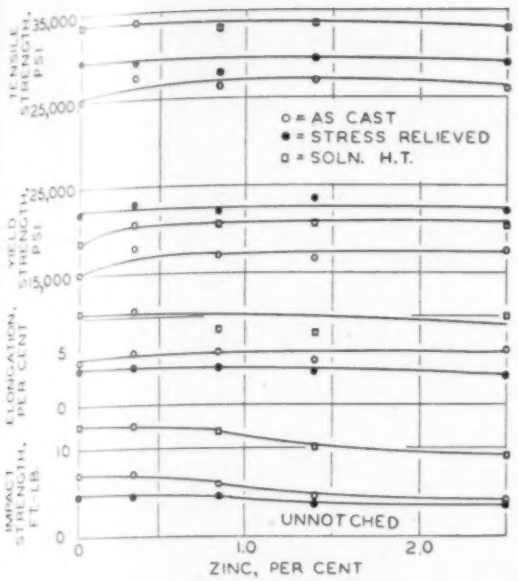


Fig. 12—Effect of zinc on as-cast, stress relieved, and solution heat treated properties.

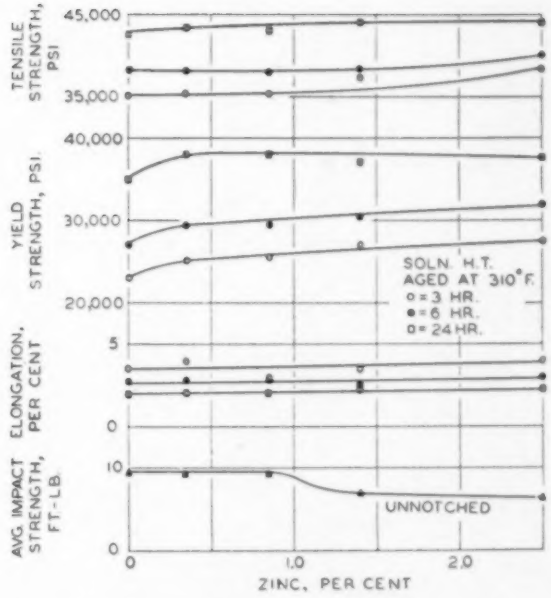


Fig. 13—Effect of zinc upon solution treating and aging at 310° F.

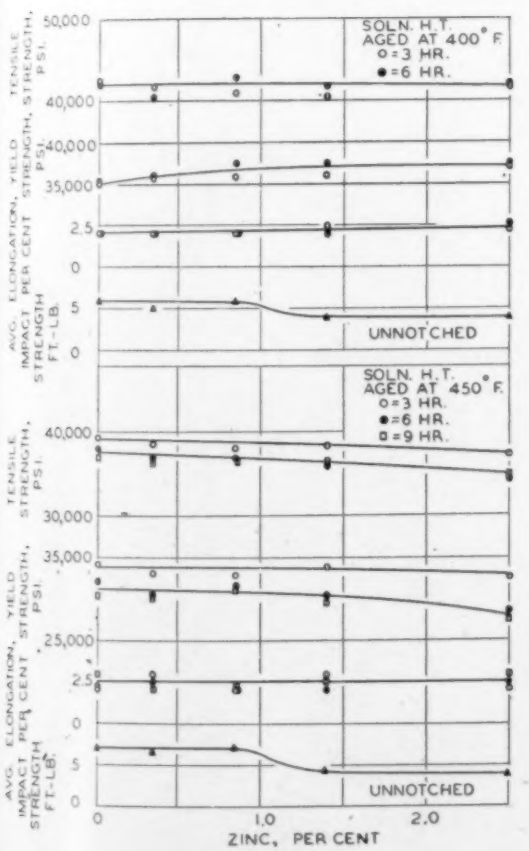


Fig. 14—Effect of zinc upon solution treating and aging at 400° F. and 450° F.

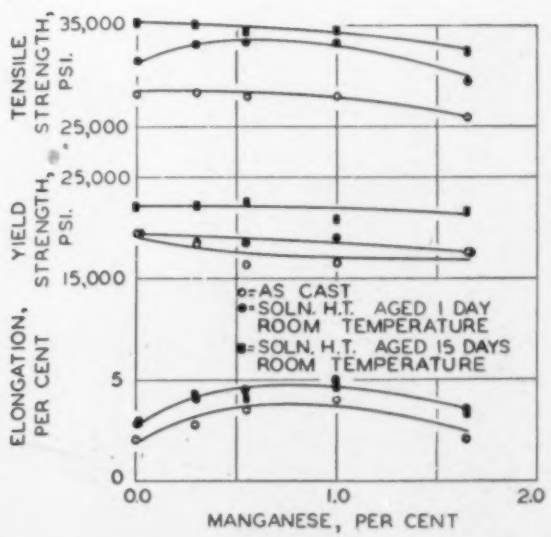


Fig. 15—Effect of manganese on as-cast and solution heat treated properties.

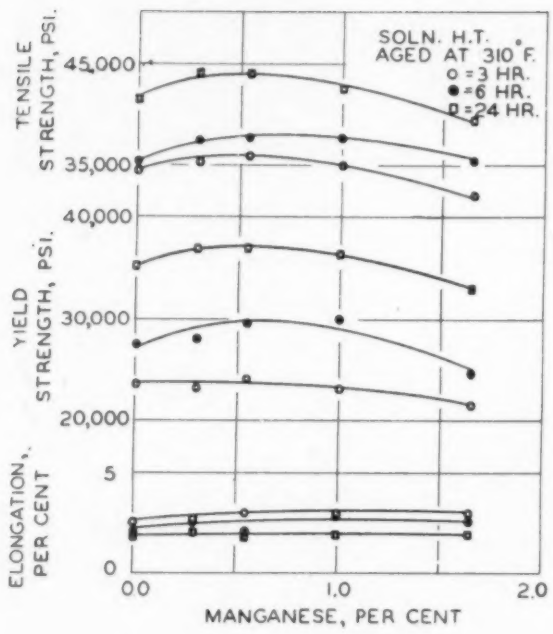


Fig. 16—Effect of manganese upon solution treating and aging at 310° F.

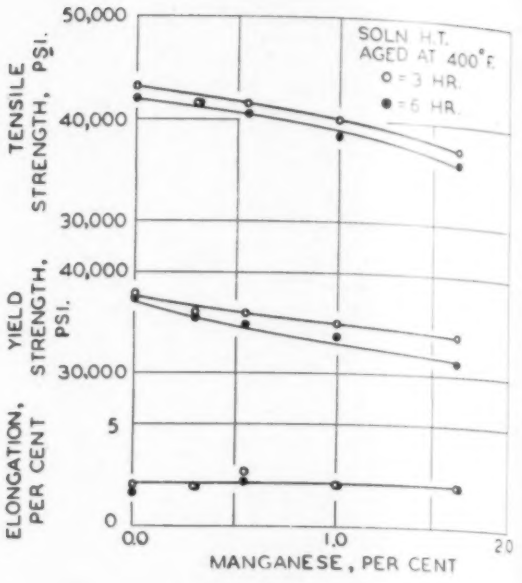


Fig. 17—Effect of manganese upon solution treating and aging at 400° F.

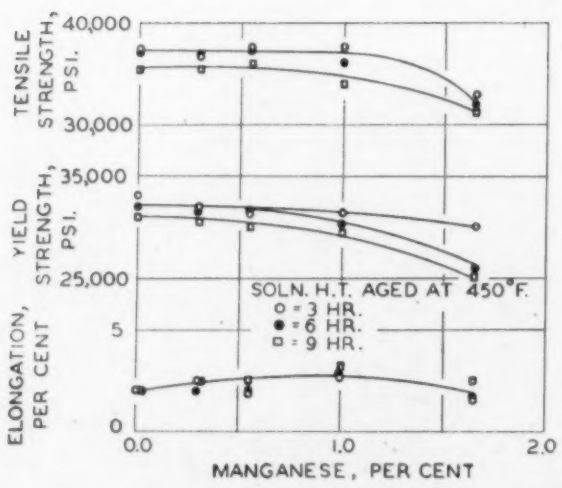


Fig. 18—Effect of manganese upon solution treating and aging at 450° F.



Fig. 19—As-quenched alloy with 1.65 per cent manganese. Alpha Mn-Si (in relief), silicon (half tone). Etchant, 1 per cent HF. 300X.

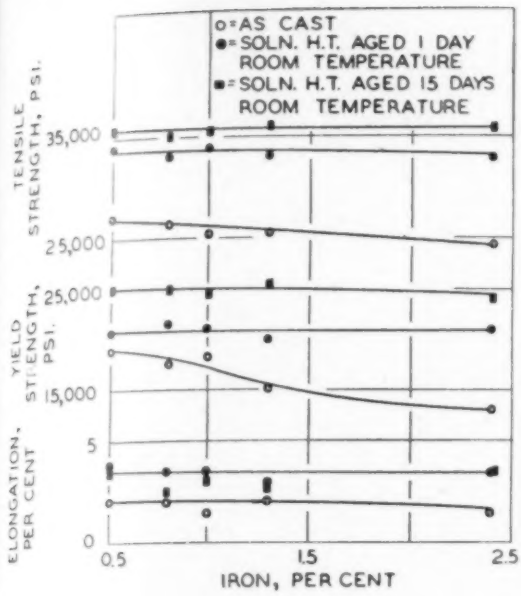


Fig. 20—Effect of iron on as-cast and solution heat treated properties.

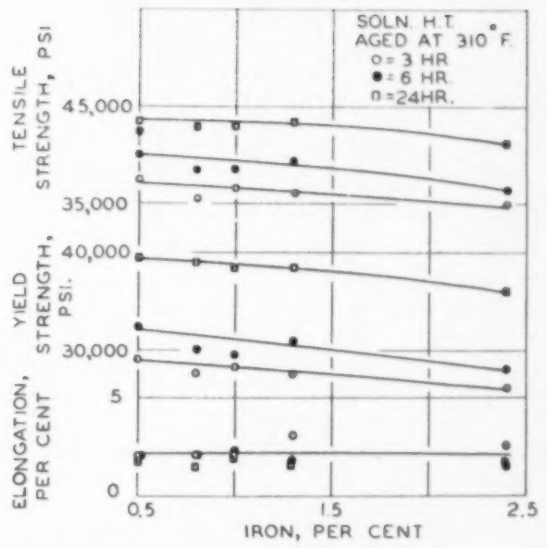


Fig. 21—Effect of iron on the properties of test bars solution treated and aged at 310° F.

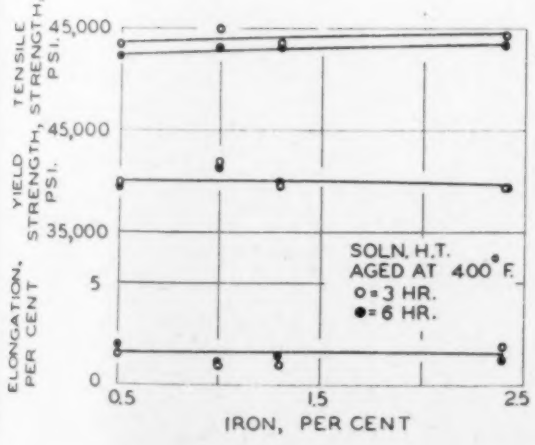


Fig. 22—Effect of iron upon solution treating and aging at 400° F.

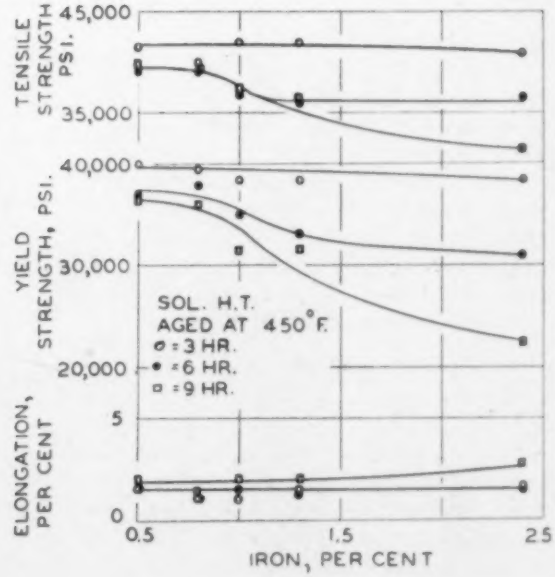


Fig. 23—Effect of iron upon solution treating and aging at 450° F.

Table 7

ALLOY COMPOSITION

Component	*Per Cent*
Copper	1.15
Silicon	5.04
Magnesium	0.52
Manganese	0.58
Iron	0.76
Nickel	0.02
Zinc	0.03
Chromium	0.20
Titanium	0.08
Aluminum	Balance

Table 8

ALLOY COMPOSITION

Component	*Per Cent*
Copper	1.33
Silicon	5.17
Magnesium	0.33
Manganese	0.03
Iron	0.77
Nickel	0.02
Zinc	0.23
Chromium	0.10
Titanium	0.10
Aluminum	Balance

Table 9

ALLOY COMPOSITION

Component	*Per Cent*
Copper	1.68
Silicon	4.90
Magnesium	0.53
Manganese	0.37
Iron	0.48
Nickel	0.03
Zinc	0.25
Chromium	0.006
Titanium	0.005
Aluminum	Balance

the alloy, although the significance of the slight changes in slope of these curves is dubious.

According to the plot in Fig. 20, iron of over one per cent appears to impair the strength properties of F150.5 alloy in the as-cast condition. However, the data used in this plot were obtained from one to 3 days after casting. Spare tensile bars of the alloy containing 2.4 per cent iron were pulled after aging 18 days at a room temperature of 75-80° F.

Mechanical Properties

Tensile strength averaged 26,100 psi., yield strength 17,700 psi., elongation 1½ per cent, and the Brinell hardness was 72. Since these properties compared favorably with those obtained on the initial alloy, it seemed advisable to check the room temperature aging characteristics of the intervening alloys to note any change in hardness which would indicate a substantial change in tensile properties.

Sections of test bars which had been retained for metallographic study were utilized for hardness measurements. The Brinell hardnesses (500-kg. load, 10-mm. ball, 30 sec.) shown in Table 10 were obtained.

It would appear that no subsequent hardening occurred in the alloy containing one per cent of iron or less. The higher iron alloys definitely hardened during the period of observation.

Final mechanical properties of these alloys in the as-cast temper are very close to those containing low iron, and indicate that it has no effect upon the alloy in this temper except to inhibit the precipitation of a hardening constituent under the casting and cooling conditions employed.

The only change in microstructure occurred in the high iron alloy. Figure 24 shows plates of primary alpha (FeSi) found in the alloy with 2.4 per cent of iron.



Fig. 24—As-quenched alloy with 2.4 per cent iron. Primary alpha Fe-Si (plates), silicon (half tone). Etchant, 1 per cent HF. 300X.

Effect of Manganese

Absence of a deleterious effect of iron on this alloy is believed to be due at least in part to the presence of manganese. This fact has been noted previously in the literature[7,8,9] although no systematic study to substantiate this claim has been forthcoming. Therefore, it was decided to attempt to prove this fact by determining the relative effects of manganese and iron in various combinations with each other.

Relative Effects of Iron and Manganese. The possibility that manganese may play an important role in the development of superior mechanical properties in F150.5 alloy led to the following investigation. A quantity of ingots of the analysis shown in Table 11 was procured.

Several ingots were melted and the mechanical properties in various tempers obtained. This metal was then remelted and manganese added to raise this element to 0.50 per cent, and properties redetermined. The metal was melted a third time and iron was added to increase it to 0.93 per cent, and again physical properties were determined.

These three melts, then, were used to obtain the properties of an F150.5 alloy containing low iron (0.34 per cent) and low manganese (0.01 per cent); low iron (0.34 per cent) and high manganese (0.50 per cent); and high iron (0.93 per cent) and high manganese (0.50 per cent) in that order.

Another set of ingots from the initial charge was melted and iron added first to raise it to 0.83 per cent. After the usual physical tests, the metal was remelted and manganese added, which raised this element to 0.65 per cent.

Second Melt

The second pair of melts indicated the properties to be obtained on the same F150.5 alloy when it contained high iron (0.83 per cent) with low manganese (0.01 per cent) and again high iron (0.83 per cent)

Table 10

ALUMINUM ALLOY HARDNESSES

Iron Content, per cent	*Aged at Room Temp., days*	*Brinell Hardness No.*
0.48	1-4	74-74
	70	74
0.80	1-4	72-74
	60	74
1.01	1-4	74-74
	40	74
1.32	1-4	65-65
	30	74
2.40	1-4	61-61
	18	72-74-72

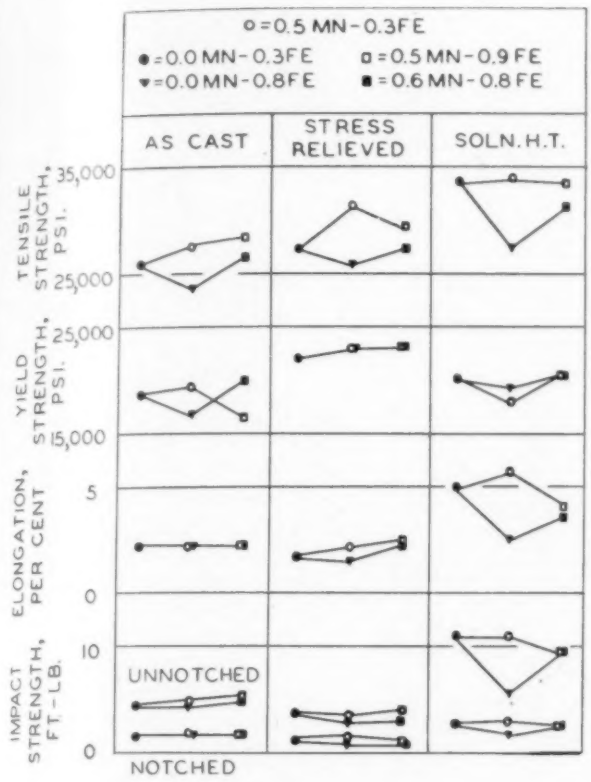


Fig. 25—Effect of various manganese-iron combinations on properties in the as-cast, stress relieved, and solution heat treated tempers.

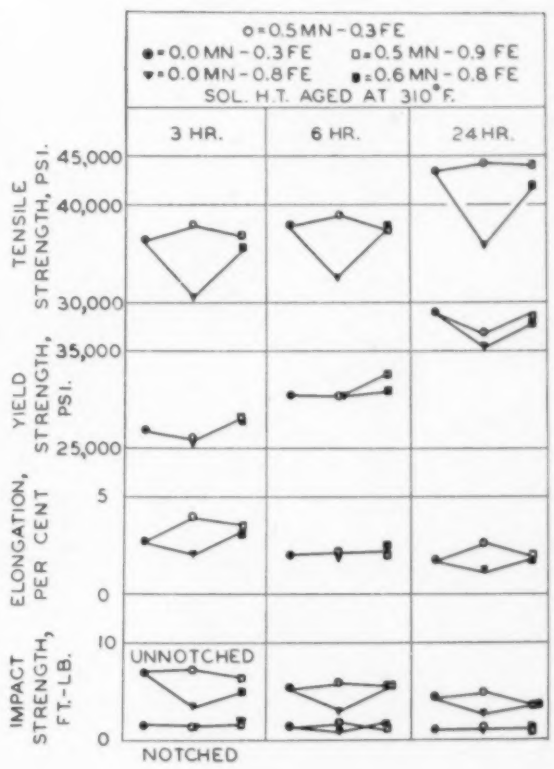


Fig. 26—Effect of various manganese-iron combinations on properties of specimens solution heat treated and aged at 310° F.

with high manganese (0.65 per cent).

The results of all five heats are plotted in Figs. 25, 26 and 27. It is readily apparent that the only alloy having inferior properties is the one with high iron and no manganese. When manganese is added to this alloy a distinct improvement in properties is noted which makes them generally equivalent to those of the initial alloy with low iron and low manganese.

Arriving at substantially the same set of properties by this circuitous method is considered an excellent check on the fact that iron is a much less critical element than manganese in this alloy. Iron under one per cent has essentially no effect if about 0.5 per cent manganese is also present. In fact, manganese appears to effect a slight but persistant improvement in even the low iron alloy.

Further, these data indicate that 0.5 per cent or 0.6 per cent iron may well be the maximum permissible if low manganese is also re-

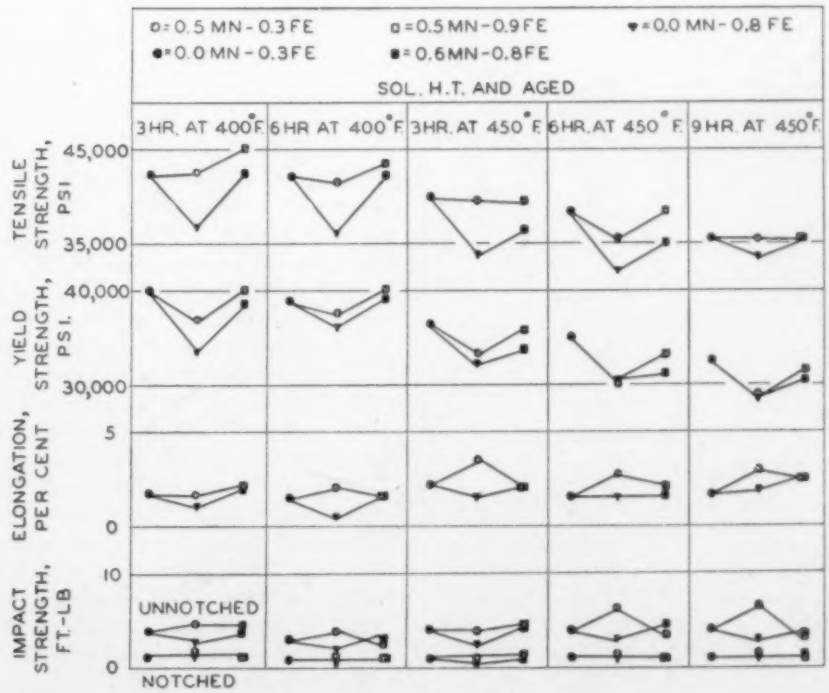


Fig. 27—Effect of various manganese-iron combinations upon solution treating and aging at 400° F. and 450° F.

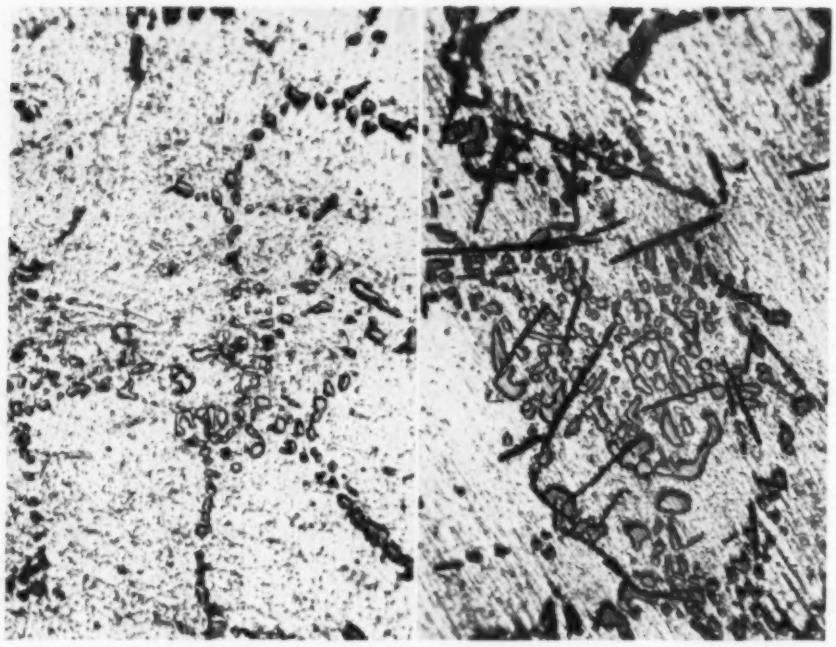


Fig. 28—Effect of iron on microstructure (as-quenched). Etchant, 1 per cent HF. 300X. Left—Initial alloy with 0.34 per cent iron. Spheroidized silicon in alpha aluminum matrix. Right—Same alloy with 0.83 per cent iron. Beta Fe-Si (needles) associated with spheroidized silicon.

quired. In this regard the present specifications appear justified, since with more than 0.5 per cent iron the beta (Fe-Si) phase field is entered with its attendant deleterious effect on mechanical properties.

Table 11

ALLOY COMPOSITION

Component	*Per Cent*
Copper	1.35
Silicon	5.52
Magnesium	0.55
Manganese	0.01
Iron	0.34
Nickel	0.01
Zinc	0.04
Chromium	0.10
Titanium	0.05
Aluminum	Balance

Table 12

ALLOY COMPOSITIONS

Component	*Initial Alloy, per cent*	*Adjusted Alloy, per cent*
Copper	1.35	1.35
Silicon	5.52	5.52
Magnesium	0.55	0.55
Manganese	0.01	0.50
Iron	0.34	0.81
Nickel	0.01	0.24
Zinc	0.04	0.37
Chromium	0.10	0.10
Titanium	0.05	0.05
Aluminum	Balance	Balance

A distinct change in microstructure accompanies the observed variation in properties. Figure 28 shows both the initial microstructure and the needlelike phase which occurred in the F150.5 alloy containing 0.83 per cent iron and 0.01 per cent manganese. No such structure could be located in the initial alloy, although an Al-Fe-Si compound must exist[10]. It is apparently too finely dispersed to be readily detected microscopically.

When 0.65 per cent manganese is added to the high iron alloy, the needlelike beta (Fe-Si) phase is suppressed and the common scriptlike habit of secondary alpha (Fe-Si) is substituted (Fig. 29). This same constituent is found in the low iron alloy containing 0.5 manganese. Heat treatment does not appear to affect the form or occurrence of these phases, and they show little tendency to go into solid solution.

Composite Effect of Greater Percentage of Minor Alloying Elements in F150.5 Alloy. Upon completion of the various investigations previously described it became apparent that an equally satisfactory specification for this alloy could be utilized that was more generous in regard to the investigated elements. The resulting alloy would have equivalent properties to one of high commercial purity. A series of tests was performed with alloys of the compositions shown in Table 12, produced from the same lot of metal and adding desired quantities of iron, manganese, nickel, and zinc to one of the melts.

Results of this test are plotted in Fig. 30. It is apparent that there is essentially no difference between these alloys in regard to mechanical properties except for the fact that the more highly alloyed material overages more rapidly at a temperature of 450° F.

As may be noted from the curves previously given in this paper, this phenomenon of accelerated overaging appears to be related to the total quantity of alloying elements rather than being an effect ascribable to any one element. All F150.5 alloys overage at temperatures of 450° F. independent of the degree of purity, but an extra per cent or two of other alloying elements increases this overaging.

It should be emphasized that the tendency for the more highly alloyed F150.5 alloys to overage at the temperatures of 450° F. after solution heat treatment is not necessarily a disadvantage.

Heating at these relatively high



Fig. 29—Structure of alloy shown in Fig. 28 (right) when alloyed with 0.65 per cent manganese. Secondary alpha Fe-Si (script) and spheroidized silicon. Etchant, 1 per cent HF. 300X.

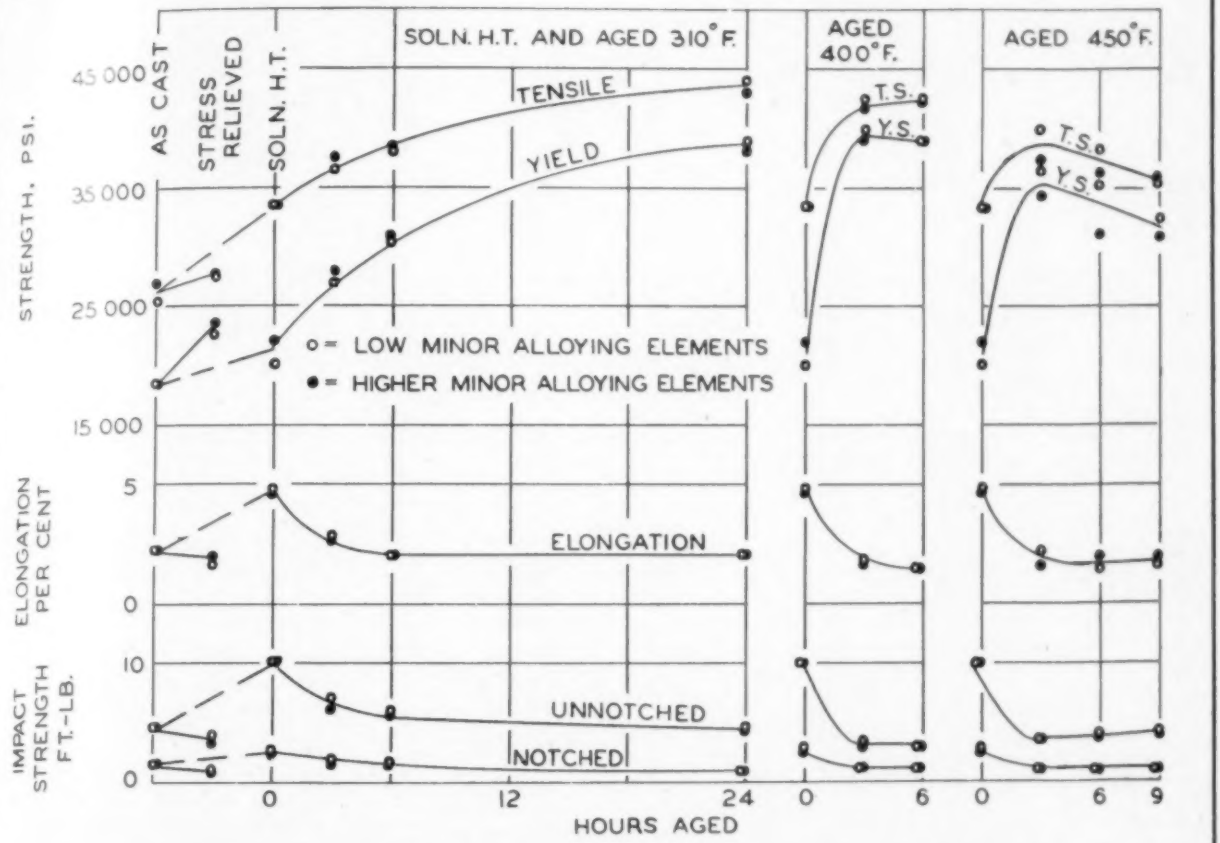


Fig. 30—Composite effect of higher percentages of minor alloying elements on an F150.5 alloy of high commercial purity.

temperatures is practiced to stabilize the alloy so that further dimensional changes at intermediate temperatures will not occur, and to relieve residual internal stresses due to quenching. Alloys which overage more rapidly may, therefore, be stabilized in shorter heat treating periods.

Summary

The effects of casting conditions and of varying amounts of magnesium, manganese, iron, nickel, and zinc have been determined on the sand cast and heat treated properties of representative commercial samples within the range of F150.5 aluminum alloy.

Pouring temperatures greatly affect the resulting grain size and physical properties. Overheating the alloy has no influence on the resultant grain size or mechanical properties providing excessive gassing has not occurred and the alloy is cooled to the optimum casting temperature. In general, the alloy should be cast as cold as possible: 1250° F. maximum temperature for standard test bars.

Magnesium is the most critical element in this alloy in its effect on mechanical properties. It tends to oxidize selectively upon melting and remelting, and the loss must be made up. It renders the alloy susceptible to heat treatment, and increasing amounts produce stronger, less ductile castings.

Nickel under 0.30 per cent has essentially no effect on the mechanical properties investigated. Zinc is likewise harmless in amounts up to one per cent. Iron may or may not be deleterious, depending upon the manganese content of the alloy.

Probably no more than 0.6 per cent iron can be safely tolerated if manganese is absent. If about 0.5 per cent manganese is also present, up to one per cent of iron will have no effect. Since manganese also improves the properties of the alloy with low iron, it appears desirable to produce this alloy with at least 0.3 per cent manganese.

Since iron pickup is a constant foundry problem, such insurance against inferior properties due to comparatively high iron should be of interest to the producer and consumer of castings alike. Probably manganese should be specified as a range of 0.30 to 0.70 per cent, while the iron content in castings should then be limited to a maximum of one per cent.

Acknowledgment

The work reported in this paper represents part of a program of research conducted at the Central Research Laboratory of the American Smelting and Refining Co. for its Federated Metals Div.

The guidance and interest of Dr. A. J. Phillips and A. A. Smith, Jr., is gratefully acknowledged. The author is indebted to Miss Doris Marshall for her assistance in carrying out the laboratory work.

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DISCUSSION

Chairman: WALTER BONSACK, National Smelting Co., Cleveland.

Co-chairman: L. W. EASTWOOD, Battelle Memorial Institute, Columbus, Ohio.

WALTER KLAYER[1]: Did I understand you to say that you heated the molten metal to 1700° F., and then poured the test bars at the various temperatures shown in Fig. 3?

MR. QUADT: That is right.

MR. KLAYER: I have data that would show that the condition would be different had you not gone so high initially.

MR. QUADT: You mean melting at a lower temperature?

MR. KLAYER: Yes, melting at 1400° to 1450° F. maximum.

MR. QUADT: What effect does that have?

MR. KLAYER: It gives finer grain size and better physical properties.

MR. QUADT: Data shown in Fig. 3 indicates a final grain size of approximately one-half mm., which is considered reasonably fine when the alloy is cooled to and poured at 1180° to 1200° F. Tensile strength in the as-cast condition was approximately 28,000 psi. which is superior for this alloy in the as-cast condition. If the tensile property can be improved by melting the alloy at lower temperatures, I don't think the improvement would be substantial.

MEMBER: Did you notice appreciably greater gas absorption in the alloy when heated to 1700° F.?

MR. QUADT: That is the big objection to overheating aluminum. We used coke-fired furnaces. The quality of hydrogen compounds normally present in the furnace atmosphere could be considered at a minimum. We polished sections of the alloys to determine whether or not there were excessive amounts of pin-hole porosity present. However, previous work which we have done on alloys of this and other types has indicated that small quantities of gas porosity have no effect on tensile properties. In this particular case, heating to 1700° F., under those conditions, no increase in pin-hole porosity was noted.

MEMBER: Is it not possible, then, that the method of heating effected the high physicals? Had the melting been done with other types of fuel, perhaps you would have had greater gas porosity and a reduction in physicals.

MR. QUADT: There is no question about that. Heating aluminum to 1700° F. is definitely undesirable with other types of fuels. In most normal melting equipment, where gas, oil, or even coke are used that should never be done. The intent of the test described was just to determine what the effect on mechanical properties would be if inadvertently the temperature approached 1700° F. in melting. The result of this and several other tests indicated that it had no effect providing the metal was not excessively gassed.

[1]Aluminum Industries, Inc., Cincinnati, Ohio.

HOODS AND PIPING FOR FOUNDRY DUST CONTROL SYSTEMS

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The Kirk & Blum Manufacturing Co.
Cincinnati, Ohio

THE first step in planning a well engineered foundry dust control system is selection and fitting of a proper and correctly designed hood to the source of dust. The source of dust is located and the hood is placed at this point, to reduce the dust concentration to that within safe limits of hygienic standards.

The second step deals with air volume, which is calculated to produce the desired result. This volume determines pipe sizes, actual hood openings and selection of exhaust equipment to produce the desired air flow.

The third step involves details of construction, such as hood and piping gauge, proper clearances, and the safety, comfort and visibility of operator. Adherence to American Foundrymen's Association and State Codes is checked and finally the detailed plan of dust control system is ready for preparation.

Hood for Shakeout

Since the crane operator requires maximum of visibility for unhampered handling of flasks, the use of an overhead hood is sometimes precluded. For this reason the "side" or "cross draft" hood is usually considered (Fig. 1). The cross draft hood does not have a roof or full extension over shakeout grate and, therefore, it is natural that a much greater volume of air must be handled, resulting in excessive power costs and heat losses.

It is well known that velocity contours of controlling air decrease quite rapidly as the distance from hood intake increases. This means that the foundry operator must increase power consumption approximately 100 per cent to operate the cross draft hood, in comparison to an overhead type or enclosure hood.

The answer to this problem could be a compromise between the straight "cross draft" and totally enclosed hood, as illustrated in Fig. 2.

Perhaps this compromise could be accomplished by means of a pneumatic finger canopy arrangement, having rather closely spaced round canvas "fingers" possibly 8 or 10 in. in diameter. They would be held in a horizontal position over the grate by means of a small supply blower mounted alongside the shakeout hood or on top. The space between tubes or fingers would afford visibility to the crane operator and the chains would slide and work their way between the tubes, with little or no difficulty.

Such a design would save 50 to 75 per cent of power costs, in comparison to straight shakeout cross draft hood. This is a typical example of the specialized design possible in foundry hoods, which is "step one" in dust control.

Selection and fitting of a properly and correctly designed hood to the dust source is the first step in planning a well engineered foundry dust control system. The required air volume determines pipe sizes, actual hood openings and selection of exhaust equipment to produce the desired air flow for effective dust removal. Hood and piping gauge, proper clearances, and the safety, comfort and visibility of operator must be considered in designing the dust control system.

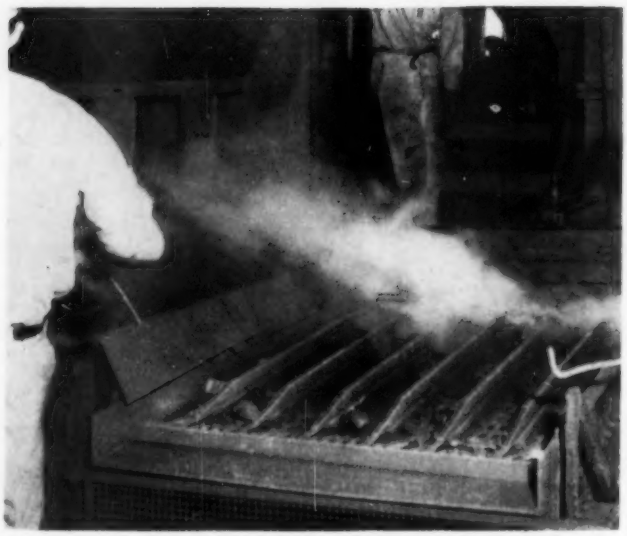
Snagging and Cleaning

Snag and cleaning departments in a foundry offer another opportunity to improve hood design (Fig. 3). Here, it is possible to apply pedal operated turntables, built to a convenient height of operator, so that smaller castings are turned and operator grinds into a fixed hood position. Of course, if the castings are larger, the cylindrical type downdraft table arrangement is preferred (Fig. 4).

If floor gratings are used, movable vertical screens or baffles are of assistance in directing the light grinding dust downward. The conveyor transfer points, elevator boot and head connections, shaker screens, mixers, mills, aerators, tailings hoppers, tunnels and the like can be usually handled with direct rectangular to round and overhead hooded connections.

Grinding Operations. Individual grinding stands or groups of wheels are served by heavy duty plate hoods,

Presented at a Session sponsored by the Safety and Hygiene and the Plant and Plant Equipment Committees at the Fiftieth Annual Meeting of the American Foundrymen's Association at Cleveland, May 6, 1946.



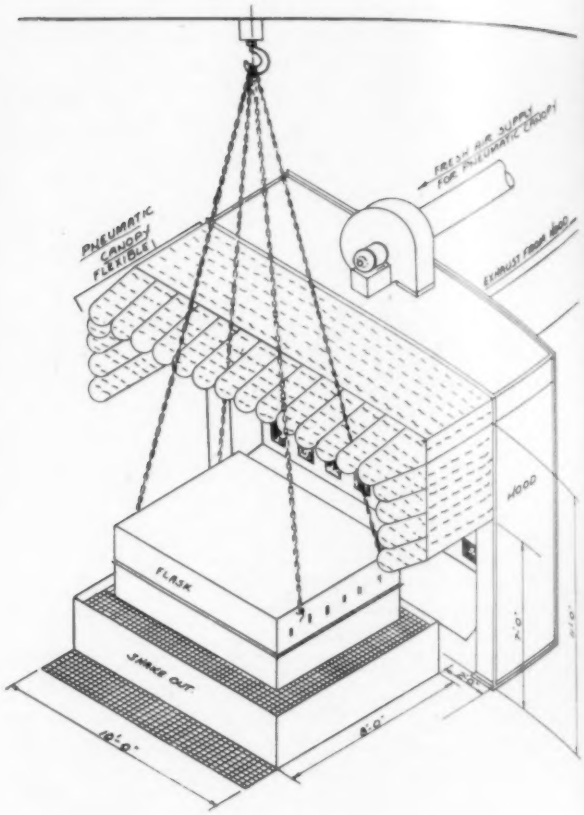
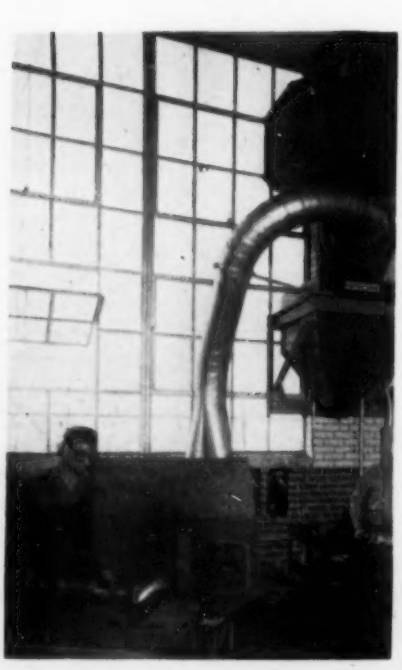


Fig. 1 (Top)—Side or cross draft hood in connection with the mold shakeout operation. Fig. 2 (Right)—Pneumatic finger canopy type of hood, a compromise between the straight cross draft and the totally enclosed hood (Patented).

some with traps built immediately below intake while others have traps in the vertical exhaust to the rear of stands (Fig. 5). Foot rests sometimes are incorporated into design, and sides and tops are hinged for wheel replacement or dressing.

Fig. 3—Pedal operated turntable in the foundry snagging and cleaning department. Note cross draft hood arrangement.



A large individual grinding wheel, either vertical or horizontal always proves an interesting problem. It is advisable to consider wheel replacement and locating the point of exhaust as close to a perfect tangent to material leaving the periphery of wheel, as possible. Sometimes on larger wheels the complete top half of hood is removable for wheel replacement.

The safety of these types of hoods, however, should be the paramount consideration. Accidents have been recorded, where the face opening of hood was too great and castings would become lodged between side of hood and wheel itself.

The horizontal grinder frequently has its own exhaust connections built in bottom. It is then only necessary to utilize a circular header at the bottom of grinder to collect the various points of exhaust.

Figure 6 illustrates a wet grinding application utilizing heavy duty steel plate hoods, with provision for ready removal of the hoods and wheels, by means of an overhead chain fall.

Mold Cooling Tunnels. Special consideration is given to the design of mold cooling tunnels (Fig. 7). These hoods are built from the floor, having a semi-circular top and served by one or two vertical exhaust stacks with fans built into stacks themselves. Removable sections are provided for applying and removing weights, if desired.

Condensation Traps. Condensation traps are sometimes used in the branch lines for the purpose of reducing moisture content which is always undesirable where accompanied by solids and dust particles in the air stream. Heavy black steel piping with electric welded seams and joined by compansion angle connections is used where temperatures

Fig. 4—Cylindrical type down draft table arrangement for snagging and cleaning castings.





Fig. 5—Dust removing arrangement for individual grinding stands and groups of grinding wheels.



Fig. 6—Dust removing hoods on a wet grinding application.

exceed 400° F. Unusual structural conditions are encountered as illustrated by cross draft fume exhaust system (Fig. 8).

Cleanouts are usually placed at 10 ft. centers along side of header or main pipe for cleaning, inspection and maintenance. Discharge stacks extending vertically through roof may be equipped either with rain cap, 90° elbow, or tapered static regain section.

The most efficient stack termination is, of course, the slightly tapered regain nozzle which converts some of the velocity pressure back into static. The rain cap and elbow offer some protection against weather but incur a slight additional back pressure to air discharge.

Pattern Shop

Pattern shop has its dust control problem. However, here we have woodworking shavings, chips and sawdust rather than foundry abrasives. The hoods and piping can be somewhat lighter in gauge, but nevertheless, require careful planning and expert fitting to the woodworking machines (Fig. 9).

Air Volume and Piping Size

In most instances, in foundry proper, space being at a premium, the hood equipment and piping from the hoods must be built into extremely limited space. This leads us into steps two and three, or the important phase of determining air volume and piping size.

Not until relatively recent years has the problem of pipe design been given its just place of importance in modern foundry dust control systems. We keep rediscovering periodically that piping plays a major role in good foundry practice and although the design and ideas are somewhat different and slightly more advanced in the last quarter



Fig. 7—Mold cooling tunnels.

century, the basic fundamentals are the same.

Fittings and Connections. Fittings and connections are used with branches brought in at very gradual angles and elbows are provided with throat radii resulting in a minimum of friction loss to air and material flow. Since the maximum amount of wear in an elbow takes place at the heel, opposite side from throat, removable plates are provided here which receive the wear of the abrasive materials handled in air stream, and are replaceable making it unnecessary to change elbow itself.

The heavy duty "one piece" elbow is usually seen in modern installations. This elbow, by virtue of being formed from one piece of sheet metal, has but one throat seam and no rivets to wear. The girth seams are formed from the metal itself providing three thicknesses of metal seam completely around the elbow at each crimp.

The majority of foundry piping installations make use of galvanized iron or steel, riveted and soldered longitudinal and circumferential seams.

Gauges of piping range from No. 18 gauge for small diameter pipes through No. 14 gauge for larger sizes. Elbows are at least two gauges heavier. Often old dust control systems may be modernized by rearranging, balancing and replacing piping, resulting in increased efficiency.

Piping Diameter

Its determination is a study in itself. Present day foundry piping design is customarily based on ve-

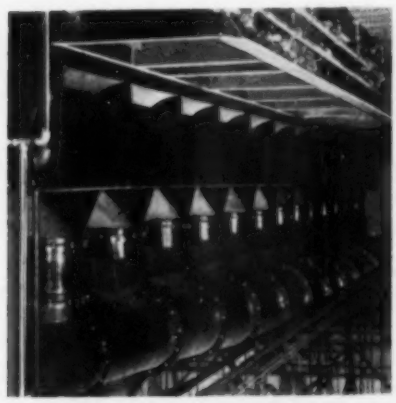


Fig. 8 (Above)—Cross draft fume exhaust system in connection with a condensation trap.



Fig. 9—Dust control system in a wood pattern shop.

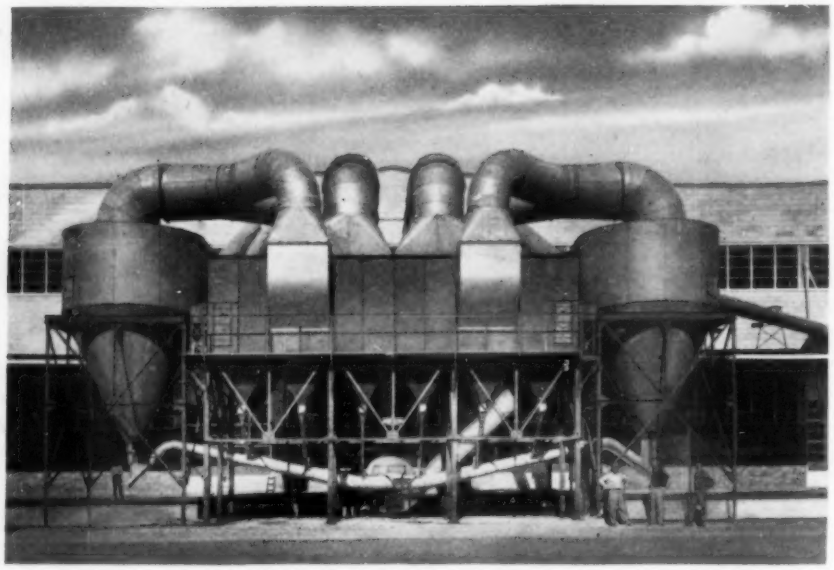


Fig. 10 (Left)—Termination of dust control piping in a large combination collection unit.

locities ranging from 3600 to 4500 fpm. or velocity pressures from 0.8 through 1.3 in. of H_2O. It is desirable to maintain as constant a velocity pressure as possible and sometimes cut-offs are provided and locked in a fixed position to accomplish this means.

The computation of losses in the entire piping system must be carefully worked out, and a study of hood entrance losses with equivalent loss in percentage of velocity pressure, plus lineal friction loss of pipe in static, branch and elbow losses, plus pressure drop across precleaning and exhausting equipment determine the total pressure at which the systems would operate.

Round piping is usually preferred to oval or elliptical and since the static pressures are quite high, in material handling systems, rectangular and square ducts, if used must necessarily have a good deal of bracing tc prevent vibration. Finally, ing of pipe as the smaller sizes used caution should be exerted in the sizwill incur considerable more power consumption if selected for velocities above 4500 fpm.

Figure 10 shows the termination of dust control piping in a large combination collection unit, the size of which may be approximated by the men in lower right hand corner.

USE OF WOOD MODELS IN DESIGNING AND PLANNING

G. W. Schuller
Supervisor, Metal Pattern Making
Caterpillar Tractor Co.
Peoria, Illinois

APPLICATION of wood models in industry is rapidly becoming a device to facilitate betterment of planning and design and an aid to quality and economy of a product.

Most of us in the various phases of skilled crafts and trades and in the fields of designing and planning must familiarize ourselves with the use of sketches, drawings and blueprints, to the end that the design or object on the drawing will be completely visualized.

We can readily see from the drawing that the particular object pictured has a length and width. The various adjacent views, sectional views and notes project onto the screen of the mind the necessary thickness. This property of visualizing the thickness or third dimensional view of an object from the drawing is a vital factor. Few people have completely mastered this valuable asset without the injection of still another factor, namely, models.

Aids Visualization

Many errors are made in design due to the lack of visual power. Have you ever heard a carpenter say that he would have moved that particular door or closet or have completely revamped the room arrangement were he to rebuild his home? Maybe this has happened in your own home even though you have poured through every detail on the drawing with the building contractor.

Many pieces of new pattern equipment for castings involving considerable coring are correct in every detail on the drawing and expertly planned as to clearances for the various cores and placement of chaplet seats, etc.; however, that same pattern equipment may go to the foundry for a casting only to find that the cores could not be assembled in the mold. Or the core assembly in the cope mold would not pass the cores set in the drag.

Even though the pattern equipment is made as per drawing and the house built as per specification, these errors, due to the lack of visual power or the inability to foresee the processing of operations in every detail, are costly to correct. In many instances, inconvenient and unfavorable circumstances are set up in trying to apply other methods to avoid the cost of redesign or reworking.

Aid in Core Assembly

The use of wood models have been employed many times in intricate and complicated core assemblies. The pattern maker has a better opportunity to see and simplify the partings of the core boxes and elimination of loose pieces. Many combinations of cores previously planned from the drawing have been combined into one core by the making of a wood model.

The foundryman also may be consulted and enlightened on the problems he may encounter. This method affords him an opportunity to see the contours of the cores easily and to voice his approval or suggest changes in assembly before any costly equipment is made. He can readily devise his system of installing the necessary venting arrangements, the placement of wires or rods or the complete method to make and assemble the cores. Many changes come about in this manner which may simplify core making, gating and molding procedures and lead to better or more economical casting.

Short Cut to Planning

The job of careful planning and designing can never be over done whether it be for a building, a set-up of operations in machining or the complete layout for the function of an industry. The use of models is a short cut to all of these factors.

Figure 1 shows a model of a complete line for the making of dry sand molds for an aluminum air-cooled cylinder head. Every detail was carefully planned and designed on paper—then a wooden model was made so the most minute detail and operation could be observed. This model was made to the scale of one-half inch to the foot. The cylinder head involved was the deep

Many errors are made in design due to the lack of visual power. The job of careful planning and designing can never be over done whether it be for a building, a set-up of operations in machining or the complete layout for the function of an industry. The use of models is truly a short cut to all of these factors. Models, preferably wood models, have a definite place in the scheme of industry and construction.

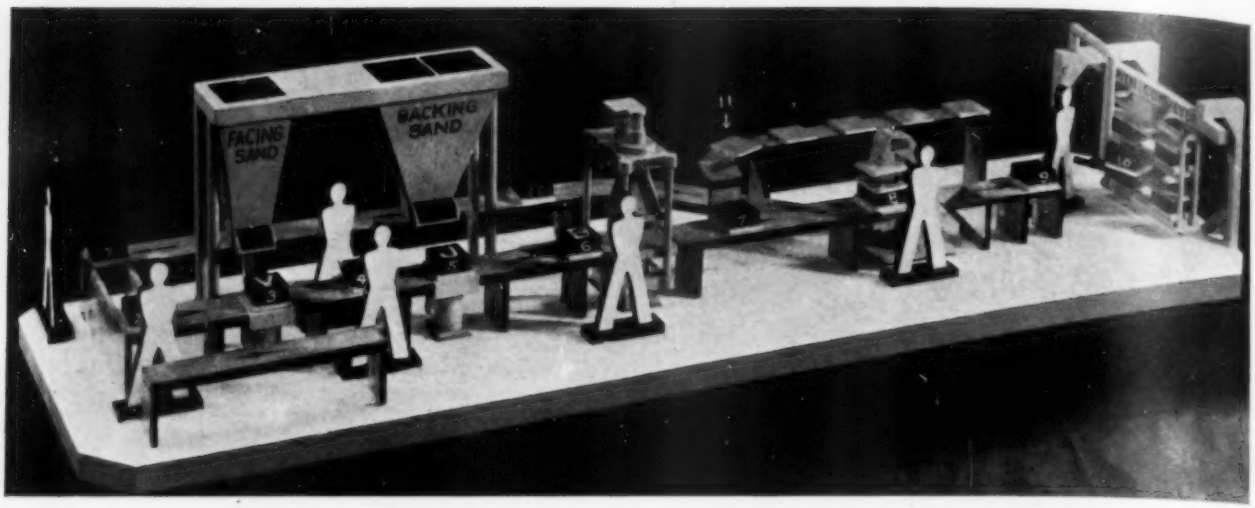


Fig. 1—Model of core making line. 1—Cleaning and washing pattern; 2—Inserting nails at pattern joint; 3—Vibrating sand into fins; 4—Adding a few additional nails; 5—Jolting backing sand over pattern; 6—Core entering blower for completion; 7—Placing drier on top of finished core; 8—Core in process of rolling over; 9—Core completed;; 10—Placing core on continuous core oven; 11—Driers returning from oven; 12—Pattern return to station No. 1.

finned type for a radial engine.

Considerable experimental work was involved in a search for a more rapid and economical method of producing these castings and the use of this model facilitated in the establishment of a well planned and progressive method of producing the molds. The "assembly line" method of operation in making the molds was used, as shown in Fig. 1.

Several sets of patterns, in core box form, were employed and progressed to stations for each operation. Every sequence of operation was worked out by the use of this model from the washing and cleaning of the pattern at the first station to the final placement of the core on the oven conveyor and the eventual return of the pattern to the starting point.

Models in Aircraft Industry

The aircraft and automotive industries are great users of models, both full scale and miniature size. Model making is a vital part of the aircraft industry and thousands of dollars are spent on a single model of many an aircraft design so the stresses on the wings and fuselage can be studied by use of the wind tunnel. Many corrections in design are made through the careful research on the model before any attempt is made to build the plane.

The contour of body design in the automotive industry is worked out by the use of full scale models. The answer to the prayer of an automotive manufacturer is a compact unit loaded with power and speed yet pleasing and alluring to the eye. In normal times, a large part of the battle of sales is fought by the careful shaping of the body models. Imagine the thousands of dollars involved in making dies for forming the body parts that can be jeopardized by a misinterpretation on the drawings. Wood models completely eliminate this hazard.

In planning the sequence of machining operations for the numerous parts comprising a motor, a pilot piece is usually machined first to "iron out" those hidden "bugs" that cannot quite be visualized on the drawings. To the planner and machinist these pilot pieces are actually his models. Many of you probably can say much about time, material and tooling that was saved or possibly even eliminated in your own plant by the use of this method.

Making of Models

You may ask these questions, "What materials are used for models?" and "Who makes them?" Several materials are used, such as wood, plaster, white metal and various other metals. Wood is generally used because of its workability and economy. Also changes can readily be made in models made of wood.

Plaster models are used when a precise shape or contour is to be held. Although they are easily broken or marred, they are often used by the engineer for the design of cylinder head inlet and exhaust ports. They are oftentimes used for the measurement of volumes of these ports and the shaping of the combustion chambers, to effect the desired volume in gas engine cylinder heads. White metal models are used when accuracy is expected to be maintained and much handling is required.

The question then arises—"Who makes the models?" The engineer, the tool designer, the machinist or foundryman? It should fall on someone with respect or partial knowledge of each of the above, or a patternmaker. The patternmaker works with the necessary materials and has a knowledge of machinery, tooling, and is well acquainted with foundry problems. He is also an engineer in his own right.

It is commonly known that models, preferably wood models, do have a definite place in the scheme of industry and construction. A model generally speaking, is actually a device for viewing the "scenes behind the scenes." Their uses are varied and many, which can be summarized as follows:

The engineer may better see his

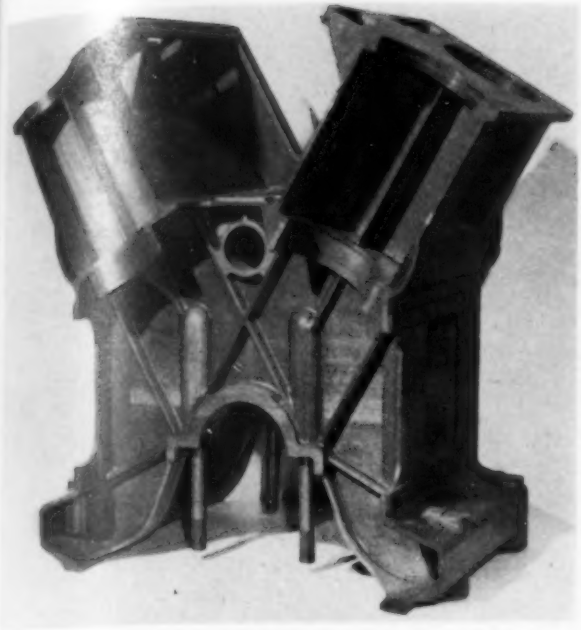




Fig. 2 (above)—Full scale model for V-type 12 cylinder Diesel engine block. Fig. 3 (right)—Full scale partial section through water jacket section of V-12 engine block.

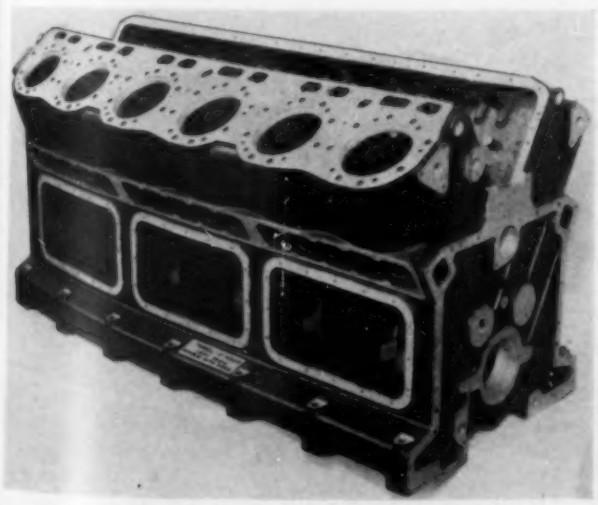
design in actual form and readily determine any weakness of structure and adjust any section to be under stress or poorly designed. A full scale section of a model "V" type, 12-cylinder Diesel engine block is shown in Fig. 2. This model and the section through the bore model, Fig. 3, were built principally for the study of design.

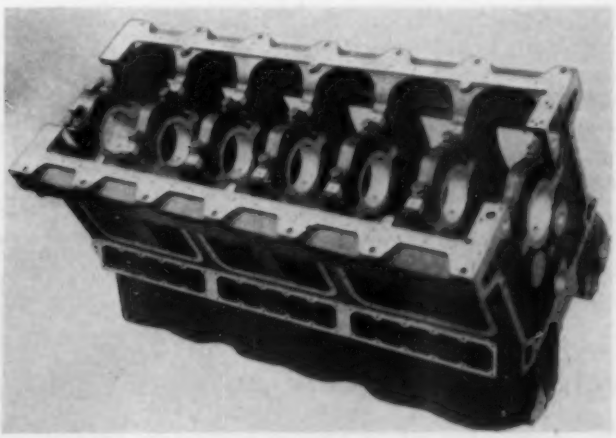
You may note the change of design in the ribbing and thin flange at the tappet chamber (shown at the tops of the "V") in the original model, Fig. 2, and the final model, Fig. 4. Also note the single strengthening rib in the water jacket of each bore in Fig. 2 to the type shown in Fig. 3 and partially visible in Fig. 4.

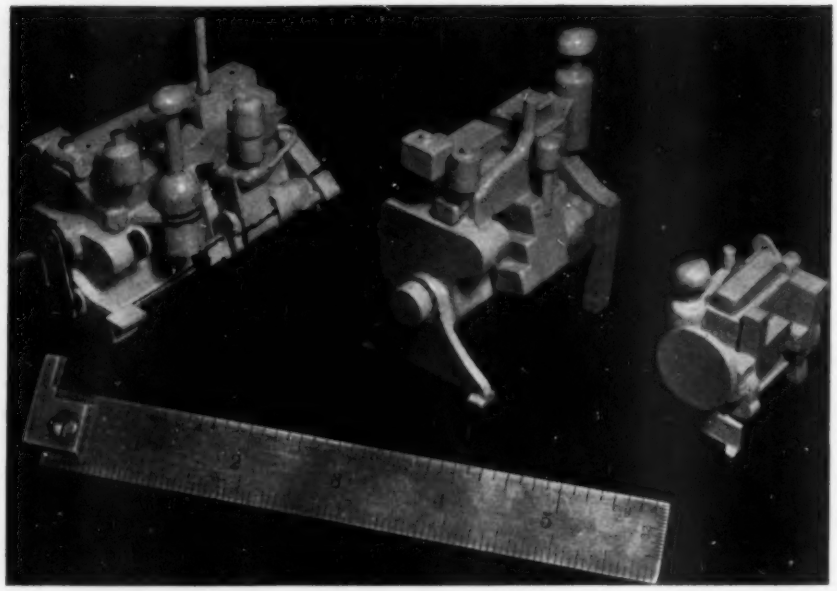
The final design incorporated a system of ribs to be directly under the studs that hold the cylinder head on the block. You can also note the change of the water opening passage on the side of the model Fig. 4 from the original model. These are but a few of the changes in design which were effected by the engineer after making a complete study of the models.

For the Planning and Tool Design Departments, the same "V-12" model was constructed for the processing of machine operations and the use of the tool designer, as shown on Fig. 4 and 5. The application of this half scale model to each of the

Fig. 4 (left)—Top view of half scale model of 12 cylinder Diesel block used by Planning and Tool Design Departments to aid in processing machine operations and design of necessary tooling. Fig. 5 (below)—Bottom view of half scale model of 12 cylinder Diesel block Shown in Fig. 4.







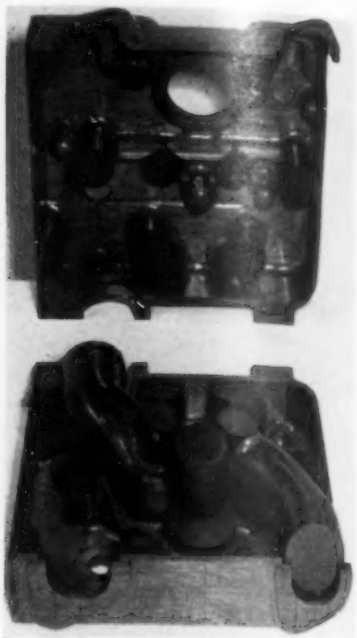


Fig. 6 (above)—Models made to one twenty-fourth scale used for assembly planning. Fig. 7 (right)—End section model of a six cylinder Diesel cylinder head used for study by engineer and pattern maker.

above departments cannot be overemphasized. The model, itself, cost approximately $2,500.00 to build and has saved many times its cost in processing and tool design alone.

It can easily be seen how far the use of this model outmodes the use of the drawing alone. The cost of the tooling in this particular block ran well over a quarter million dollars. Figures 4 and 5 have dealt with a half scale model of a large engine block. Let us look at still another phase of planning as shown by Fig. 6. These small engine models are made 1/24 scale and were built for assembly planning. They cost from $65.00 to approximately $125.00.

From the study of these models, a hanger system for the engine assembly line has been devised to be used from the beginning of assembly to the completed engine. The engines are then planned to be "run in" on the same hangers, thereby saving an estimated two and a half hours "running in" time and approximately 2,500 gallons of fuel oil per month. Models have a place in Planning and Tool Design Departments.

For the patternmaker and foundryman, much can be said about models as they are both concerned with difficulties that may be encountered in the foundry, as previously mentioned in this paper. The model shown in Fig. 7 is an end section of a six-cylinder head. The finished head casting is one casting for all six cylinders. This model was built principally for engineering and experimental departments, but returned to pattern shop for use while building pattern equipment.

This particular cylinder head presented a problem in the making of the jacket cores, as the inlet and exhaust ports are on opposite sides of the head creating a very thin sectional core. Many bosses in this head were changed in contour to permit the proper venting of cores. Many parts are redesigned to eliminate excess coring and deep green sand pockets in molding by the consultation between the patternmaker, foundryman and engineer.

For the die maker and forging manufacturer, Fig. 8 shows a wooden model of a tractor track link which is a forging incorporating many different radii and blends from one radius to a larger one. The parting line and draft angles are clearly shown on this model. Hence one of the reasons for a model is to eliminate a possible misinterpretation of the drawing. A

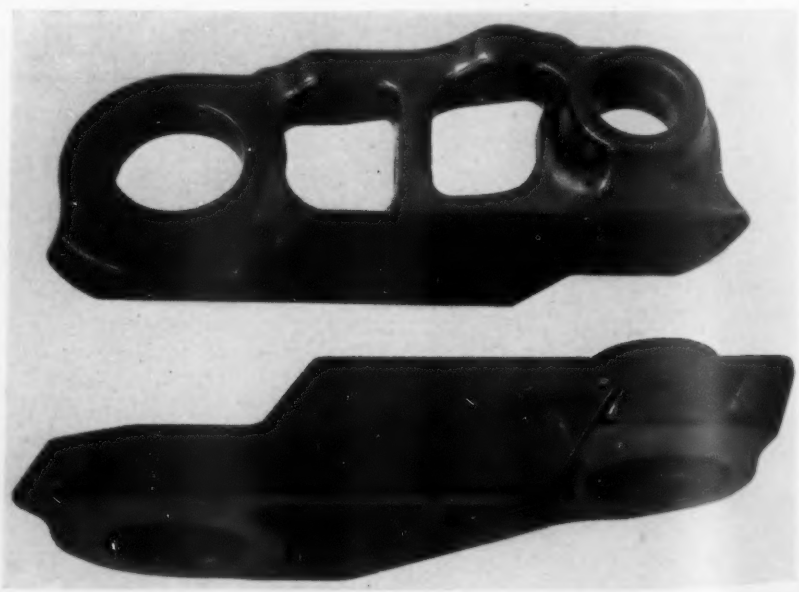


Fig. 8—Model of track link for track type tractors.

model for this type of piece is usually cast of white metal and finished to accuracy. It is then approved by the engineer and sent to the forging manufacturer along with a drawing. A duplicate white metal model is kept by the Inspection Department. This method eliminates a possible costly error in die making due to differences of blue print reading. A problem arose in connection with a diagonal brace for the underside of a tractor. This brace is a fabricated part, the bearing end piece being a forging. This part caused considerable discussion among the engineers relative to the junction of the neck of the bearing and the neck of the fabrication. The part of the bearing forging inserted in the neck of the brace was lightened out to an "X" section and plug welded by a small circle and fillet welded at the neck.

A wooden model of the neck section was decided upon. Modeling clay was laid on this same model, shown in foreground of Fig. 9, to effect a strengthening in design at this weak section. (Note the step in model to show the addition of the modeling clay.) The final product of the strengthened brace is shown by the model in background of Fig. 9. This change meant the bearing forging must also be redesigned. Fig. 10 shows a close-up of this bearing.

Modeling clay was again added to the model which would show the effect of the design to meet the requirements of the engineer and forging manufacturer. Note the parting lines and lightening out of desired forging. Through the use of a wood model and some modeling clay, a stronger brace was effected and the drawings corrected.

There are many miscellaneous applications for models. It was desired to choose from drawings, a suitable type and case letter to adopt and apply to a "Caterpillar" radiator top tank and side plate. It was necessary to design these so that they could be read from a height of 70 in. when looking at the radiator of the tractor. Three full scale wood models of the top tank were made, each with a different case letter. These models were made so that top management could quickly choose and adopt a particular type letter to be used throughout Caterpillar products.

In conclusion, the writer would like to add that the use of wood models in designing and planning does pay dividends. Although only a few illustrations are mentioned in this paper, there are countless opportunities for using such models in the industry. Each of the various phases of manufacture are rapidly becoming conscious of the problems of the other through the use of models—the engineer is becoming more interested in foundry problems—the tool designer is becoming familiar with the parting lines and draft angles on the pattern, etc.

The model tends to merge each problem into one unit so each branch of industry can actually lessen the problems of the other. The initial cost of the model is oftentimes repaid many times through the saving of time, material and energy in processing and manufacturing a product.

DISCUSSION

Chairman: FRANK C. CECH, Cleveland Trade School, Cleveland.

Co-Chairman: V. J. SEDLON, Master Pattern Co., Cleveland.

PETER RETTIG[1]: I would like to ask who constructs these models?



Fig. 10—Close-up view of model of bearing forging.

MR. SCHULLER: The patternmaker in our plant makes all the models, because he has a fair knowledge of engineering. He also has a good knowledge of tool making and tool designing. The patternmaker does the actual making of the models.

MR. RETTIG: Does the same man follow the job right through then?

MR. SCHULLER: He follows the job through so far as the model is concerned. The patternmaker makes the model from a drawing. The model may go to several stations, such as the engineering department, planning department, and possibly tool design. In many instances, changes are made before the eventual part is completed.

I brought with me a model of an oil pump that was made to simplify molding and coring and to simplify cleaning of the casting all the way around. Many changes came about in design and making of the equipment due to this model.

MR. CECH: I would like to ask what led up to the use of the models in the first place?

MR. SCHULLER: Inability to visualize a part from a drawing necessitated the making of models.

MR. CECH: Was there any particular job coming through the shop that necessitated it?

MR. SCHULLER: This particular oil pump housing had many ports that crossed each other and lead into one another, hence it was quite a job to visualize. So we started out by making the model. As the model progressed, we changed some of the ports to eliminate and simplify quite a lot of coring. We also changed the design of the part, in accordance with the engineering depart-

[1]Rettig Pattern Co., Cleveland.



Fig. 9—Models of diagonal brace in process of redesign.

ment, so we could clean the casting.

MR. CECH: I am wondering whether the engineering department themselves requested the making of the model originally. How long have you been using models?

J. E. KOLB[2]: We have been using models for years. Mostly the engineering department request models. It is easier for a designer who thoroughly understands his own drawing to impart information to one or two individuals who might be making the model. Then a group of engineers who would be working on the component parts of something such as a motor could study the model and in five minutes see all the different phases insofar as engineering, machine operations, and planning is concerned for a well-designed part.

We, in the pattern shop, can get our information, you might say, free, or in other words, the engineers pay for this model. We can see a lot of short cuts and improvements which we think would be beneficial for foundry practice that we request of the engineers as we go along with the model. The small crankcase model that was shown was requested by the planning department for machine tool purposes.

From the model, the sequence of all of their machine operations was planned. They could look at that model and locate all the drilled holes, tapped holes, milled surfaces, and bores. A group of men working in the planning department who are concerned with the processing of that part through the shop, could visualize in far less time from the model than twice the number of men could by reading the blue print only.

MEMBER: I did not notice whether Mr. Schuller stressed the importance of models to the foundry division. He seemed to be speaking about the engineers, tool shop, and machine shop. Models are of a great help to the foundry division. There are many angles that a foundryman can see from a model, that he cannot see from a drawing or get from an explanation.

At our plant we used models for the last 20 years. It has been one of our big items and big forces of help to our foundry. If Mr. Schuller could elaborate on the usefulness to the foundry, I think it would help the foundrymen.

MEMBER: Some years ago I had to make a pattern of a stirrer which interlaced with another stirrer. The engineering department did not know whether these two blades would interlock in the mixing. I proceeded to make a model, geared it so that one pattern would turn inside the other. I found that I had the thing which could not be put on blue prints and it was a great help in that case. The model was used as the pattern, hence it did not cost anything.

MR. SCHULLER: That was one time where a pattern could be used actually as a model. Many times a model is made before we have a pattern. You were fortunate in this case.

It is true models do play an important part so far as the foundry is concerned. It is a knack to have the power to complete visualization. It is really not a foundryman's job to be able to read a drawing completely. By the use of a model he can see the actual part. Many important changes came about through a foundryman's suggestion that was not seen by the patternmaker, the engineer, or anybody else.

MEMBER: The use of models in the airplane industry was spoken about. But in the ship building industry during the war we used full scale models of the interior of submarines. The company thought it worth while to spend a thousand dollars to construct the building just to house them. They were used primarily for the ship business. The ship fitter and pipe fitters, in bending copper pipes in the model avoided the use of an extra number of men working in the actual hull.

While the electricians or welders were working in the various compartments, the pipe fitters worked in the model. When the compartment was clear they fitted pipes right in and assembled them in place. They also made models of the engine room out of wood, using patternmakers and joiners for that work, and saved a great amount of labor in transporting it to the hull.

MR. SCHULLER: That is another good illustration in the use of a model.

A. H. TRAPP[3]: I have found it quite valuable in the last few years to make models of certain parts of a cylinder head. Let us consider the valve in the head where we have the intake port. It sometimes turns out to be quite complicated because they generally are made with a fork. The details are sometimes difficult to show on the drawing, and it is hard to visualize them. A series of cross sections are made on the drawing, but it is hard to visualize how the finished part will look.

Of course, it makes it rather difficult for the patternmaker to make the core box according to that lay-out. The patternmaker generally consults the foundryman or coremaker during construction of the model. Questions arise. Just how would you make that core? How will we part that box? How will it work best in your core room? These are some questions which may arise during the making of the model.

We have found it quite valuable to make up a little model of the core proper. Then we decide how to part it. For a number of years now we have been blowing cores, using core boxes for blowers. We all have been confronted with the problems of keeping core boxes tight. If you have a straight parting, you would not have much trouble. However, if you are going to have irregular partings those are generally the problems for the patternmaker, because they wear out frequently and you have to tighten them.

We have found in numerous cases that it is hard to visualize these things from a drawing, and it is hard to explain to our foundrymen just what problems we have. I am very much in favor of models. Of course, we have not gone too deeply into it, as yet, but there are numerous places where models will really help a lot, especially in determining on some complicated blower equipment.

MR. SCHULLER: There are thousands of uses for models. I have just touched on a very few illustrations. But a model actually merges all problems into one problem. The engineer is beginning to be conscious of the patternmaker: the patternmaker is conscious of the engineer, and so forth. The tool designer even is beginning to wonder how patterns are made. Models are not new, but they are being employed more readily than they have been. They pay for themselves many times over.

[2]Caterpillar Tractor Co., Peoria, Ill.

[3]Minneapolis-Moline Power Implement Co., Minneapolis.

HIGH TEMPERATURE SAND TESTING

Sixth Annual Progress Report

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Cornell University
Ithaca, N. Y.

INTRODUCTION. This, the sixth annual report of research on the behavior of steel foundry sands at elevated temperatures covers work done during the calendar year 1945. The investigations covered by this report were conducted at Cornell University by A. F. A. Research Fellow, Douglas C. Williams. The Fellowship is provided by the American Foundrymen's Association under its Technical Development Program and the work is under the general direction of the Steel Sand Subcommittee of the Committee on the Physical Properties of Foundry Sands at Elevated Temperatures.

Various miscellaneous and annual reports have set forth from time to time over the last eight years the purposes of this Research Fellowship and of the Steel Sands Subcommittee, hereinafter referred to as "6b7." In Mr. Williams' report for 1944 and in the report of the Chairman, Mr. Werner Finster, the specific aims of the work at that time were defined.

As work progressed on the "single variable" program of testing, using commercial dilatometers, it became increasingly apparent that results were not adequate for research purposes, although general trends in the effect of variable of composition and technique could be effectively established. It was felt by 6b7 that in order to obtain a fundamental understanding of the behavior of foundry sands at elevated temperatures it would be necessary to construct a special research furnace.

It developed that Cornell University was willing to provide funds for this undertaking and Mr. Williams set about to build the "ideal" furnace, which is now in operation. In order to continue simultaneously with the program of practical testing on the commercial instruments, A.F.A. provided funds for an assistant to Mr. Williams.

Also to enable the test work to be done on the latest equipment Mr. Dietert, of the Harry W. Dietert Company donated a new 1944 model dilatometer to A.F.A. for the work of 6b7. Except for certain items of general equipment such as baking ovens and miscellaneous room temperature testing equipment the Cornell laboratory is now well equipped for both fundamental research and practical testing.

It will be readily realized that maximum attention to each of the varied phases of the work at Cornell has been impossible and progress in any one direction has been slower than desired. Progress upon the new furnace, for example, has been hampered by the emphasis upon practical testing which 6b7 felt was essential to the continued support of the practical foundryman.

As work progressed, however, it became more and more apparent that division of the work at Cornell was resulting in lost motion, and in the last analysis that the fundamental investigations would be essential to optimum interpretation and use of the practical test results. At the meeting of the Sand Committees at Cleveland, December 6 and 7, 1945, it was unanimously decided to focus attention on the research furnace until it had been brought to a state of perfection suitable for fundamental work, and then to undertake programs of practical investigation.

Accordingly, the report on the work done to date on the single variable program is relegated to an appendix to the main topic. The original program of co-operative testing between Cornell University and the Naval Research Laboratory has been temporarily discontinued until attention is again focussed on the practical test work.

In addition to his formal report Mr. Williams has prepared a color movie showing the sand specimens subjected to different conditions of handling and heating. This is his first attempt to record the behavior of foundry sands at elevated temperatures on film and although not the polished job of a professional it is clear that much can be learned from movies to supplement that

Sixth Annual Progress Report of the investigation of steel foundry sands at elevated temperatures, sponsored by the Subcommittee on Steel Sands, Committee on Physical Properties of Foundry Sands at Elevated Tempertures and conducted at Cornell University.

Presented at a Sand Research Session at the Fiftieth Annual Meeting of the American Foundrymen's Association, May 8, 1946, at Cleveland.

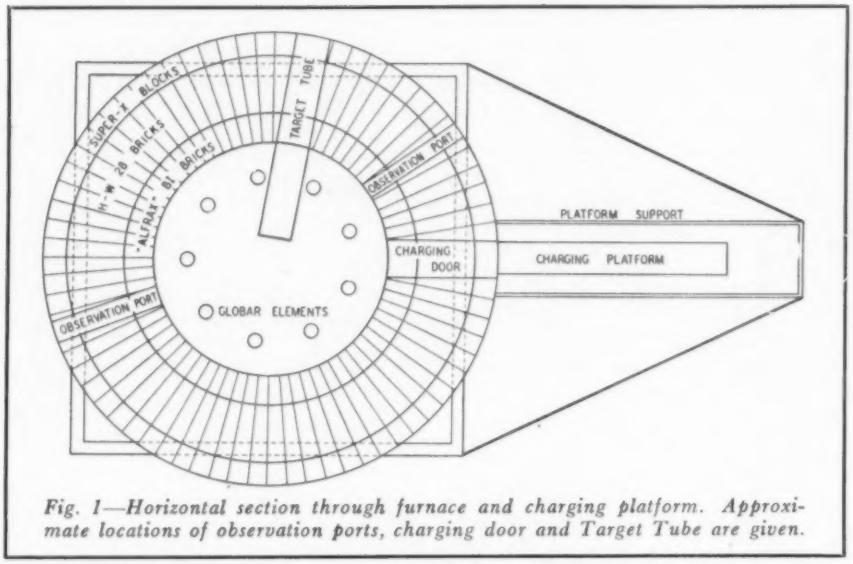


Fig. 1—Horizontal section through furnace and charging platform. Approximate locations of observation ports, charging door and Target Tube are given.

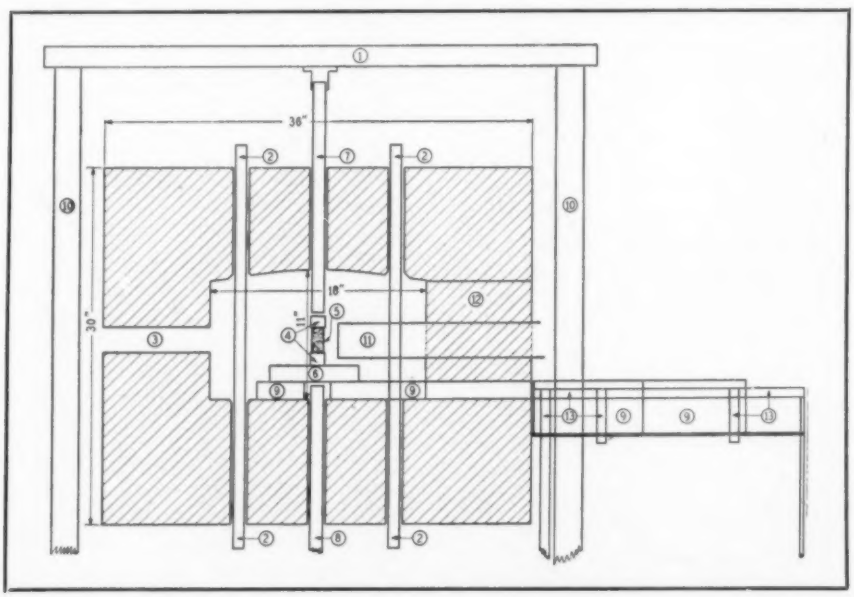


Fig. 2 (Above)—Vertical section through furnace and charging platform. Specimen location is shown as well as refractory specimen support that operates as a skid. Parts are identified as follows: (1) rigid top cross-piece; (2) nine Globar heating elements; (3) observation port; (4) ceramic disks; (5) test specimen; (6) compression and transverse specimen support; (7) fixed upper loading post; (8) moving lower loading post; (9) ceramic skidway charging platform; (10) top cross-piece support; (11) radiation pyrometer target tube; (12) charging door (full height of furnace); (13) charging platform support.

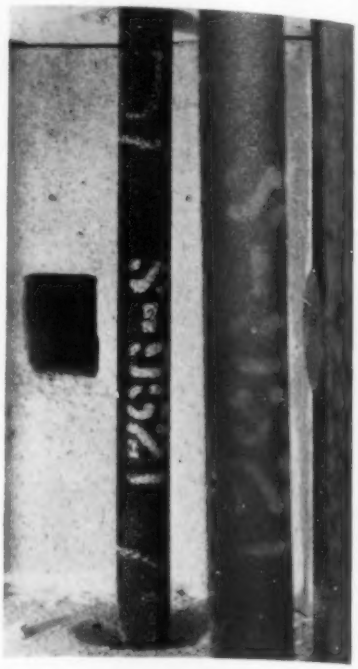
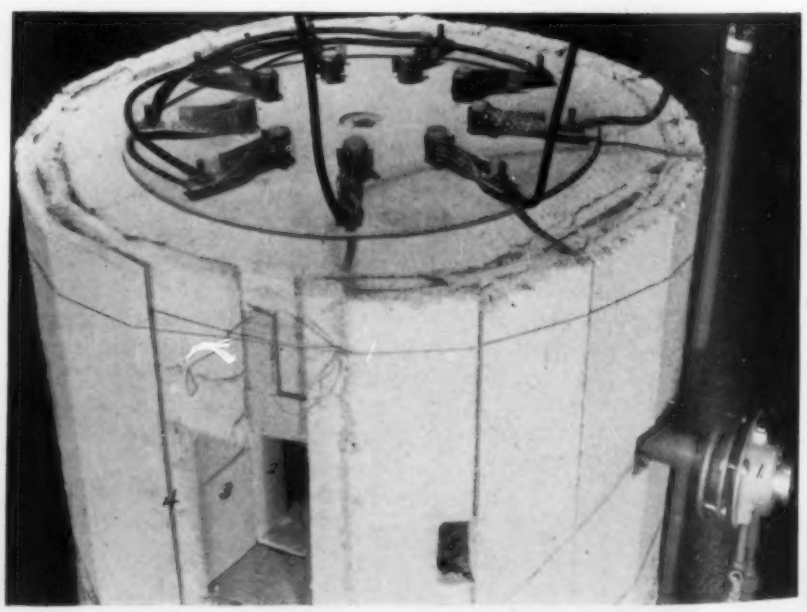


Fig. 3—View of interior of furnace showing (1) observation port opening, (2) Globar heating element, (3) Target Tube end (Rayotube sighted upon this end for temperature control), (4) chamfered opening for heating elements.

Fig. 4 (Below)—Outside view showing furnace without sheet metal casing; (1) total radiation pyrometer with Target Tube for temperature control; (2) inner lining; (3) intermediate insulation; (4) outer insulation; (5) observation port; (6) electrical connections to heating elements.



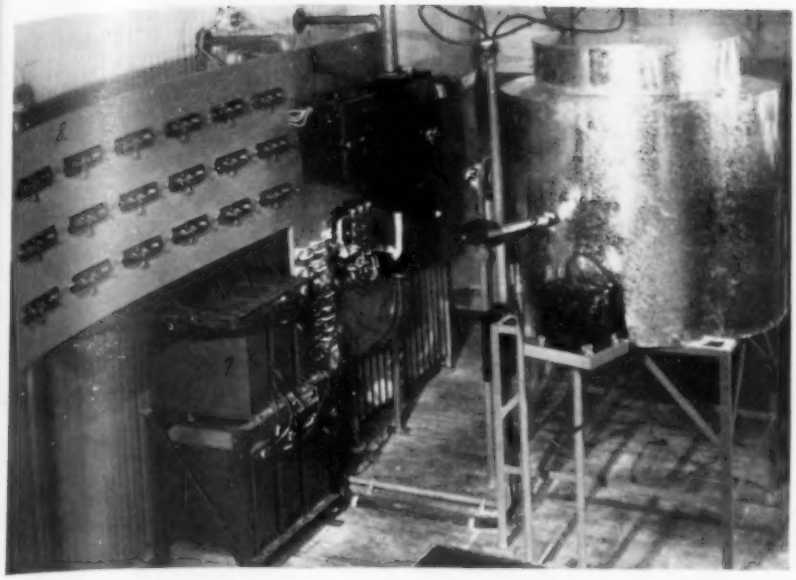


Fig. 5—View of testing equipment; (1) furnace; (2) optical pyrometer; (3) Rayotube; (4) main switch; (5) controller; (6) magnetic contactor; (7) capacitors, and (8) switch panel.

which can be seen by the naked eye or measured by instruments.

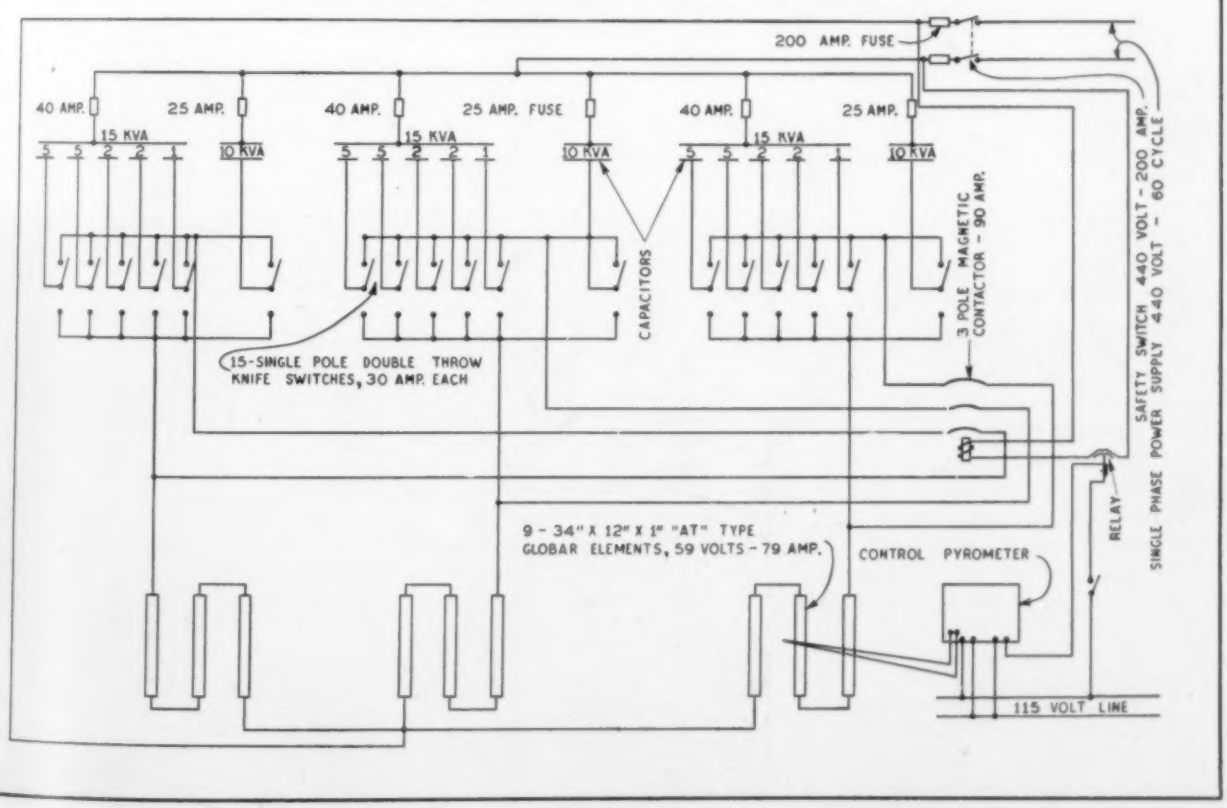
Mr. Finster, in his report as Chairman of 6b7 for 1944, made the following observation, "To pursue its work, this Sub-committee needs the whole-hearted support of the members of A.F.A. and their patient understanding that fundamental information must first be gained before practical application can be carried out intelligently."

His words are as applicable today as they were then except that we have more closely approached the time when we can obtain the accurate, reproducible, fundamental results needed. In every new field of endeavor progress is always slow, and in spite of its unquestioned value, research never appears economical of time or money. Your chairman is certain that the work of the next year will prove the foresight and patience of A.F.A. in entering this field of investigation well founded.

I take this opportunity to express appreciation for the interest and assistance of all persons connected with the work of 6b7, particularly Professor J. R. Moynihan and Dr. H. Reis of Cornell University and the persons who comprise the membership of the Sub-committee. Especial acknowledgment is due the Research Fellow, D. C. Williams for his tireless effort and unflagging interest in his work.

Sub-committee personnel are:
Howard F. Taylor, *Chairman*
John Rassenfoss, *Vice-Chairman*
Charles W. Briggs
Harry W. Dietert
Werner Finster

Fig. 6—Electrical wiring diagram for elevated temperature furnace.



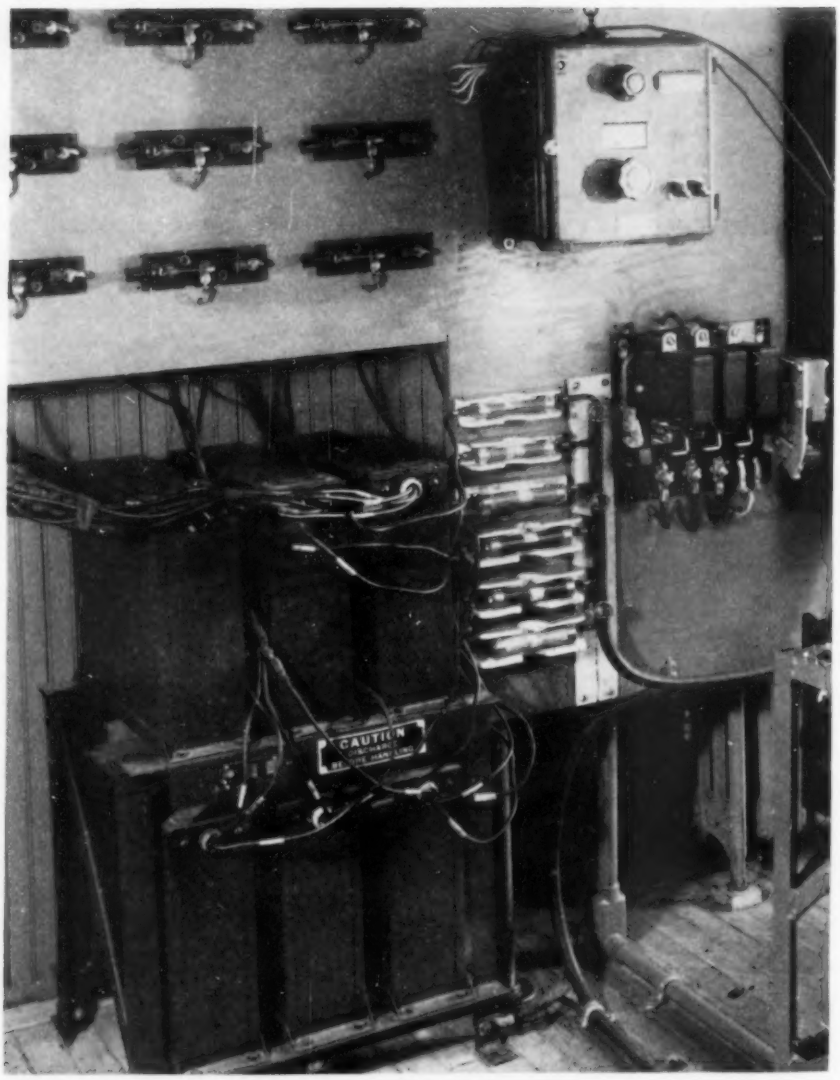


Fig. 7—View showing installation of Capacitors.

1. The test specimen should be seen when being tested.

2. The furnace should have large heat capacity and power input so that introduction of the test specimen would cause little or no drop in furnace temperature.

3. Within the testing furnace the zone around the test specimen should be of uniform temperature.

4. The furnace should be large enough to make transverse tests at elevated temperatures.

5. The hydraulic loading system should be equipped with a means to vary rate of application of load.

Furnace Construction

In designing the furnace the writer consulted Mr. B. A. Bovee, Chief Engineer, Globar Division, Carborundum Company and construction of the furnace was started the latter part of 1944, at Cornell University. The University, through Dean S. C. Hollister, provided the necessary funds to procure required materials.

Figures 1 and 2 show horizontal and vertical sections through the furnace. The diameter of the furnace is 18 in., the chamber height is 11 in., and 9 heating elements are used.

Clyde Jenni
Hobart N. Kraner
J. R. Moynihan
Robert E. Morey
Earl Woodliff
Douglas Williams

Respectfully Submitted,
Howard F. Taylor, *Chairman.*

Research Furnace

To clearly indicate characteristics of the new research furnace it appears desirable to describe its construction in detail before recording the few experimental results obtained during the short time the unit has been in operation. For fundamental research upon foundry sands at elevated temperature it was determined necessary that the equipment have the facilities to meet the following requirements:

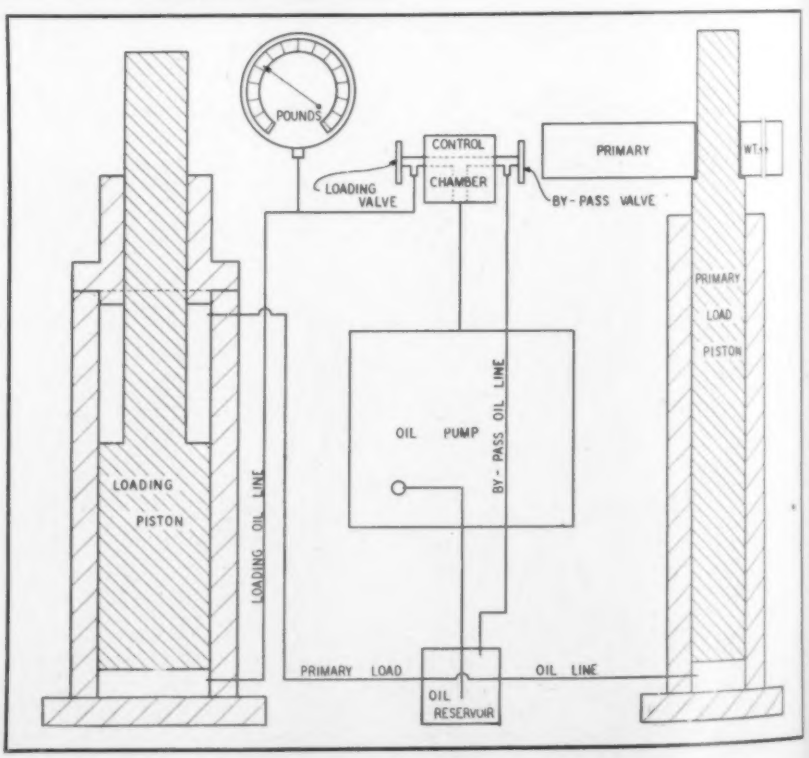


Fig. 8—Diagram of hydraulic pressure loading system.

Table 1

HOT COMPRESSIVE STRENGTH OF TEST SPECIMENS AS DETERMINED USING NEW RESEARCH FURNACE AND COMMERCIAL DILATOMETER

Mixture No.	Moisture Content, per cent	Exposure Time in Min.	Temperature Control Setting, °F.	Hot Compressive Strength in psi. Using Research Furnace	Hot Compressive Strength in psi. Using Commercial Dilatometer
8	5.3	1	2500	32	42
8	5.3	2	2500	9	13
8	5.3	4	2500	20	49
3	5.3	2	2500	3	22
3	5.3	16	2500	0	7
4	5.3	1	2500	0	6
4	5.3	2	2500	8	12
4	5.3	4	2500	0	0
4	3.4	1	2500	0	2
4	3.4	2	2500	7	7
4	3.4	4	2500	0	0
4	3.4	8	2500	0	0
4	3.4	12	2500	0	0
4	3.4	16	2500	0	0
7[1]	3.2	16	2500	22	18
3[1]	5.3	2	2500	3	38
3[1]	5.3	8	2500	0	7
3[1]	5.3	16	2500	0	10
7[2, 1]	5	8	1600	66	63
7[2, 1]	5	16	1600	88	84
1[4, 2]	3.4	2	1600	25	27
1[4, 2]	3.4	8	1600	58	84
1[4, 2]	3.4	16	1600	47	179
1[4, 1]	3.4	2	1600	31	27
1[4, 1]	3.4	8	1600	93	104
1[4, 1]	3.4	16	1600	97	272

[1]6 rams [2]7.5 min. wet mixing [3]3 rams [4]10 min. wet mixing

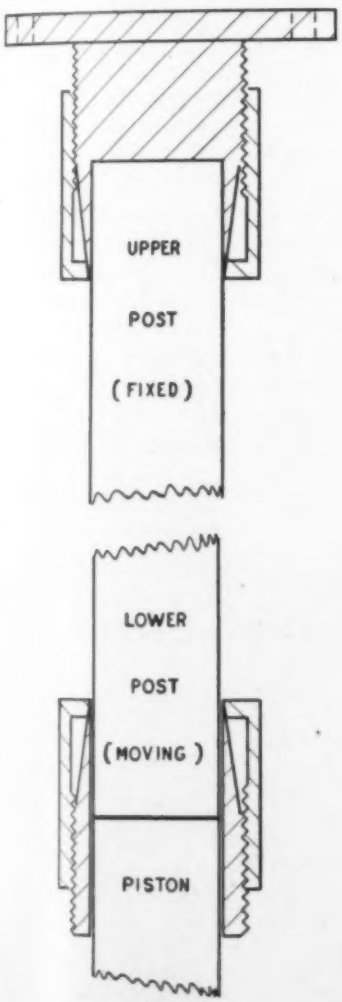


Fig. 9—Sketch of post holders used on research testing equipment.

For a furnace of this size the operating cost (electric power) is not as great as was expected. Below are shown power consumption and temperatures (optical pyrometer measurements) of a heating element and furnace wall.

Table 2

SAND MIXTURES TESTED

Mixture No.	Ingredients	Per Cent
1	Western Bentonite	4
	Sand	96
2	Western Bentonite	4
	Silica Flour	10
	Sand	86
3	Western Bentonite	4
	Corn Flour	1
	Sand	95
4	Southern Bentonite	4
	Sand	96
5	Southern Bentonite	4
	Silica Flour	10
	Sand	86
6	Southern Bentonite	4
	Corn Flour	1
	Sand	95
7	Fire Clay	4
	Sand	96
8	Fire Clay	4
	Corn Flour	1
	Sand	95

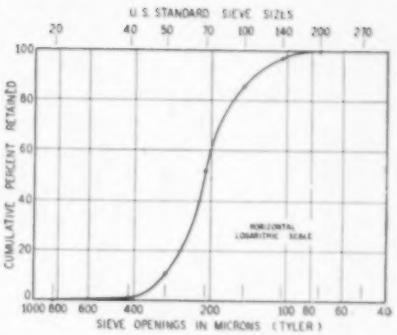


Fig. 10—Cumulative per cent distribution curve for N. J. No. 60 sand used in investigations. Relative location of U. S. Standard Sieve sizes are given.

(1) T.C.S. (temperature control setting).....1960° F.
Power consumption during 45 hr. aver. 6.2 kw. per hr.
Temperature of heating element.....2080° F.
Temperature of furnace wall .....1970° F.

(2) T.C.S.....2500° F.
Power consumption during 40 hr. aver. 8.3 kw. per hr.
Temperature of heating element .....2610° F.
Temperature of furnace walls .....2505° F.

Much more insulation would be required to reduce the temperature difference between the elements and the furnace wall. This 100° F. temperature difference, at the present time, is not considered harmful.

Figures 3 and 4 are two views of the furnace showing construction. Figure 3 is a view looking into the furnace through the charging door and Figure 4 shows the almost completed furnace without the sheet metal casing. Figure 5 shows the furnace and control equipment except the hydraulic loading system. The sheet metal top was found unnecessary and is no longer used.

In order to secure best results with the Globar heating elements, capaci-

Table 3

SCREEN ANALYSIS OF No. 60 N. J. TEST SAND

U.S. Standard Sieve Size	Per Cent Retained
On 20	0.00
40	0.51
50	10.15
70	40.95
100	35.40
140	10.90
200	1.92
270	Trace
Pan	Trace
	99.83

Table 4

HOT COMPRESSIVE STRENGTH TEST VALUES FROM SEVERAL SERIES OF TESTS UPON THE SAME MIXTURES SHOWING THE VARIATION OBTAINED FROM SERIES TO SERIES

T.C.S. Exposure Time in Min.	1600° F. 2	1600° F. 8	1600° F. 16	2000° F. 2	2000° F. 8	2000° F. 16	2500 F. 2	2500 F. 8	2500 F. 16
				Mixture No. 1					
Test Series No. 1	127	338	405	126	153	173	221	6	0
2	147	400	580	182	183	221	108	6	0
3	151	519	632	99	177	159	—	—	—
4	70	277	410	—	—	—	—	—	—
				Mixture No. 2					
1	194	850	over 1000	366	307	304	194	24	22
2	176	414	391	190	378	396	224	50	46
				Mixture No. 3					
1	99	277	236	102	65	147	46	0	0
2	103	325	370	165	66	158	22	2	6
3	81	312	450	130	61	133	—	—	—
				Mixture No. 4					
1	40	131	215	40	29	87	21	4	0
2	31	164	272	40	24	65	20	1	0
3	27	107	193	55	44	82	17	0	4
4	44	148	256	—	—	—	125	5	5
5	—	—	—	—	—	—	27	0	0
6	—	—	—	—	—	—	27	0	0
				Mixture No. 5					
1	60	341	620	150	78	178	131	9	4
2	176	359	366	147	84	216	80	9	20
3	—	—	—	75	52	122	120	2	0
				Mixture No. 6					
1	59	125	185	40	29	87	20	4	0
2	31	164	272	40	24	65	14	0	4
3	41	123	232	43	34	92	92	5	5
4	50	115	247	—	—	—	9	0	0
				Mixture No. 7					
1	36	51	64	78	302	343	26	34	40
2	21	55	78	31	172	202	29	2	0
3	—	—	—	18	144	231	—	—	—
				Mixture No. 8					
1	56	103	90	36	147	212	24	5	5
2	37	76	92	33	227	270	13	3	11
3	32	42	40	20	169	193	42	11	10

tors were used to supply continuous rather than intermittent power. When the equipment is operating, sufficient capacitance is switched on continuously to maintain a temperature slightly below the chosen T.C.S. This requires the controller to adjust a relatively small amount of additional capacitance to maintain the desired T.C.S. Figure 6 gives the wiring diagram used and Figure 7 shows installation of the apparatus.

Since a study of the effect of the rate of loading of test specimens is to be one of the subjects of investigation it was necessary to provide a means for variable loading. This was done by allowing oil from a constant speed pump to flow to a chamber to which two valves are connected. The rate of piston travel is controlled by opening or closing the by-pass valve. An auxiliary primary load piston is incorporated to increase the tare weight. This feature is desirable when small loads are applied to the test specimens and eliminate any fluctuation of the gauge pointer due to pump action. The oil line from the primary weight piston is connected to the upper side of the loading piston. Figure 8 shows a diagram of the hydraulic system as suggested and made by A. W. Jenkins, Cornell University.

Centering of the ceramic loading posts is accomplished by a type of holder suggested by Associate Professor W. C. Andrae, Cornell University and is shown in Fig. 9. Several slots are cut in the tapered portion of the holder fitting around the post. By simply tightening the outer knurled collar the post is centered accurately and firmly. This method can be adapted to any dilatometer.

The use of the "AT" type Globar heating element has eliminated the need for cumbersome cooling water and drain connections used for Glotube and other installations.

A Rayotube (total radiation pyrometer) sighted into a target tube is used for temperature control instead of a thermocouple whose hot junction is exposed. The use of an unprotected platinum—90 per cent platinum, 10 per cent rhodium thermocouple hot junction for continued temperature control above 2400° F. is not satisfactory for research work. The Rayotube has its target tube end about 2 in. from the test specimens. Because of the large heat capacity of the furnace and small temperature drop on the introduction of the test specimens the reaction to temperature changes is adequate. However, too few experiments have been made, to date, to determine the overall satisfactoriness of the Rayotube for this purpose.

Test Results. Since completion of the research furnace only a small amount of time has been available for test work. To date it has been possible to make only superficial investigation of the comparison between specimens tested in the research furnace and the commercial dilatometer (1944 model). Table 1

Table 5

HIGH COMPRESSIVE STRENGTH OF TEST SPECIMENS WITH A BATCH MOISTURE CONTENT OF 3.4 ± 0.2 PER CENT

T.C.S. Exposure Time in Min.	1600° F. 2	1600° F. 8	1600° F. 16	2000° F. 2	2000° F. 8	2000° F. 16	2500° F. 2	2500° F. 8	2500° F. 16
				Mixture No. 1					
Test Series No. 1	115	368	551	183	301	344	12	—	—
2	120	422	508	182	183	221	—	—	—
3	57	204	417	118	225	206	—	—	—
				Mixture No. 4					
1	88	283	453	25	19	45	12	5	0
2	22	71	118	27	15	34	6	0	0
				Mixture No. 7					
1	—	—	—	—	—	—	23	7	13
2	—	—	—	6	71	132	8	5	17

gives the results of parallel tests for the H.C.S. (hot compressive strength) using the two pieces of testing equipment. The mixtures referred to by number are indicated in Table 2. The rate of loading used was 1 in. per min. table rise under no-load condition for both units. The specimen breaking time for each furnace averaged 6 sec. at a T.C.S. of 1600° F. Since so many H.C.S. tests at a T.C.S. of 2500° F. gave zero psi. an average breaking time probably has little meaning.

From Table 1 it will be seen that particularly for the shorter exposure times the test values check reasonably well for approximately half the tests.

Material. For the tests reported in Table 1, No. 60 N. J. Sand (washed and dried) was used. Figure 10 shows the cumulative distribution curve for this sand whose mechanical sieve analysis is shown in Table 3.

The western and southern bentonites, fire clay, corn flour and silica flour used as additions to the sand were obtained from commercial sources. A commercial moisture teller was used for these determinations. The specimens were tested after being given 3 or 6 rams.

General Information. One type of split specimen tube has been in use at Cornell University for over a year. A detailed account of comparison tests between the split type and usual one-piece specimen tubes was reported by the writer[1]. The conclusions of that report are given as follows: "A split specimen tube has been found to be superior to a one-piece specimen tube for the following reasons:

1. The rammed "structure" of the test specimen is better preserved.
2. Specimens from a certain sand mixture could not be stripped from the one-piece tube without damage.
3. With specimens from sand mixtures containing western bentonite the reproducibility of test values within any one batch was consistently appreciably higher.
4. The type of split tube used in the investigation was found to be easier to use than the one-piece tube."

All hot compressive strength data was obtained from test specimens (1-1/8 in. diameter and 2 in. high) under the following general conditions:

1. Green sand specimens were used.
2. Specimens were double-end rammed.
3. Each test specimen in all tests was unconfined on its surface parallel to its axis.
4. The atmosphere within the testing furnace will be called oxidizing. This is to distinguish it from the procedure of testing under "own atmosphere."

The mold for making transverse test specimens has just been made and data for this type of test will be obtained for a future report.

The latest model of the commercial dilatometer has incorporated in its design features found on the research furnace at Cornell University.

Table 7

HOT COMPRESSIVE STRENGTH OF SAND TEST SPECIMENS WITH 4 AND 10 PER CENT FIRE CLAY IN MIXTURE

T.C.S. Exposure Time in Min.	*1600° F.*			*2000° F.*			*2500° F.*		
	2	*8*	*16*	*2*	*8*	*16*	*2*	*8*	*16*
Per Cent Fire Clay									
Three Rams at 3.4 Per Cent Moisture									
4	7	27	31	6	71	132	8	5	17
10	17	39	82	34	254	337	42	8	19
Six Rams at 3.4 Per Cent Moisture									
4	22	30	44	14	108	140	14	7	20
10	26	107	81	34	294	379	40	10	22
Three Rams at 5.4 Per Cent Moisture									
4	21	55	78	18	144	231	26	34	40
10	55	192	139	70	454	578	85	18	24
Six Rams at 5.4 Per Cent Moisture									
4	27	67	79	24	178	202	15	10	22
10	77	174	259	92	443	671	104	22	33

Table 6

HIGH COMPRESSIVE STRENGTH OF SAND TEST SPECIMENS UNDER GIVEN CONDITIONS

T.C.S. Exposure Time in Min.	*1600° F.*			*2000° F.*			*2500° F.*		
	2	*8*	*16*	*2*	*8*	*16*	*2*	*8*	*16*
Per Cent Bentonite	*Three Rams*								
2	71	277	438	84	173	172	85	7	5
4	70	277	410	99	177	150	108	6	0
6	138	562	741	270	305	300	147	7	7
	Six Rams								
2	97	307	544	116	215	196	100	7	7
4	97	322	525	122	217	208	100	2	2
6	140	587	853	227	347	347	177	7	7

Appendix

This appendix contains information and data obtained for the Foundry Mixes "single variable mentioned by Chairman Taylor in his introduction.

Since the A.F.A. was given a 1944 model dilatometer we have accumulated considerable H.C.S. data for the eight mixtures shown in Table 2. For convenience, the mixture numbers will be referred to throughout this report.

All data reported was obtained from testing green sand specimens. A limited amount of data from tests on dry sand specimens indicates no change in the trends developed by the green sand tests. For all tests, 1-3/8 in.×1 in. "mullfrax" S disks were used.

The writer in his Fourth Annual Progress Report* suggested that it be kept in mind that the hot compressive strength test values should be

[1] D. C. Williams "Split Type Specimen Tube for Elevated Temperature Sand Testing," *Transactions*, AMERICAN FOUNDRYMEN'S ASSOCIATION, vol. 53, p. 248, (1945).

*D. C. Williams, "Fourth Progress Report on Investigations of Physical Properties of Steel Foundry Sands at Elevated Temperatures," *Transactions*, AMERICAN FOUNDRYMEN'S ASSOCIATION, p. 979, vol. 52 (1944).

Table 8

HOT COMPRESSIVE STRENGTH TEST DATA FROM LAST SERIES OF TEST FOR 8 MIXTURES. THIS DATA WAS USED FOR THE BASIS OF CERTAIN GENERAL STATEMENTS CONCERNING THE EFFECT OF SOME VARIABLES UPON THE HOT COMPRESSIVE STRENGTH

T.C.S.	*1600° F.*			*2000° F.*			*2500° F.*		
Exposure Time in Min.	*2*	*8*	*16*	*2*	*8*	*16*	*2*	*8*	*16*
Mixture No.									
	Three Rams at 3.4 Per Cent Moisture								
1	57	204	417	118	225	206	81	3	3
2	104	402	544	182	345	380	201	29	31
3	19	89	252	54	63	104	63	6	7
4	22	71	118	27	15	34	6	0	0
5	44	239	388	75	52	122	74	9	4
6	24	73	123	2	7	9	9	0	0
7	7	27	31	6	71	132	8	5	7
8	7	0	0	2	11	16	3	1	2
	Six Rams at 3.4 Per Cent Moisture								
1	92	383	434	163	273	247	106	2	7
2	183	773	902	254	469	462	265	46	47
3	48	202	307	83	79	133	85	7	7
4	24	80	187	30	25	40	13	0	0
5	95	327	592	102	70	205	94	19	20
6	30	102	173	2	7	13	12	0	0
7	22	30	44	14	108	140	14	7	20
8	7	0	1	0	13	19	5	5	12
	Three Rams at 5.4 Per Cent Moisture								
1	70	277	410	99	177	159	108	6	0
2	194	850	1000	190	378	396	256	50	46
3	81	312	450	130	61	133	22	2	6
4	27	107	193	41	26	51	27	0	0
5	60	341	620	104	71	182	83	9	20
6	50	115	247	48	34	74	11	0	0
7	21	55	78	18	144	231	26	34	40
8	32	42	50	20	169	193	13	3	11
	Six Rams at 5.4 Per Cent Moisture								
1	97	322	525	122	217	208	100	2	2
2	284	1000	1000	264	463	478	269	24	33
3	88	420	507	172	79	157	28	7	10
4	27	142	207	47	27	72	18	0	0
5	102	487	633	112	92	254	96	17	26
6	62	157	340	62	39	87	22	0	0
7	27	67	79	24	178	303	15	10	22
8	27	50	64	20	178	252	17	6	12

considered as "nominal" values since they are referred to the original cross-sectional area of the test specimen. The actual cross-sectional area of either the rind or the core, produced by the exposure of the test specimen to shock heating, is less than the original cross-sectional area of the test specimen.

This cross-section area of the rind and core has been found to vary according to the mixture tested. If the "true" cross-sectional area of the part of the test specimen, that is, either the rind or the core, which supported the externally applied load was known, it is entirely probable that our present information regarding the effect of variables would undergo an alteration.

Summary of Several Series of Tests. Table 4 gives H.C.S. data obtained from several series of tests upon similar mixtures made from the same stock of ingredients under similar conditions of testing. Each test specimen was given three rams and the moisture content of all sand batches was 5.4 per cent plus or minus 0.2 per cent. Our limited knowledge of optimum mixing techniques and the method of ramming the test specimens may possibly contribute to the spread in test values.

While there is not complete agreement of test data between the various series of tests upon the same mixture, several unmistakable trends are evident. Later in this report the discussion of room temperature test variations indicates we might expect poor reproducibility between sand batches.

Data found in Table 4 indicates the following general conclusions:

1. At a T.C.S. (temperature control setting) of 1600° F. the H.C.S. of the test specimens increases as the time of specimen exposure to heat increases.

2. Test values for all series of tests at a T.C.S. of 2000° F. indicate a smaller magnitude of variation than at a T.C.S. of 1600° F. Comparison with tests at a T.C.S. of 2500° F. is perhaps unwise because of the very low H.C.S. values at this T.C.S.

3. At a T.C.S. of 2000° F. fire clay probably develops the greatest H.C.S. after a specimen exposure to heat of 8 min.

4. At a T.C.S. of 2500° F. the H.C.S. is greatest at the 2-min. exposure and at longer exposures the H.C.S. is generally below 10 psi.

5. Western bentonite imparts greater H.C.S. to the test specimen than either southern bentonite or fire clay. The possible exception is the longer specimen exposure times for fire-clay at a T.C.S. of 2000° F.

6. Silica flour increases the H.C.S.

7. A one per cent addition of corn flour does not significantly effect the H.C.S.

Table 9

EFFECT UPON HOT COMPRESSIVE STRENGTH WHEN RAMMING OF TEST SPECIMEN IS VARIED FROM 3 TO 6 RAMS

T.C.S.	*1600° F.*						*2000° F.*						*2500° F.*					
Rams—	*3*			*6*			*3*			*6*			*3*			*6*		
Exposure Time in Min.	*2*	*8*	*16*	*2*	*8*	*16*	*2*	*8*	*16*	*2*	*8*	*16*	*2*	*8*	*16*	*2*	*8*	*16*
Mixture No.																		
1	In.	In.	—	—	In.	In.	De.	De.	De.	De.	De.	De.	In.	—	—	—	—	—
2	In.	In.	In.	In.	In.	In.	—	—	—	—	—	—	In.	In.	In.	—	In.	In.
3	In.	In.	In.	In.	In.	In.	In.	—	In.	In.	—	In.	De.	—	—	De.	—	—
4	—	In.	In.	—	In.	—	In.	In.	In.	In.	—	In.	In.	—	—	In.	—	—
5	In.	In.	In.	—	In.	—	In.	In.	In.	—	In.	In.	—	—	In.	—	In.	In.
6	In.	In.	In.	In.	In.	In.	In.	In.	In.	In.	In.	In.	—	—	—	In.	—	—
7	In.	In.	In.	—	In.	In.	In.	In.	In.	In.	In.	In.	In.	In.	In.	—	—	—
8	In.	In.	In.	In.	In.	In.	In.	In.	In.	In.	In.	In.	In.	—	—	In.	—	—
	91			75			70			63			41.5			29		
Per Cent Showing	83						67						35.5					

In. = Increase in H.C.S. De. = Decrease in H.C.S. — = No change in H.C.S.

8. Certain mixtures (3-4-5-6) at a T.C.S. of 2000° F. have their lowest H.C.S. at an exposure time of 8 min.

We have much less test data from different series of tests at a batch moisture content of 3.4 per cent plus or minus 0.2 per cent. This data is collected in Table 5.

From Table 5 it is evident that the H.C.S. of test specimens bonded with Western bentonite is greater than when bonded with Southern bentonite.

Varying the Amount of Bentonite Addition. A limited study was made concerning the effect of 2, 4 and 6 per cent western bentonite. The moisture of all batches was 5.4 per cent and all test specimens were given 3 or 6 rams and tested in the green state. Table 6 gives the H.C.S. data obtained.

The following conclusion can be made from the data Table 6:

1. Two and 4 per cent western bentonite additions will be found to have the same H.C.S. for any T.C.S.

2. A 6 per cent western bentonite addition roughly doubles the H.C.S. when compared to the 2 and 4 per cent additions.

3. An increase in ramming increased the H.C.S. except for certain test conditions at a T.C.S. of 2500° F. (very low test values).

Varying the Amount of Fire Clay Addition. A limited study was made concerning the effect of 4 and 10 per cent fire clay in a sand-fire clay mixture. Table 7 gives the H.C.S. data obtained.

A change in the amount of fire clay from 4 to 10 per cent produced an increase in the H.C.S. Increasing the moisture content increases the H.C.S. whereas increasing the ramming apparently has little effect at either level of fire clay addition. This observation is in line with the general statements deduced from Tables 11 and 12.

Latest Series of Tests

The data from recently completed tests is given in Table 8. The specimens were tested in the green condition and were given 3 or 6 rams. Two moisture contents (3.4 and 5.4 per cent) were used. The specimens were exposed to heat for 2, 8 and 16 min. For any particular test condition five tests were averaged for the 3-rammed specimens, while three tests were averaged for the 6-rammed specimens. The number of tests conducted for this series are: (1) H.C.S.—1550, (2) Moisture—130.

Effect of Moisture Variable. Trend in the effect of varying the moisture upon the H.C.S. can be determined from the data of the test series shown in Table 8. Varying the moisture content is considered to have no effect when, (1) compared values are within 10 per cent and (2) the psi. is 10 or less. The effect of moisture at any condition of testing is shown in Table 9. It is commonly accepted that increasing the moisture increases the H.C.S. but none of the eight mixtures tested has a uniform reaction to an increased moisture content. Table 9 gives the relative effect between mixtures containing the three types of clay at two degrees of ramming.

The information from Tables 9 and 11 indicate that (1) mixtures containing fire clay are most responsive to an increase in H.C.S. as the moisture content increases, (2) that mixtures containing western bentonite are least responsive to an increase in H.C.S. as the moisture content increases, and (3) under certain conditions an increase in moisture content of western bentonite mixtures will decrease the H.C.S.

Effect of Varying the Number of Rams. The effect of varying the ramming upon the H.C.S. is determined from the data of the test series shown in Table 8. The effect of ramming at any condition tested is shown in Table 10.

Table 12 gives the relative effects between mixtures containing the three types of clay.

The information from Tables 10

Table 10

EFFECT UPON HOT COMPRESSIVE STRENGTH WHEN MOISTURE CONTENT OF MIXTURES IS VARIED FROM 3.4 TO 5.4 PER CENT

T.C.S.	1600° F.						2000° F.						2500° F.					
Per Cent Moisture—	3.4			5.4			3.4			5.4			3.4			5.4		
Exposure Time in Min.	2	8	16	2	8	16	2	8	16	2	8	16	2	8	16	2	8	16
Mixture No.																		
1	In.	In.	—	In.	In.	In.	In.	In.	In.	In.	In.	In.	In.	—	—	—	—	—
2	In.	In.	In.	In.	In.	—	In.	In.	In.	In.	In.	In.	In.	In.	—	De.	De.	De.
3	In.	In.	In.	—	In.	In.	In.	In.	In.	In.	In.	In.	In.	—	—	—	—	—
4	—	—	In.	—	In.	—	—	In.	In.	In.	—	In.	In.	—	—	—	—	—
5	In.	In.	In.	In.	In.	—	In.	In.	In.	—	In.	In.	In.	In.	In.	—	In.	—
6	—	In.	In.	In.	In.	In.	—	—	—	In.	In.	In.	—	—	—	In.	—	—
7	In.	—	In.	—	—	—	In.	In.	—	In.	In.	In.	In.	—	—	De.	De.	De.
8	—	—	—	—	In.	—	—	—	—	—	—	In.	—	—	—	—	—	—
Per Cent Showing Increase	67			50			67			83			40			8		
	62.5						75						25					

In. = Increase in H.C.S. De. = Decrease in H.C.S. — = No change in H.C.S.

Table 11

HIGH COMPRESSIVE STRENGTH TESTS ON SPECIMENS WITH THREE TYPES OF CLAYS IN MIXTURE

	—Due to Increase in Moisture Content Per Cent of Tests Showing—		
	Increase in H.C.S.	Decrease in H.C.S.	No Change in H.C.S.
Mixtures containing Western Bentonite	48	15	37
Mixtures containing Southern Bentonite	65	—	35
Mixtures containing Fire Clay	77	—	23

Table 12

TESTS ON SPECIMENS WITH THREE TYPES OF CLAYS UNDER GIVEN CONDITIONS

	—Due to Increase from 3 to 6 Rams, Per Cent Showing—		
	Increase in H.C.S.	Decrease in H.C.S.	No Change in H.C.S.
Mixtures containing Western Bentonite	70	4	26
Mixtures containing Southern Bentonite	55	—	45
Mixtures containing Fire Clay	28	8	64

Table 13

POSITION OF TEST SAND MIXTURES ARRANGED RELATIVE TO INCREASING HIGH COMPRESSIVE STRENGTH

T.C.S.	1600° F.		2000° F.		2500° F.	
Per Cent Moisture—	3.4	5.4	3.4	5.4	3.4	5.4
Mix. No.	7	7	8	7	8	7
	8	4	6	8	4	8
	3	8	7	4	7	4
	4	6	4	6	6	6
	6	5	3	1	3	3
	5	1	5	5	5	5
	1	3	1	3	1	1
	2	2	2	2	2	2

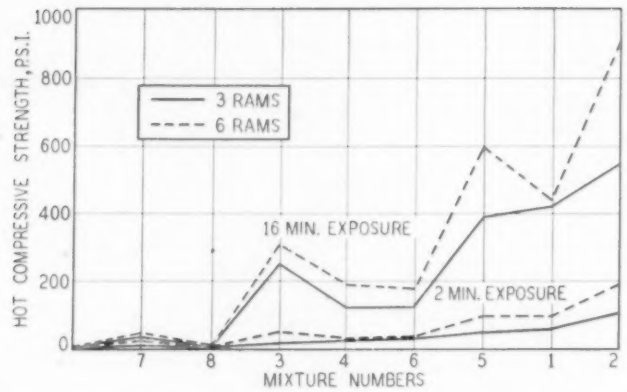


Fig. 11—Relative position of mixtures determined from hot compressive strength; moisture content, 3.4 per cent; temperature control setting, 1600 ° F.

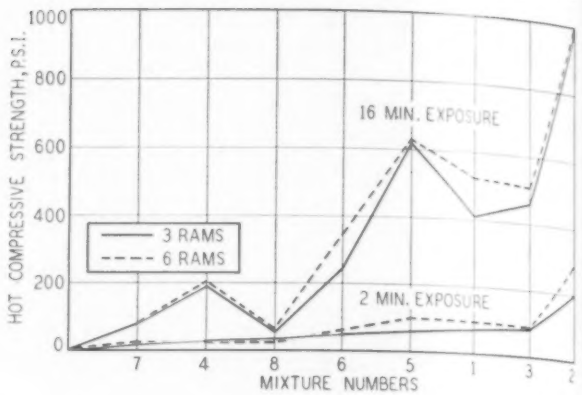


Fig. 12—Relative position of mixture determined from hot compressive strength; moisture content, 5.4 per cent; temperature control setting, 1600 ° F.

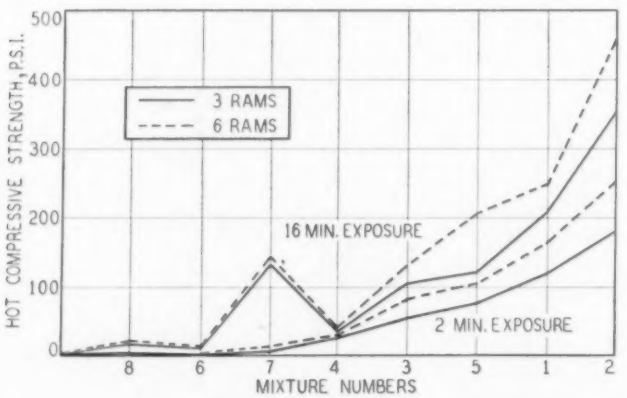


Fig. 13—Relative position of mixtures determined from hot compressive strength; moisture content, 3.4 per cent; temperature control setting, 2000 ° F.

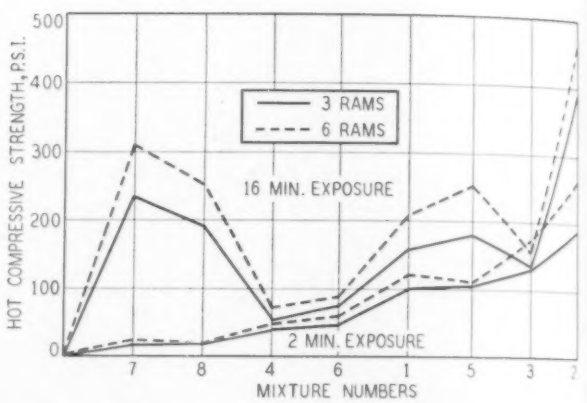


Fig. 14—Relative position of mixtures determined from hot compressive strength; moisture content, 5.4 per cent; temperature control setting 2000 ° F.

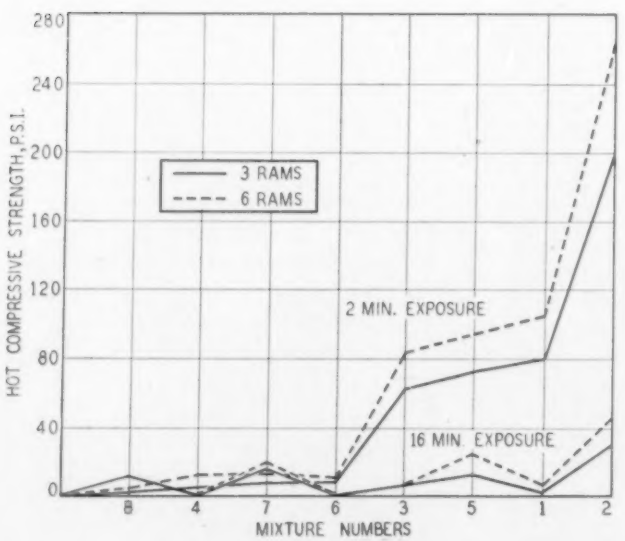


Fig. 15—Relative position of mixtures determined from hot compressive strength; moisture content, 3.4 per cent; temperature control setting, 2500 ° F.

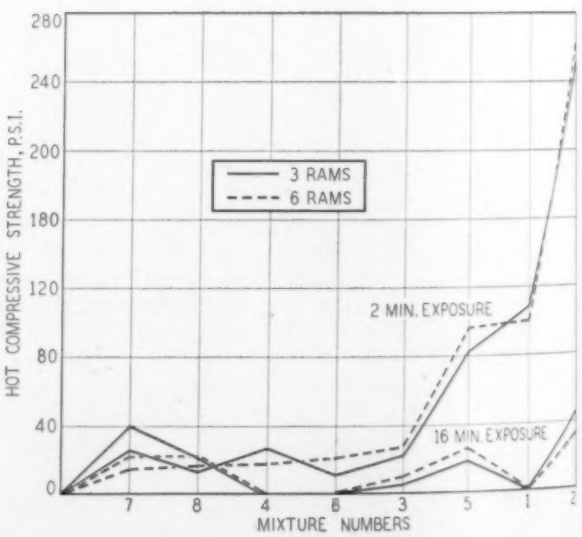


Fig. 16—Relative position of mixtures determined from hot compressive strength; moisture content, 5.4 per cent; temperature control setting, 2500 ° F.

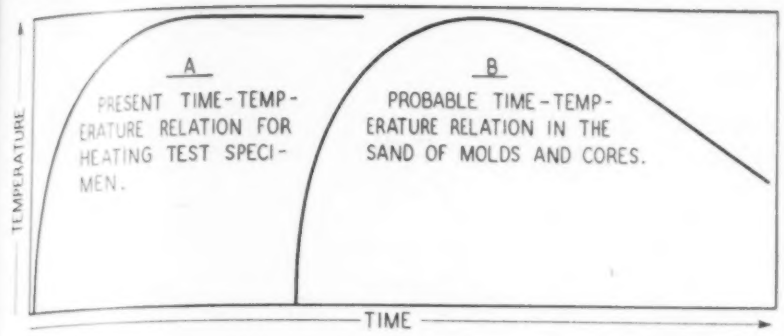


Fig. 17—Curve A shows present time-temperature relation for heating test specimen. Curve B shows probable time-temperature relation in the sand of molds and cores

and 12 indicate that (1) mixtures containing western bentonite are most responsive to an increase in H.C.S. as the ramming is increased, (2) mixtures containing fire clay are least responsive to an increase in H.C.S. as the ramming is increased.

Position of the Various Mixtures with Regard to Their H.C.S. Figures 11 to 16 show the relative position of the mixtures as judged by the H.C.S. data from Table 8. The curve for the 8-min. exposure time is generally parallel to the others and does not have much effect upon the relative positions. The results for the 2-min. exposure time are the basis for the order chosen and with few exceptions the 16-min. exposure time has no effect upon the relative order.

The position of the mixtures arranged relative to increasing H.C.S. is given in Table 13.

It will be noticed that the fire clay mixtures (7 and 8) will be found among the first three mixtures at any T.C.S. and moisture content, and the western bentonite mixtures (1 and 2 and occasionally 3) will be among the last three, with the southern bentonite mixtures (4 and 6) intermediate in the relative order.

New Type of H.C.S. Test. The present manner of heating the laboratory test specimen for H.C.S. can be described graphically by Figure 17-A. It is shown that the specimen reaches some temperature which is maintained. Figure 17-B is suggested as being possibly more applicable to the time-temperature relationships in molds and cores.

One laboratory test procedure which approaches the relationship shown by Figure 17-B has been tried. The test specimen is inserted into the testing furnace previously heated to the desired T.C.S. The test specimen is subjected to exposure at this T.C.S. for a predetermined length of time, after which the T.C.S. is lowered a chosen amount. At the time the lower T.C.S. is attained the test specimen is loaded to determine the H.C.S.

Time has allowed only a few tests of this type to be made at Cornell. The length of exposure at the higher T.C.S. was eight minutes. The T.C.S. was then lowered 250°-400°-500° or

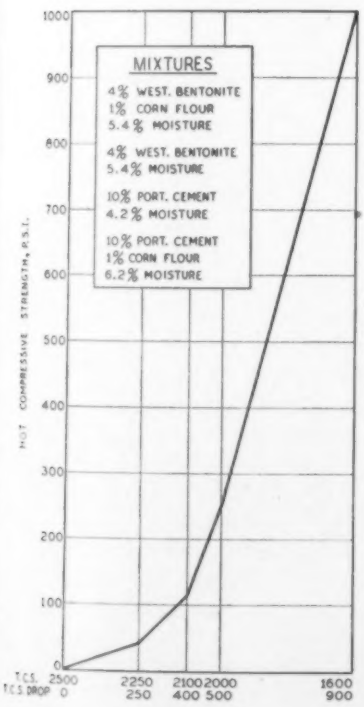


Fig. 18—Hot compressive strength of certain mixtures after an 8-min. exposure at 2500 ° F. followed by continued exposure until a chosen lower temperature control setting is reached.

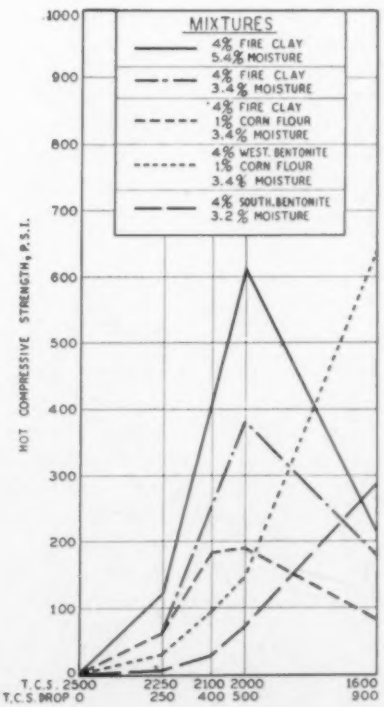


Fig. 19—Hot compressive strength of certain mixtures after an 8-min. exposure at 2500 ° F. followed by continued exposure until a chosen lower temperature control setting is reached.

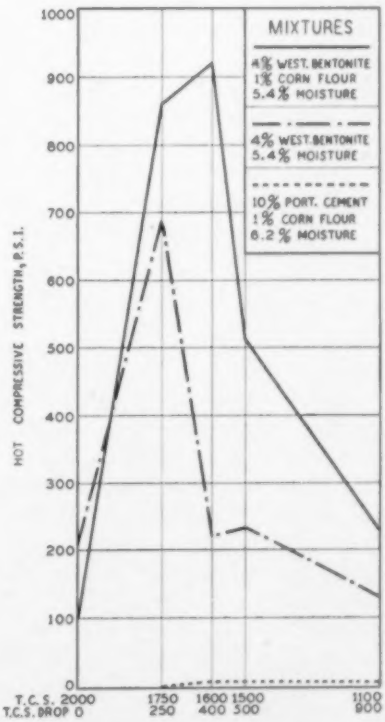


Fig. 20—Hot compressive strength of certain mixtures after an 8-min. exposure at 2000 ° F. followed by continued exposure until a chosen lower temperature control setting is reached.

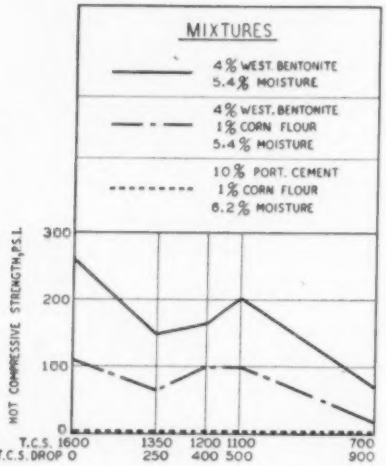
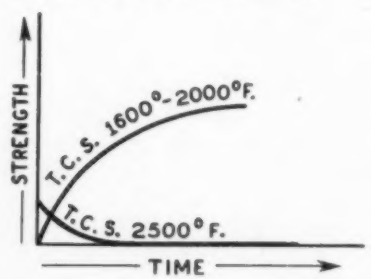


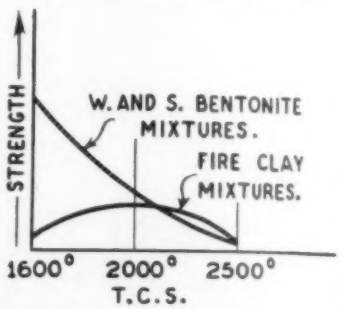
Fig. 21—Hot compressive strength of certain mixtures after an 8-min. exposure at 1600° F. followed by continued exposure until a chosen lower temperature control setting is reached.

Variable	*Effect*
Grain Size	Increases as grain size diminishes—50 mesh and 70 mesh sand tested.
Time of Temper in Sealed Container	No change during 6½ hr. after mixing—15-min. test intervals.
Time of Mixing	Little or no effect upon fire clay—varied effect upon Western Bentonite.
Silica Flour	Increases H.C.S. when added.
Moisture Content	Increases in 62 per cent of tests as moisture is increased: at T.C.S.-1600° F. — 83% increase; at T.C.S.-2000° F. — 67% increase; at T.C.S.-2500° F. — 35.5% increase
Ramming	Increases in 54 per cent of tests as ramming is increased: at T.C.S.-1600° F. — 62.5% increase; at T.C.S.-2000° F. — 75% increase; at T.C.S.-2500° F. — 25% increase
	Relative strength of clays (4 per cent additions)—West. Bentonite greater, South. Bentonite and Fire Clay about same.
	Corn Flour (1 per cent addition)—slight variable effect.
	Moisture variations most likely to vary H.C.S. of fire clay mixtures, West. Bentonite mixtures least likely.
	Ramming variations most likely to vary H.C.S. of West. Bentonite mixtures—fire clay mixtures least likely.
	Variation in West. Bentonite additions—2 and 4 per cent equal—6 per cent doubles H.C.S.
Type of Bond	Variation in fire clay additions—10 per cent fire clay doubles 4 per cent H.C.S.

H.C.S. = Hot Compressive Strength; T.C.S. = Temperature Control Setting.



Specimen exposure time (min.)—strength (psi.) relationship.



T.C.S. (°F.)—strength (psi.) relationship.

Fig. 22—General effect of variables upon the hot compressive strength of laboratory test specimens.

900° F. as the testing conditions required. As soon as the furnace attained the lower T.C.S. the load was applied to the test specimen. One curve is shown in Fig. 18 for all the mixtures tested.

Figure 19 shows the results obtained with other mixtures tested starting at a T.C.S. of 2500° F. Moisture content of the mixture appears to decidedly effect the H.C.S. Mixture No. 6 of Figure 18 and 5.4 per cent moisture while the same mixture of Fig. 19, had 3.4 per cent moisture. Also in Fig. 19, mixture No. 7 is compared at 2 moisture levels. Mixtures having fire clay additions indicate a lowering of the H.C.S. as the T.C.S. drop is increased. This may be important where early casting shakeout is the practice.

Figure 20 shows the results for mixtures tested when the beginning T.C.S. is 2000° F. It is to be noted that the portland cement mixture at any testing condition had little or no H.C.S. If the green cement mixture would withstand pouring conditions this mixture would be excellent from a shakeout viewpoint. In certain molds a layer of this sand might aid the shakeout if placed close enough to the casting to become heated to a high temperature.

Figure 21 shows the results for mixtures tested at an original T.C.S. of 1600° F. Except for the portland cement mixture the magnitude of test values is much lower than those indicated in the preceding figures.

Some type of test which approaches the time-temperature relationship shown in Figure 17-B might be valuable used alone or in addition to the present type of test.

General Effect of Each Variable. All general statements result from an analysis of data obtained when using the 1944 model dilatometer. Figure 22 lists the chosen variables and a short general statement of their effects upon the H.C.S. Possibly some exceptions to these statements will become evident later.

New Movie. A motion picture film has been prepared which shows the reaction of test specimens from the eight mixtures of Table 3 upon exposure to shock heating. The compression breaks are also recorded. As yet a complete study of the film has not been made. As an indication of the value of recording such data on film it is obvious from observing the compression breaks that cracking of the specimen surface often begins at the places where the fingers touched the test specimen during handling. Ability to reproduce test values within any particular batch has been bettered by *not touching* the specimen with the fingers.

One method of handling the specimen is as follows. Two curved drying plates (supplied by H. W. Dietert Company) are shortened to 1½ in. in length. After the specimen has been stripped from the tube it is lightly gripped with the two plates and turned to a horizontal position resting upon one of the plates. A 1⅜ in. × 1 in. refractory disc is placed against one end of the test specimen and the specimen and disc then set upon the table in a vertical position. The top concave

disc is placed upon the specimen and the specimen is tested as usual.

Conclusion

Seven variables were chosen to be studied regarding their individual effects upon the H.C.S. The conclusions are summarized in Fig. 22 showing these variables and their effects. The data obtained also permits the following general statements:

1. At a T.C.S. (temperature control setting) of 1600° F. the H.C.S. of the test specimens increases as the time of specimen exposure to heat increases.

2. Test values for all series of tests at a T.C.S. of 2000° F. indicate a smaller magnitude of variation than tests conducted at a T.C.S. of 1600° F. Comparison with test values at a T.C.S. of 2500° F. is perhaps unwise because of the very low H.C.S. values at this T.C.S.

3. At a T.C.S. of 2000° F., fire clay probably develops the greatest H.C.S. after a specimen exposure to heat of 8 min.

4. At a T.C.S. of 2500° F. the H.C.S. is greatest at the 2-min. exposure and at longer exposures the H.C.S. is generally 10 psi. or lower.

5. For 4 per cent additions western bentonite imparts greater H.C.S. to the test specimen that either southern bentonite or fire clay.

6. Silica flour increases the H.C.S.

7. A one per cent addition of corn flour does not significantly effect the H.C.S.

8. At a T.C.S. of 2000° F. certain mixtures (3-4-5-6) indicate a minimum H.C.S. at an exposure time of 8 min.

Better reproducibility of test values have resulted by eliminating the handling of the test specimen with the fingers. A motion picture study of test specimen at a T.C.S. of 2500° F. indicated cracking began first at the place where there were finger marks.

One new kind of testing procedure for elevated-temperature studies is proposed as it may more closely approach the temperature-time relationship in the sand mixture after mold cavity is filled.

DISCUSSION

Chairman: H. F. TAYLOR, Massachusetts Institute of Technology, Cambridge, Mass.

Co-Chairman: J. A. RASSENFOSS, American Steel Foundries, Indiana Harbor, Ind.

H. MEYER[1]: Does the line between the upper and lower part of your specimen occur if you double-end ram or single-end ram the specimen?

MR. WILLIAMS: We double-end rammed the specimen. Theory indicates that the line will occur with single-end ramming. I have not tried single-end ramming as yet until a safety control is installed on the research furnace. Then work will proceed on 2x2-in. specimens. We probably will get the same type of failures within the test specimen.

CO-CHAIRMAN RASSENFOSS: The southern Bentonite specimens you showed appeared to be cracked. Cracking occurred after cooling, did it not?

MR. WILLIAMS: Yes.

CO-CHAIRMAN RASSENFOSS: Were the specimens in one piece in the furnace?

MR. WILLIAMS: No.

CO-CHAIRMAN RASSENFOSS: Did they break apart in the furnace?

MR. WILLIAMS: When we took the specimens out of the furnace, it was immediately evident that they were in at least two parts.

CO-CHAIRMAN RASSENFOSS: But they stayed in one piece in the furnace.

MR. WILLIAMS: They were side by side and cracked while in the furnace. It was possible to see them crack through the furnace observation port. When removing them from the furnace on the disc you carried together both parts or whatever number of parts there were and set them out together on a block to cool. Then you could pull them apart and see the different shapes. They crack in the furnace under shock heating.

CO-CHAIRMAN RASSENFOSS: What was that heating temperature?

MR. WILLIAMS: The temperature of those specimens was 1600° F.

CO-CHAIRMAN RASSENFOSS: Does this cracking persist as furnace temperature is increased? Have you carried out such experiments?

MR. WILLIAMS: Yes. We have conducted most of the tests at 2700° F. The specimen was heated for 20 min. The glassy form of the clay is formed under those conditions. When you bring out the specimen and set it at room temperature there is a possibility of retaining the test specimen. If you just held the specimen (using western or southern Bentonite) at 2000° F., and set it out in the room, in 15 or 20 min. it will disintegrate. So it was necessary to get up to a temperature at which the specimens could be kept together for later examination.

CHAIRMAN TAYLOR: I should think it would be rather dangerous to assume that bad cracks were present during the actual testing just because they were there at room temperature because of the plastic condition of the sand at high temperatures. As the specimen cools down, cracks would tend to form more than they would when hot and would indicate an exaggerated effect.

MR. WILLIAMS: In the high temperature film we have taken at 2500° F., those cracks were evident. You can see them only in the rind, however. It will be only when we can actually make the study at the high temperature that I think we can show that those cracks are present at the beginning of the time the test specimen is loaded. It seems to me that the cracks are in the specimen before the load is applied to the specimen.

MEMBER: If the crack occurred at those temperatures, would the permeability and strength be adversely affected or affected to a great extent? Would permeability and strength really mean anything?

MR. WILLIAMS: If a metal test specimen had already failed before you tested it, you probably would not test it.

MEMBER: Does it have a crack in it before testing?

MR. WILLIAMS: That is right. It has already failed by the time you applied the load to it. I have calculated the hot compression strength on the basis of the effective area. Instead of having hot strength of 200 or 300 lb., we are getting up to 6,000 and 8,000 lb. hot strength because the affected area is greatly reduced. That is just a calculation.

I have found by using the calculations and the nominal values which we get in the furnace that there is a ratio between the hot strength at three rams and six rams. You can figure that if you get a certain hot strength at three rams, at six rams it will be 25 per cent higher. I obtained that figure from recalculating the results of about 500 test specimens. I did the same thing with the green strength on these same mixtures and found that, on the average, if you increase the ramming from three to six rams, you will get a 25 per cent increase in green strength. However, whether there is any relation between the two is yet to be determined.

[1]General Steel Castings Co., Granite City, Ill.

A.F.A. COMMITTEE REPORT ON

MANGANESE BRONZE

This paper discusses hard spots experienced in machining manganese bronzes covered by Federal Specification QQ-B-726c, Grades A, B and C. It also discusses maximum limits of tin and lead in relation to attainment of yield strength and elongation in Grade B alloy.

G. K. Dreher
Olds Alloys Co.
Southgate, Calif.

THIS REPORT is the result of experience of industrial representatives who supplied information and data herein contained on manganese bronze. It is the result of practical shop research as well as laboratory tests. It is generally a consensus of opinion and is of value because it represents war time experience.

The report discusses identification and possible correction of hard spots appearing in high strength manganese bronze castings made according to Federal Specification QQ-B-726c, Grades A, B, and C. Table 1 and 2 show the chemical and physical requirements of this specification. This report also discusses whether the maximum limits of tin and lead as now expressed in the specification are too liberal in relation to attainment of the yield strength and elongation requirements in Grade B alloy.

In gathering and summarizing these data on manganese bronzes the Committee found that wide variations in practice existed in the foundry industry. Within the extremes of composition used, there must be one set of optimum properties and corresponding chemical requirements which would serve as a basis for a more closely controlled and, consequently, better engineering alloy than exist under present broad range of compositions.

The metallurgical experience of the Committee would indicate that in most alloys an extremely wide range will result in performance which will vary according to the composition of the material.

During war-time production several questions arose in connection with Manganese Bronze. These questions (seven in number) were presented to a group of outstanding non-ferrous metallurgists in the country for consideration. Results and answers to these questions are presented in question and answer form.

Hard Spots Experienced

Several manufacturing plants experienced hard spots in manganese bronze alloys which made machining of some castings difficult. One question presented is as follows:

I. Is the presence of these hard spots due to improperly alloyed material, the formation of metallic compounds or oxides during the melting and casting, or to any other cause?

The cause of hard spots, according to the survey, is as follows:

1. Previously contaminated raw materials in the charge are a contributing factor. This can be in the form of scrap or in the choice of rusty iron and/or aluminum oxidized through manufacturing operations or exposure to the elements.

2. Poor melting practice, permitting formation of excessive oxides and overheating of the metal with subsequent improper feeding and secondary shrinkage may also be a cause. This includes unclean crucibles and furnaces.

3. Turbulence due to improper gating procedure permitting the formation of oxide, and

4. Lack of care in the original alloying of the ingot melt may also be causes.

Remedy for Hard Spots

II. What remedy can be offered for the solution of these hard spots?

From the survey made it is evident that certain definite steps can be taken to eliminate these hard spots.

1. High quality back scrap or virgin materials must be selected for the basic metals. This would involve a selection of high quality iron in the purest possible state. Care should be taken to avoid use of alloy irons for this purpose. The Committee recommends that low carbon soft irons be used.

Aluminum, likewise, should be of highest purity and preferably not an alloy. The committee recommends that trimmings and clippings contaminated during manfacturing operations be avoided where possible and that straight high purity aluminum be used.

2. In preparation of the alloy, care should be exercised to avoid moisture-containing back scrap that might have a tendency to assist in creation of added oxides in the resulting alloy.

3. A review of the chemical requirements reveals that several inconsistent results are possible because of the range of aluminum and iron permitted. Pre-alloy,

Presented at a Brass and Bronze Session of the American Foundrymen's Association at the Fiftieth Annual Meeting, May 10, 1946, at Cleveland. It is a report submitted by the Non-ferrous Representative of the Ordnance Research Committee of A.F.A.

uniting the iron with the aluminum and pure copper, is definitely required.

The other alternatives would be to overheat the bath of metal to insure complete solution and/or melting of the iron and aluminum, or to permit the bath to remain molten for a period long enough to insure complete solution.

The use of a pre-alloy would eliminate these two undesirable methods. Inasmuch as iron and aluminum unite to form $FeAl_3$ the composition should be so adjusted to permit absorption of all of the iron in this manner. This compound normally consists of approximately 41 per cent iron and 59 per cent aluminum. In all cases the ratio should therefore be approximately 3 parts aluminum to 2 parts iron, with the aluminum in excess of this ratio at all times.

Present specifications permit free iron to remain outside of the $FeAl_3$ composition in such a manner that it must be absorbed into the copper. Unless this absorption is definitely assured, free iron will be present in the final alloy. By having a copper-aluminum-iron pre-alloy a ready method of introducing these elements into a copper-zinc bath is achieved.

If the above procedure is followed, pouring temperatures can be held to a minimum and the metal will be retained in its best quality through the smelting and alloying processes, and subsequently during the remelt for casting purposes.

4. *Casting Practice.* All castings made in manganese bronze, and particularly in Class B and C, alloys, must be gated so as to avoid turbulence during the pouring operations and so risered as to feed the entire casting soundly to prevent honeycombing or shrink pockets which provide opportunities for oxidation and subsequent difficulty in machining.

Effect of Tin and Lead

III. What effect does the tin and lead content have upon the physical properties of these alloys?

Both tin and lead appear to have deleterious effects upon manganese bronzes in general.

1. Tin and lead in small amounts apparently have no effect in the Class A alloy. A summary of 114 heats totaling approximately 2,500,000 lb. of alloy indicated that no rejections were caused by tin contents ranging from 0 to 0.9 per cent and lead from 0 to 0.3 per cent.

2. Tin definitely affects elongation values of Class B and C manganese bronze. By so reducing the elongation, it is reasonable to expect that increases in tensile strength will occur. This is not always true, however, in that any embrittling of an alloy would tend to create conditions by which the ultimate tensile properties could be caused to fall off through the inability of the alloy to readjust itself under strain, as is the case when a desirable elongation is present. A practical summation of this effect is set forth in Table 3 of Class C alloy heats.

Table 3 shows that the copper, aluminum and tin vary. Iron and manganese were kept constant at 2.60 and 3.75 per cent, respectively. The metallurgists who were contacted unanimously recommended that tin be deleted entirely from the specification or that limitations be placed at 0.02 per cent *maximum.*

3. *Lead.* The presence of lead generally reduces the physical properties and should be maintained to 0.02 per cent maximum with 0.01 per cent or less desired.

Tin and Lead Changes

IV. What changes would you recommend in the tin and lead content in the specification?

1. Opinion states that tin should be specified 0.02 per cent max. for Class B and C alloys.

2. Opinion states that lead should be specified at 0.02 per cent maximum for Class B and C alloys.

V. Have you identified any other minor elements which are deleterious to the quality of this particular alloy and, if so, in what quantities are they harmful?

1. Very little work has been done on this subject although reference has been made to the harmful affects of silicon which should consequently be kept well under 0.05 per cent.

2. The general impression seems to be to eliminate all impurities when dealing with sensitive alloys of this type. These impurities would include such elements as antimony, phosphorus, chrome, cobalt, bismuth, and other metals which might be ordinarily found in a non-ferrous foundry.

3. Specific reference to the harmful effects of phosphorus and calcium on Class B and C alloys were mentioned in replies to the questionnaire.

Table 2

PHYSICAL REQUIREMENTS, FEDERAL SPECIFICATION QQ-B-726c

	Physical Properties		
Class	*Tensile Strength, psi. (min.)*	*Yield Strength,* psi. (min.)*	*Elongation Per cent in 2 in.*
A	65,000	25,000	20
B	90,000	45,000	20
C	110,000	60,000	12
D	60,000	20,000	20

*Yield Strength at 0.2 per cent set.

Elongation and Yield

VI. What is your experience with elongation and yield strength requirements of composition B of this specification?

1. Opinion indicates that the 20 per cent minimum elongation is inconsistent with the nature of this alloy and should be reduced to approximately 18 per cent.

VII. What recommendations would you make in regard to the specification of these alloys from both the chemical and physical standpoint?

Table 1

CHEMICAL REQUIREMENTS, FEDERAL SPECIFICATION QQ-B-726c

	Elements, in Per cent							
Class	*Cu*	*Zn*	*Al*	*Mn*	*Fe*	*Sn (Max.)*	*Pb*	*Ni (Max.)*
A	55-60	Rem.	0.50-1.5	1.50*	0.40-2.0	1.00	0.40*	0.50
B and C	60-68	Rem.	3.0-7.5	2.5-5.0	2.0-4.0	0.50	0.20	—
D	56-62	Rem.	1.50*	1.50*	2.00*	1.50	—	—

*Maximum.

Table 3

PROPERTIES OF CLASS C ALLOY HEATS

Mechanical Properties			Elements, per cent		
Tensile Strength, psi.	Yield Strength, 0.5 per cent	Elongation, per cent	Copper	Aluminum	Tin
116,000	71,000	16.4	61.50	5.50	0.00
40,600	*	0.2	61.50	5.50	0.50
46,000	*	0.2	61.50	5.50	0.90
95,000	49,500	14.3	68.00	5.40	0.00
95,300	56,400	6.0	68.23	5.40	0.56
100,100	50,600	16.3	68.00	5.40	0.00
99,800	55,500	8.9	68.40	5.40	0.56
95,000	*	2.0	64.20	5.50	0.55
106,500	66,800	14.0	64.02	4.66	0.00
95,800	69,000	3.5	64.00	4.60	0.61

*Yield not determined for fear of breaking extensometer due to brittleness of the alloy.

1. We would propose that the chemical requirements of Class B and C alloys be altered to correspond to these shown in Table 4. The values in Table 4 define the composition more accurately and enable a more ready meeting of the physical requirements. This premise has been checked by commercial producers and verified as a practical set of standards to follow for this purpose.

2. Physical properties would be altered to correspond to those shown in Table 5.

Conclusions

I. Raw materials for Class B and C alloys should be specified as follows:

1. Virgin metal, if possible.

2. Clean scrap of known analysis as follows:

a. Low carbon rust and dirt-free iron preferably of the highest purity. (Avoid alloy iron.)

b. Aluminum clippings cleaned and dry. (Avoid alloy aluminum particularly those containing silicon.)

c. Copper scrap cleaned and dry.

d. Back scrap of known purity cleaned and dry.

II. Procedure for manufacturing these alloys should be specified as follows:

1. Proper pre-alloy containing the total amount of aluminum, iron, manganese or alloys plus other minor constituents with some copper should be the first step.

2. The above is to be added to the charge in the alloying melt.

3. Overheating the metal should be avoided in both the alloying and casting stages. A maximum temperature of 2400° F. is recommended.

III. Specify a ratio of aluminum to iron be maintained at 3 minimum to 2 maximum with the excess in favor of aluminum.

IV. Specify for Class B and C alloys that tin, lead, silicon, phosphorus, calcium, chrome, cobalt, antimony and other components be held to an absolute minimum in all classes of manganese bronze.

V. Alter the present chemical requirements of Federal Spec. QQ-B-726c to correspond to those shown in Table 6.

VI. Alter the physical requirements to correspond to those shown in Table 5.

VII. Recommend that castings be produced from previously created ingot metal to insure uniformity.

Table 5

RECOMMENDED MODIFIED PHYSICAL PROPERTIES FOR FEDERAL SPECIFICATION QQ-B-726c

Class	Tensile Strength, psi. (min.)	Yield Strength,* psi. (min.)	Elongation in 2 in., per cent (min.)
A	65,000	25,000	20
B	90,000	45,000	18
C	110,000	60,000	12
D	60,000	20,000	20

*Yield Strength at 0.2 per cent set.

Reporting Personnel

The metallurgists who have conferred in studying and solving these questions were as follows:

J. E. Crown, Master Mechanic, Naval Gun Factory Foundry, U. S. Navy Yard, Washington, D. C.

W. B. George, Metallurgist, R. Lavin & Sons, Inc., Chicago.

Jack Klement, Chief Metallurgist, Ampco Metal, Inc., Milwaukee, Wis.

B. A. Miller, Chief Metallurgist, The Baldwin Locomotive Works, Cramp Brass & Iron Foundries Division, Philadelphia.

Wm. Romanoff, Vice-President and Technical Superintendent, H. Kramer & Co., Chicago.

Roger Keeley, Ajax Co., and George Bradshaw, Philadelphia Navy Yard, concurred with B. A. Miller in the information set forth by him.

Not all of the statements contained herein represent the unanimous opinion of the participants. On most points there is general agreement with specific exceptions which are not identified in the body of the report.

Table 4

PROPOSED MODIFIED CHEMICAL ANALYSIS OF CLASS B AND C ALLOYS

Element	Content, per cent Class B	Class C
Copper	56-66	60-68
Zinc	Bal.	Bal.
Aluminum	2.0-5.0	4.5-7.5
Manganese	2.5-4.0	3.5-5.0
Iron	1.25-3.5	2.0-4.0
Tin	0.02 Max.	0.02 Max.
Lead	0.02 Max.	0.02 Max.

Table 6

RECOMMENDED MODIFIED CHEMICAL REQUIREMENTS OF FEDERAL SPECIFICATION QQ-B-726c

Element	Content, per cent Class A	Class B	Class C
Copper	55-60	56-66	60-68
Zinc	Balance	Balance	Balance
Aluminum	0.50-1.5	2.0-5.0	4.5-7.5
Manganese	1.50	2.5-4.0	3.5-5.0
Iron	0.60-2.0	1.25-3.5	2.0-4.0
Tin (max.)	1.00	0.02	0.02
Lead	0.40	0.02	0.02
Nickel (max.)	0.50	0.3*	0.2*

*Not desired.

References

1. A. H. Hesse, E. To. Myskowski, B. M. Loring, "The Effect of Various Elements on the Mechanical Properties and Dezincification of Manganese Bronze," *Transactions*, AMERICAN FOUNDRYMEN'S ASSOCIATION, vol. 51, p. 821 (1943).

2. G. P. Halliwell, "The Effect of Lead on Some Mechanical Properties of Manganese Bronze," *Transactions*, AMERICAN FOUNDRYMEN'S ASSOCIATION, vol. 51, p. 837 (1943).

3. J. T. Robertson, "The Utilization of Fired Cartridge Brass in Cast Manganese Bronze," *Transactions*, AMERICAN FOUNDRYMEN'S ASSOCIATION, vol. 52, p. 527 (1944).

DISCUSSION

Chairman: G. P. HALLIWELL, H. Kramer & Co., Chicago.

Co-Chairman: A. H. HESSE, R. Lavin & Sons, Inc., Chicago.

E. J. BOYLE[1]: We have done considerable work on alloys of this type in our laboratories during the last few years. Our alloys contain a higher manganese content than manganese bronze. We have found that silicon is probably the most critical element of all. In alloys that contain iron in the amount of 1 per cent, we have found that silicon reduces tensile strength as much as 5,000 psi. if present in amounts as low as 0.1 per cent. I think some work in that direction should be carried on with manganese bronze.

MR. DREHER: You say that is in the presence of iron? That is a very interesting bit of information.

MR. BOYLE: We traced this effect down and actually isolated iron-silicon-manganese compound, and determined their compositions. These compounds were quite hard and could lead to difficulties in machining.

GEORGE DALBEY[2]: In regard to silicon in manganese bronze, a heat of manganese bronze was accidentally contaminated with silicon to the extent of 0.1 of 1 per cent. It was noticed that when standing a sludge formed on top of molten metal when the temperature dropped slightly. We allowed the metal temperature to drop to somewhere near the pouring temperature, collected some of the sludge and found 1.5 per cent silicon and about 9 per cent iron in the sludge. In the liquid manganese bronze, the silicon was a trace and the iron had dropped off to about 0.6 per cent. In other words, there was a concentration of iron and silicon in this sludge. This sludge was composed of massive segregations of blue constituent; very often seen in manganese bronze; entraining considerable manganese bronze. The entrained manganese bronze had a normal alpha-beta structure.

MR. DREHER: I assume the silicon helped free the iron from the alloy and formed a separate lump.

MR. DALBEY: Yes, apparently it did.

MR. BOYLE: We ran segregation tests on alloys containing iron and silicon and found that we got a segregation of this manganese-iron-silicon compound at the top of the melt when we had appreciable silicon present.

MR. DREHER: Would you assume that possibly the iron-silicon compound would be a difficult, brittle, metal to handle?

MR. BOYLE: Yes, it is.

MEMBER: I missed the question about these hard spots, the difficulty of machining, and what causes them.

MR. DREHER: There is no direct answer except that free iron might have been present.

MEMBER: Were they real or imaginary? Was it established that there were hard spots?

MR. DREHER: It has definitely been established that there was difficulty in machining.

G. M. THRASHER[3]: I have seen such specimens taken from small bushings with hard spots that were cast from this 100,000-psi. tensile grade of manganese bronze. The spot always seemed to occur at the same place, near the gate. I could not understand why we had so many always at that one spot. Why did they not circulate to some other spot in the casting?

MR. DREHER: Has anyone made any investigation of the chemical composition of the casting versus the gate or riser?

MR. DALBEY: I wonder if the hard spots could not be accounted for by the sludge that forms on slightly cooling with a segregation of the iron and silicon at these points? I noticed the sludge did not move very well. It would get into some corner and stay there, and the manganese bronze would run on through and around it.

G. P. HALLIWELL[4]: I would like to corroborate the effect of silicon. Dr. H. W. Gillett gave a paper entitled "Role of Silicon in Non-Ferrous Castings" which appears in TRANSACTIONS, American Foundrymen's Association, vol. 46, p. 413 (1938). In a discussion of that paper, I presented several photomicrographs of manganese bronze of practically the same composition to which varying amounts of silicon had been added.

The high zinc equivalence of silicon is seen in the change from a 50 per cent alpha-beta structure in an alloy of no silicon to a completely beta structure in a similar alloy containing 0.6 per cent silicon. Whether or not the needle-like formation seen in these alloys is free silicon or iron-silicide, I do not know. Certainly they do not have the crystalline habit or color of the usual iron-rich constituent seen in manganese bronzes.

The hard spots found in manganese bronze may be two things; i.e., (1) an iron-silicon slag resulting from combination of silica in the crucible with iron oxidized from the alloy; (2) iron partially alloyed with the base metal, and generally results from an initial imperfect alloying of the iron.

MR. DALBEY: At one time we received several carloads of ingot and noticed with some we had trouble with meeting specifications and with others we did not. We took cross sections of different ingots and roughly ground them down and then etched them in ammonium chloride solution. In some of these ingots, rust clusters would show on the surface of the sections and iron segregations were found in these locations in the ingots. In other ingots the surface of the sections would stay perfectly clear. This is one way to spot this unalloyed segregated iron rather easily.

MR. HALLIWELL: Are they both magnetic?

MR. DALBEY: Yes.

MR. BOYLE: I might say in the identification of these compounds, that we separated them by dissolving the alloy in concentrated nitric acid and obtained a residue which we found to be high in manganese, iron, and silicon.

MR. HALLIWELL: Does that distinguish between the usual blue color of the normal iron constituent present in manganese bronze? We do not know what that composition is. Does it distinguish between that and the material you are thinking about?

MR. BOYLE: I do not know the answer to that.

MR. HALLIWELL: The yield strength at 0.2 per cent permanent set is more difficult to obtain because it is necessary to apply and release the load several times before the corresponding stress is obtained. There is always the danger of over-shooting the required stress.

If the yield is obtained by the 0.2 per cent offset, then a stress-strain curve must be obtained or plotted. On low strength alloys the 0.5 per cent yield under load gives approximately a 0.2 per cent permanent set, which will vary with composition and resulting tensile properties. The 0.5 per cent yield under load, however, is a definite property, simple to obtain, requires no stress-strain curve, and eliminates most of the personal equation.

On some high strength manganese bronzes the 0.2 per cent yield by offset is higher than the yield at 0.5 per cent extension under load.

MR. DREHER: There is a difference also in the yield strength that you achieve by the two methods.

MR. HALLIWELL: Very much so.

MR. DREHER: And I imagine it is higher by the method expressed here.

MEMBER: I believe that is an extension under load, 0.2 per cent plus the arbitrary extension. I think on the C Grade, from memory it is something like 0.0063 instead of 0.005.

MR. HALLIWELL: Is that 0.2 per cent under load?

MR. DREHER: They have calculated from their experience that with an extension of 0.0062 in., upon releasing the load the test bar will take a 0.2 per cent set.

MR. HALLIWELL: What alloys were you referring to?

MR. DREHER: That is the 110-C Classification.

MR. DREHER: May I quote from the reference I have? This is by William Romanoff of Kramer & Co. "It is proved by tests here that you would have considerable difficulty meeting the yield strength on the Grade B if the offset

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[1]The New Jersey Zinc Co., Palmerton, Pa.
[2]U. S. Navy, Mare Island Navy Yard, Vallejo, Calif.
[3]R. Lavin & Sons, Inc., Elmira, N. Y.
[4]H. Kramer & Co., Chicago.
[5]Bureau of Ships, Washington, D. C.

method were used. The extension under load method which paragraph 3 permits gives about 5,000 lb. higher yield strength. On the Grade C, the 0.2 per cent offset conversely gives about 3,000 lb. higher yield strength. In other words, the definition for yield strength should be as suggested above." That is the reason given there. I think that is another point that should be worked on.

MR. HALLIWELL: It should be standardized.

LT. COMDR. B. O. BROUK[5]: I was particularly interested in the discussion on variation of specifications. During the war, we were particularly concerned about variance in the spread of both the Federal specifications and Navy specifications. One of the broadest specifications, I think, is 49-B-3, our regular manganese bronze with 65,000 psi. tensile. We did quite a bit of work on salt water corrosion on this composition and we came to the conclusion that we can reach two definite phases in the alloy.

By that I mean, we can get the alpha-beta microstructure and we can get a straight beta. We found that the beta is definitely detrimental as far as any salt water stress corrosion or fatigue is concerned and under light load we have had definitely a few failures. We did a lot of research work at the Portsmouth Navy Yard on failures in some of our applications and we traced them down to a theory that has not been given much consideration before, and that is with regard to the aluminum-zinc ratio. In other words, we feel that the power of inoculation with aluminum is about six to one, with zinc remaining constant.

We have found that you could get it with approximately 39 per cent zinc, copper at 58 per cent, aluminum content held at 0.8 per cent, manganese held definitely at 1 per cent, and iron not over 1.2 per cent. We could get between 42 to 45.5 per cent of zinc plus aluminum with this composition. We could get a definite alpha-beta microstructure with this analysis. With the alpha-beta microstructure we get a particular type of grain formation whereby we do not get the stress corrosion failure that we do on the beta microstructure. The physical properties on the beta microstructure are definitely higher than on the alpha-beta microstructure.

That would be of interest to some of the ingot manufacturers. You could produce an ingot of the beta microstructure and meet the physical properties on the 46-B-25 ingot-bronze specification very easily. But in our application, after we applied it to our physical properties, regardless of the fact they passed the specifications, they will not give the salt water corrosition resistance that the Navy requires. The work to be done further is possible modification of the specification. The Navy is working on that and I think we have definitely come to the conclusion that we want alpha-beta microstructure.

MR. DREHER *(author's closure)*: The author believes that this report, and more especially the discussion growing out of it, has opened the way toward innumerable necessary investigations which can be accomplished in industrial, professional or collegiate laboratories. The unknown factors in the control of Manganese Bronze are far too many for a series of alloys which have found such widespread use.

ECONOMICS OF CUPOLA OPERATIONAL CONTROL

F. K. Vial
Vice President
Griffin Wheel Co.
Chicago

The author stresses the importance of cupola operational control. A scientific method of analysis of heat developed and heat distribution in the cupola is discussed. Much useful data is presented in table and chart form for reference.

EVERY OWNER or operator of a cupola should know to what extent he is securing the full value of the fuel he is using.

A publication for the cupola operator should contain not only the physical constants and conversion relationships necessary for the calculation of problems dealing with the development and absorption of heat, but also detailed tables of thermal balances over a wide range of coke ratios and of air volumes indicating the weight and volume of the air required for combustion.

It should also contain the weight and volume of the CO_2, CO and nitrogen in the resultant gases for twenty different stages of combustion; the Btu. developed for each stage, with corresponding flame temperatures; the rate of combustion, and rate of melting.

It should further contain Btu. absorbed by the iron; Btu. used in the formation and superheating of slag; Btu. lost in converting CO_2 to CO; Btu. remaining in effluent gases; and finally the efficiency of absorption by the iron, efficiency of combustion, and the overall thermal efficiency.

Cupola Operation Tables

Such a compilation would contain the specific heats of the various gases of combustion from zero to 4000° F.; the specific heats of carbon, slag, iron, etc., through the entire range from zero to 3000° F., the heat content in preheated blast from 100 to 1500° F., sufficient for the combustion of 1 lb. of coke at various stages of combustion.

These tables, to be of value, must be so complete that any desired quantity can be taken directly from them without calculation. It is the author's purpose to present a preliminary group of tables of this character, to be used until something better is furnished.

Figure 1 and 10 are presented to show the tendencies of variation in coke ratios and air volumes as they relate to temperature of the iron melted, rate of melting, stage of combustion, etc. These charts have not been verified in all their extremities, hence they are only intended to indicate tendencies rather than exact quantities.

Heat Balances

Table 1 is a summary of heat balances for the entire range of cold blast cupola practice; a companion table for pre-heated blast is in preparation.

Most publications on cupola practice are subject to the criticism that they are too general in their application to solve definite problems. It is therefore the intention in this discussion to be very definite even at the expense, in certain cases, of being only approximately correct.

The principles of regeneration indicate the possibility of the reduction of coke consumption by 30 per cent as compared with ordinary cold blast practice, or the possibility of increasing the iron charge by 50 per cent without increasing the coke charge.

The objective is always to produce iron of the highest quality, at the least expense. There is no thermal reason why 16 lb. of iron should not be melted, at 2700° F., for each pound of coke used.

The heat developed in the Combustion Zone must be considered from two standpoints:

First, the high temperature portion suitable for melting iron; and,

Second, the remaining portion having a temperature below 2500° F. that can be used only for preheating purposes.

In order to obtain higher efficiencies it is necessary to recover the excess low temperature heat after it has served its purpose for preheating and carry it to the high temperature side of the equation, and thus balance the heat required for both purposes.

The reputation of the cupola for economy comes about by comparison with other types of melting having lower efficiencies. Among them are the following:

Melting Method	*Efficiency Per Cent*
Crucibles	3 to 6
Air Furnace	10 to 15
Regenerative Open Hearth	20 to 30
Cupolas	30 to 50

It is only relatively, however, that the cupola can be said to be economical. Every operator witnesses daily 50 per cent of the potential heat of the coke burst into flame and rush upward through the stack, to be lost in the atmosphere, having served no useful purpose. The operator is not asked to make an accounting of what portion of the potential heat he used and what he did not use.

This condition does not indicate a high appreciation of our responsibility toward the conservation of Nature's fuel resources.

It is the purpose of this discussion to point out the character and origin of this waste, with suggestions for improving the thermal balance sheet consistent with smooth operating conditions.

The present varying operating

conditions are best shown by reviewing the results of 5000 cupolas operating daily in the United States and Canada, melting more than 20,000,000 tons of iron annually.

Variations in Details of Operation

There are wide variations in details of operation of individual cupolas, even though they may be turning out identical products. The most important of these variations for which, in the main, there is no apparent reason, and which are reflected in the wide differences in the cost of melting, are:

Pounds of iron melted per lb. of coke	7 to 15 lb.
Coke for melting one ton of iron	150 to 300 lb.
Rate of burning coke per sq. ft. of cupola area per hr.	120 to 180 lb.
Rate of melting per sq. ft. of area per hr.	800 to 1800 lb.
Temperature of iron as tapped	2550 to 2850°F.
Efficiency of combustion	50 to 70 per cent
Efficiency of heat absorption	55 to 70 per cent
Overall efficiency	30 to 45 per cent

While the above items which relate to combustion are the most important from the standpoint of cost, yet there are many other items, though less important, which vary over a wide range and in the aggregate constitute a considerable item of expense. Among these are:

Item	*Per Net Ton Melted*
Brick for lining and patching—	using .80 to 2.00 brick
Molding sand—	using 1200 to 3200 lb. per day
Limestone—	using 12 to 25 lb.
Fireclay—	using 2½ to 8 lb.
Firesand—	costing 2 to 10 cents
Labor, Charging Room—	costing $.50 to $1.00
Gannister—	

Further, there is wide variation in the method of making up the cupola. This includes:

Chipping and patching the cupola;
Placing the sand bottom;
Placing and lighting the bed coke;
Size of the charges;
Method of tapping and slagging;
Method of handling materials;

all of which indicates that cupola operations are not standardized at the present time and the costs of delivering a ton of molten iron to the foundry vary through a wide range, with no analysis of a heat balance sheet.

Graphical Illustration of Variations in Practice. In order to visualize the final result of these variations, Fig. 1 has been prepared to indicate the range of operation in ordinary cupola practice and, incidentally, to indicate the trend in the relationship that exists between coke ratios, air volumes, temperature of iron, rate of melting, etc.

The vertical scale indicates the temperature of the iron; the horizontal scale is the rate of melting. The curves bearing upward from the horizontal indicate the lb. of iron melted per lb. of coke, 7 to 11 lb. The vertical curves indicate the air volumes, 150 to 350 cu. ft. per sq. ft. of cupola area per min. In a 60-in. cupola, having 20 sq. ft. of area, these volumes become 3000 to 7000 cu. ft. per min.

Curves are also shown to indicate the relation of percentage of CO_2 to air and coke volumes, blast pressure in oz. and lb. of coke per ton of iron melted.

This indicates the wide range of choice which the cupola operator has in producing ordinary results. If the highest temperatures of iron are required, then the higher amounts of coke and higher air volumes must be used.

For ordinary purposes where 2700° F. is ample, a choice may be made of varying the coke per ton of iron melted per hr. from 182 to 286 lb. This is a difference of 100 lb. per ton melted, producing the same result as far as the melted iron is concerned, if the volume of air is in inverse proportion to the coke.

This represents a variation of $.75 per net ton melted for fuel alone.

Figure 1 is introduced simply to show the trends in varying coke ratios and air volumes. These are summarized in Table 1.

Effects of Variable Conditions on Fuel Economy

The effect of these variations on the economy of melting can best be illustrated by referring them to a large industry in which forty cupolas are melting 1,500,000 tons of iron ore per annum, the product being identical in each case.

The author refers to the chilled wheel industry, members of which are united by an Association having an engineer of tests who, with his staff of inspectors, reviews the product of each foundry daily, rejecting such of the product as fails to meet chill requirements. Each month a detailed report is made showing, among other things, the coke ratios and percentages of product that is rejected.

Coke used varies 150 lb. per ton melted which, at $16.00 per ton, in-

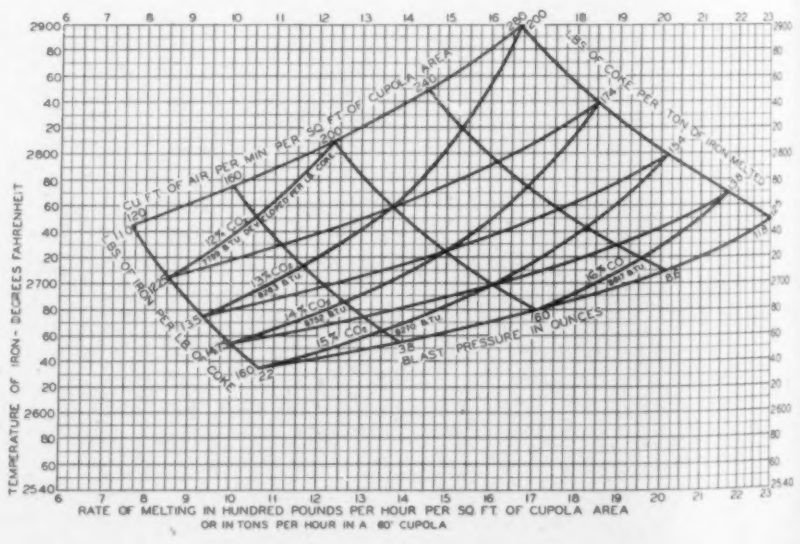


Fig. 1—Usual range of variations in cupola practice with blast at atmospheric temperature. Chart shows the relation of iron to coke ratios, varying blast pressures, and air volumes to rate of melting, temperature of the iron, efficiency and rate of combustion.

Table 1

EFFECT OF CHANGES IN AIR VOLUME AND COKE RATIO

Blast Volume / Coke, lb. per ton	*Increase* / *Constant*	*Constant* / *Increase*	*Increase* / *Increase*
Rate of Combustion	Increases	Increases	Increases
Stage of Combustion (CO_2, per cent)	Increases	Decreases	Constant
Temperature of Combustion Zone	Constant	Constant	Constant
Depth of Combustion Zone	Increases	Constant	Increases
Depth of Reducing Zone	Decreases	Increases	Increases
Temperature of Reducing Zone	Increases	Decreases	Constant
Heat Developed per lb. of Coke	Increases	Decreases	Constant
Rate of Melting	Increases	Decreases	Increases
Temperature of Iron	Increases	Increases	Increases
Loss of Heat (CO_2 to CO)	Decreases	Increases	Constant
Temperature of Stack Gases	Increases	Decreases	Constant
Amount of Slag per lb. of Coke	Increases	Increases	Increases
Temperature of Slag	Increases	Increases	Increases
Humidity Loss	Increases	Decreases	Increases
Combustion Efficiency	Increases	Decreases	Decreases
Absorption Efficiency	Decreases	Decreases	Decreases
Overall Thermal Efficiency	Increases	Decreases	Decreases

cluding handling charge, amounts to $1.20. The loss for failure to control properly depth of chill is from zero to 8 per cent. One per cent loss amounts to 15 cents per ton melted; 8 per cent therefore represents $1.20.

The two items of coke and rejected product may amount to $2.40 which, if applied to 1,500,000 tons annually, equals $3,600,000 hanging in the balance. Not all cupolas can cut their costs $2.40 per ton, but some can; and all can do better than they are now doing.

The remarkable feature of this analysis is that the cupola operator using the least coke for melting usually has the best record for chill control. This, however, does not necessarily indicate the relation between the amount of coke used and cupola regulation, for it is the most careful operator having the best housekeeping that has the lowest cost and most satisfactory product.

Importance of Operational Control

What is true in the car wheel industry is doubtless true throughout the entire foundry industry. Failure to recognize the vital elements involved in cupola operational control costs the industry millions of dollars each year.

At $1.00 per ton, the annual excess cost would be $20,000,000. Hence, it is of supreme importance to develop this phase of cupola education which involves the responsibility of management, the responsibility of the metallurgical staff, and the responsibility of the cupola operator. Here again not all cupolas can save $1.00 per ton, but some can reduce their costs much more than $1.00 per ton melted.

It is often said that we have no desire to save coke. This means that there is no desire to sacrifice quality for an apparent saving in fuel. The author is sure, however, that no one wishes to throw away coke, limestone, silicon, manganese or any other foundry material after the requirements for that material have been satisfied.

It is surprising how many cupola operators fork the coke into the cupola, thus having no accurate record of coke ratios. Many other items of a like nature might be mentioned. There has been entirely too much superstition and individual preference in the operation of cupolas instead of the use of well-established basic principles.

J. E. Hurst, in his book *Melting Iron in the Cupola,* published in England in 1929, said: "The fact that coke charges in many foundries still are left to the personal judgment of the cupola attendant raised the question at once, 'Are our coke ratios too high?' Conceivably even today, careful analysis of normal cupola practice will be repaid by a still further saving in coke."

It is quite evident from the above observations that notwithstanding the great economic importance of the operational control of cupolas, involving millions of dollars each year, there is a wide variation in the operating conditions with respect to the use of materials per ton melted, which indicates a decided absence of standardization based upon scientific analysis of actual requirements.

The question then arises: What is the basis of cupola control, and how can the scientific analysis of actual requirements be established?

Scientific Method of Analysis

The scientific approach to cupola problems is based upon a careful observation of facts developed by experimentation under controlled conditions, and is concerned only with the reproducible facts which can be observed and verified by any properly trained person.

It is the classification and coordination of such facts that determine the behavior of certain combinations and interpret natural laws. It is the development of this system of scientific methods that is responsible for the enormous rate at which the secrets of nature have been unravelled during the past 150 years, with its tremendous impetus to industrial development.

The objective of the application of the scientific method to cupola control is to acquire complete knowledge of the reactions which take place in the various zones of the cupola which point out limitations and bottlenecks in the efficiency of combustion, the efficiency of heat transfer, and finally the overall thermal efficiency of the cupola for any condition under which the cupola is operating.

Tabulation of Factors Relating to Operational Control

If we are to follow the scientific method, it is essential that each fact established in its relationship to other facts be recorded and tabulated for guidance in the solution of problems with respect to all factors involved, making it unnecessary to make calculations for any special operating condition.

These tables should be as complete and as voluminous for the solution of cupola problems as tables of logarithms are for obtaining the exponent that corresponds with any desired number or vice versa.

It would be rather awkward if we were compelled to stop and calculate logarithms, sines, cosines, tangents, etc., every time we wished to use them. In fact, it would not be

done, for in the first place we do not know how, and, secondly, it would take too long. And for the same reason we do not calculate heat balances. We are not familiar with the method, and the operation is too time-consuming.

What we require is reference tables giving complete details of the results from all combinations of coke ratios and air volumes.

Samples of such tables are shown as an appendix to this paper, consisting of fundamentals required in the analysis of cupola thermal problems, with examples of the application of the fundamental to the development of heat balances for specific cases which are shown in two ways: first, on the basis of 1 lb. of coke consumed, and second, on the basis of 1 lb. of iron melted.

Cupola Efficiency

A study of heat balances is the only method of bringing a cupola up to its maximum thermal efficiency, which is usually the condition in any melting furnace for most satisfactory melting conditions. The elements entering into heat balances and which are governed by operational control are:

Ratio CO_2 to CO which underlies all efficiencies. This ratio depends upon velocity of gases through the cupola.

Velocity depends upon vol. per min.

Volume is regulated by blast pressure.

Synchronization of time of melting a charge of iron and time of burning a charge of coke is essential.

Height of bed coke and velocity of gases regulate synchronization of melting and burning.

Efficiency of combustion depends upon percentage of coke burned and percentage not burned.

Heat developed by combustion is a variable.

Sensible heat carried to combustion by superheated coke.

Distribution of heat through the various zones of the cupola.

Excess heat in preheating zone.

Bottleneck in melting zone.

Percentage of heat below 2500° F. not available for melting.

Weight and temperature of gases at various levels of the cupola.

Waste in carbon not burned.

Waste in sensible heat through stack.

The above are the items which control efficiency of combustion, efficiency of heat absorption, overall cupola efficiency, and point to ways and means for a much better heat distribution.

Zones of the Cupola

For the purpose of showing how these items are related one to another, and the method by which they may be controlled, it is necessary to dissect the cupola and separate it into its various zones, each of which has its separate function to perform; the sum total of which constitutes cupola control.

Beginning at the top of the cupola, these zones are:

The Stack, which carries the effluent gases from the cupola to the atmosphere, representing a large waste of heat.

Preheating Zone, which may be from 6 to 10 ft. in height, extending from the charging door to the melting zone. It is here that the melting stock is heated from the atmospheric temperature to 2000° F.

Melting Zone, which lies immediately below the preheating zone and has for its function the changing of iron from the solid to the liquid state.

Reducing Zone, sometimes called the protective zone, which is just below the melting zone and which has a depth of from 15 to 30 in., filled with coke, at a temperature of 3000° F. or more. Its function is to protect the iron from oxidizing influences and to regulate the efficiency of combustion.

Combustion Zone. The layer of coke lying between the top of tuyeres and continuous with the reducing zone is the combustion zone. As its name indicates, this zone of coke from 6 to 12 in. in thickness is the point where the air is introduced and combustion takes place, eliminating all the free oxygen from the gases and producing heat in sufficient quantity for the requirements of each zone of the cupola.

Tuyere Zone, which occupies the space from top to bottom of tuyeres.

The Crucible, which occupies the space between the bottom of tuyeres and the sand bottom.

Unit of Heat Measurements

Since heat is the essential agency in melting iron, it is important that a few of the general laws which underlie the development, transformation, distribution and transfer of heat within the cupola should be established.

There is nothing more universally known than that when the temperature of carbonaceous material is raised to the point of ignition, and brought into contact with air, it burns until entirely consumed, during which process a large amount of heat is developed.

In order to measure and compare quantities of heat, it is necessary to have a unit of measure. For this purpose scientists have established British thermal unit, called the Btu. This unit is described as the amount of heat required to raise the temperature of 1 lb. of water 1° F. For foundrymen, however, who are not scientists, it may be easier to visualize as a unit of measure the quantity of heat contained in 1 lb. of liquid iron at 2700° F. This is equivalent to 540 Btu., which is made up of three parts:

1. Preheating to melting temperature	360 Btu.
2. Transforming from the solid to liquid	70 Btu.
3. Superheating	110 Btu.
Total	540 Btu.

Carbon Through Cupola

Passage of 1 lb. Carbon Through Cupola. Now let us follow a lb. of carbon as it passes from the charging door, at atmospheric temperature, downward through the cupola to the combustion zone, where it comes in contact with the air blast and, by the process of burning, is converted into a gas and returns upward through the cupola and is discharged through the stack to the atmosphere. (See Fig. 2.)

Preheating Zone. The first observation is that when first charged into the cupola the carbon, then at atmospheric temperature, is surrounded by upward flowing gases having a temperature of 800° F. more or less. As the charges descend, the gases become hotter and hotter until the top of the melting zone is reached, when the tempera-

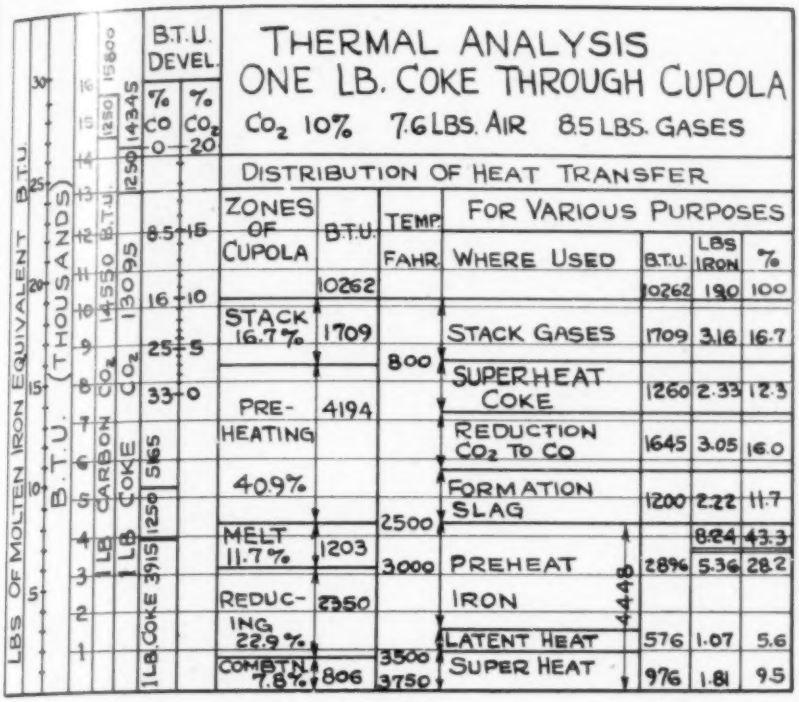


Fig. 2—Thermal analysis of 1 lb. of coke passing through the cupola.

ture of the charges has reached 2000° F. and the gases 2500° F. or more.

Melting Zone. The coke continues its travel through the melting zone, becoming preheated to a higher temperature while the iron is being changed from a solid to a liquid. After passing through this zone the carbon arrives at the reducing zone, where the temperature of the gases is 3000° F.

Reducing Zone. During the passage through this zone each lb. of coke is washed by 10 or 12 lb. of molten iron, in addition to a small quantity of slag, and is surrounded by an atmosphere of CO_2 which has a strong affinity for carbon and in consequence uses part of the carbon to reduce a portion of the CO_2 to CO.

Inasmuch as this reaction is heat consuming the portion of the carbon used for this purpose not only produces no heat but destroys a sizeable quantity of the heat already produced. This reaction will be more fully described later.

Combustion Zone. The carbon that remains after passing through the reducing zone arrives at the combustion zone where the temperature is 3500° F., the carbon having a temperature of 3000° F. or more. At this point air is introduced at atmospheric temperature at the rate of 12.1 lb. of air (151 cu. ft.) to each lb. of carbon burned. The carbon immediately enters into combustion, resulting in the following equation:

12.1 lb. air + 1.0 lb. carbon = 3.68 lb. CO_2 + 9.42 lb. nitrogen.

Total gas, 13.1 lb.

NOTE: In this calculation the 0.16 lb. of excess oxygen shown in Table 24 of the Appendix has been neglected.

Before following the carbon through the process of combustion and the subsequent distribution of heat to the various zones of the cupola, a diversion will be made to explain the amount of heat developed by combustion, the per cent of coke burned and not burned, and the establishment of the CO_2:CO ratio. See p. 519 for continuation.

Heat Developed from 1 lb. Carbon. During the process of combustion 1 lb. of carbon develops 14,550 Btu. The pound of carbon, however, at the time of combustion had a temperature of 3000° F., and since carbon has a mean specific heat of approximately 0.42 it would contain 1260 Btu.

The heat developed in the combustion zone from each pound of carbon burned, therefore, is:

By combustion	14,550 Btu.
Sensible heat	1,260 Btu.
Total	15,810 Btu.

Temperature of Gases. Since the specific heat of the gases at 4000° F. is approximately 0.31, the temperature of the gases becomes

$$\frac{15{,}810}{13.1 \times .31} = \text{(Approx.) } 3900° \text{ F.}$$

The gases, however, do not have this temperature because, as before stated, each pound of carbon has 10 or 12 lb. of molten iron passing over it, and also the ash in the case of coke is converted into slag. These two items represent a temperature loss of about 400° F., so that the gases as they leave the combustion zone have a temperature of approximately 3500° F.

At this temperature, and in the presence of carbon, CO_2 is very unstable, and if it remains in contact with incandescent carbon for a sufficient length of time each atom of oxygen will pick up an additional atom of carbon, resulting in the formula

$$CO_2 + C = 2CO;$$

in which all the CO_2 is reduced to CO.

It is evident that one half the carbon of the resultant $2CO$ was obtained from the CO_2 and one half from some other source, hence the following equation can be written:
3.68 lb. CO_2 + 1 lb. carbon + 9.42 lb. N = 4.68 lb. CO + 9.42 lb. N = 14.1 lb.

and

4.68 lb. CO + 9.42 lb. N = 2 lb. carbon + 2.68 lb. oxygen + 9.42 lb. N = 14.1 lb.

Heat Developed by 2 lb. Carbon. The heat developed in the combustion zone by 1 lb. of carbon when completely burned to CO_2 is 15,810 Btu.

The reduction of 1 lb. of carbon in the form of CO_2 is attended with a loss of 5850 Btu.

In the case of the cupola, the carbon used for the reduction has a sensible heat content of 1260 Btu. per lb. of carbon which, deducted from the 5840 Btu., reduces the net loss to 4590 Btu.

The net result for the 2 lb. of

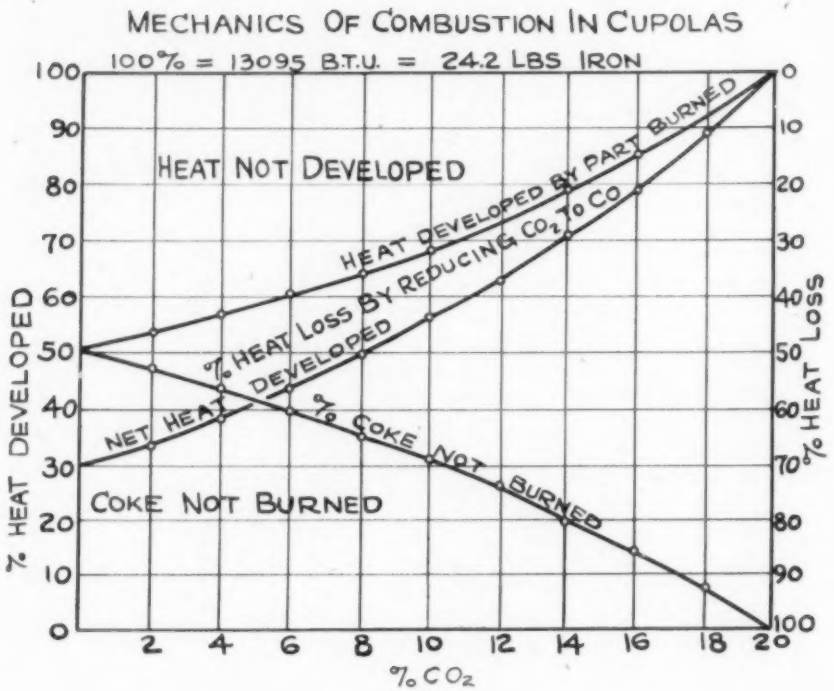


Fig. 3—Diagram illustrates mechanics of combustion in cupolas.

carbon which have disappeared is the following:

Complete Combustion—	
1 lb. carbon	15,810 Btu.
Incomplete Combustion—	
Heat Loss, additional lb. carbon	4,590 Btu.
Net heat in 2 lb. carbon	11,220 Btu.
Heat developed, 1 lb. carbon	5,610 Btu.

This indicates the exceedingly rapid loss of heat from the development of CO, consequently the objective is to control the reduction of CO_2 to a minimum consistent with minimum oxidation of silicon, manganese, iron, etc.

There are several reasons why the combustion within the cupola cannot be complete. In fact, we do not desire complete combustion, for in that event there would be a very heavy oxidizing atmosphere in the melting zone which would use up a large percentage of the alloys silicon and manganese, besides burning the iron.

If the coke were burned to CO, or ordinary commercial gas, there would not be sufficient heat to raise the temperature of the gases high enough to melt the iron, and the cupola would freeze up.

It is therefore self-evident that of necessity the reduction of CO_2 to CO must be partial, in which case a certain percentage of the theoretical heat in the coke is sacrificed to protect the melting zone from too great an oxidation. There is some relation between cu. ft. of CO_2 and the pounds of iron melted that must be observed, and this relationship varies with the stage of combustion.

Stage of Combustion. In complete combustion the volume of CO_2 is 20 per cent of the resultant gases, with no CO, in which case all the coke is burned in the combustion zone whereas in incomplete combustion the CO represents 33⅓ per cent of the total gas with no CO_2, in which case only one half the coke is burned and the other half is picked up in the reducing zone and develops no heat.

No cupola is operated at either of these extremes. Every cupola uses a stage of combustion in which both CO_2 and CO are present in varying proportions and the percentage of coke that reaches the combustion zone, and the percentage used before reaching the combustion zone depends on the ratio of CO_2 to CO, as shown in Tables 2 and 3.

Ratio of Coke Burned to Coke Charged

Tables 2 and 3 indicate that there is a wide range in the net amount of heat that may be developed from each pound of coke charged into the cupola, and it has also been shown in Fig. 1 that the number of pounds and temperature of the iron melted by each pound of coke likewise covers a wide range.

The principal item regulating the amount of heat developed in proportion to the rate at which coke disappears is the ratio of the coke that is burned to the amount which disappears in the reduction of CO_2 to CO. This ratio is fixed by the velocity at which the gases pass through the cupola, which in turn is fixed by the volume of gases per min. These ratios for various stages of combustion are shown in Fig. 3.

The lower curve shows the per cent of coke not burned, which is entirely lost as far as its heat producing effect is concerned. If no further loss were sustained, the heat developed by the portion burned

Table 2

RELATION OF PER CENT CO_2 TO PER CENT OF COKE BURNED AND NOT BURNED*

CO_2, per cent	Coke Not Burned, per cent	Coke Burned, per cent	Btu. Developed in Combustion Zone	Equivalent Lb. Iron	Btu. Loss: Coke Not Burned	Equivalent Lb. Iron	Reduction CO_2 to CO	Equivalent Lb. Iron	Net Btu. Developed	Equivalent Lb. Iron	Efficiency, per cent
20.0	0	100	13,095	24.2	0	0	0	0	13,095	24.2	100.0
14.0+	20	80	10,476	19.4	2,619	4.8	1,096	2.0	9,380	17.4	71.7
12.5	25	75	9,821	18.2	3,274	6.0	1,316	2.5	8,505	15.7	65.0
9.0+	33	67	8,730	16.2	4,365	8.0	1,775	3.3	6,955	12.9	53.5
0.0	50	50	6,548	12.1	6,547	12.1	2,632	4.9	3,915	7.2	29.9

*Portion not burned used in reducing CO_2 to CO.

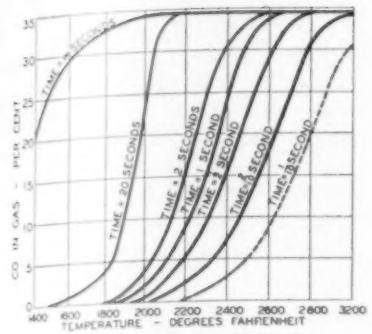


Fig. 4—Effect of time of contact and temperatures on production of CO from CO_2 and coke. (After Haslam and Russell.)

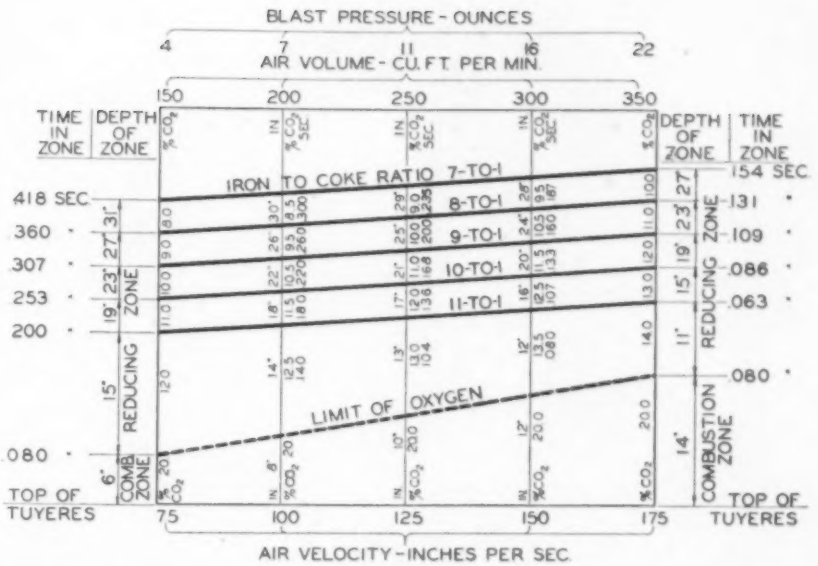


Fig. 5—Height of zones and time for reactions in the bed coke of a cupola.

would be shown by the upper curve.

However, the process of taking up coke for the reduction of CO_2 to CO is a heat consuming process which is represented by the difference between the curve marked "Heat Developed" and the curve marked "Net Heat Developed." The percentage of heat in the net heat developed curve multiplied by 13,095 Btu. will produce the net heat developed shown in Tables 2 and 3.

Inspection of Tables 2 and 3 and Fig. 3 indicates that insofar as economy of fuel is concerned, a very important item is the regulation of the ratio of the CO_2 to CO; and this is the function of the cupola operator.

Time and Velocity of Gases in Preheating Zone. The basis of regulation is the control of the ratio of CO_2 and CO. This ratio depends upon the depth and temperature of coke in the reduction zone and the velocity with which the gases pass through this zone. Velocity depends upon the volume per minute of the gases passing through the cupola.

Depth of the coke in the reducing zone in connection with the velocity of the gases may be expressed in terms of time. The effect of the time that the CO_2 gas remains in the reducing zone has been worked out thoroughly by the Bureau of Mines at Washington, D. C.

It has been placed in the form of a graphical diagram by Haslam and Russell of the Massachusetts Institute of Technology. See Fig. 4. The per cent of CO gas developed is shown by the vertical scale and the temperature of the coke by the horizontal scale, and the heavy vertical lines indicate the length of time the CO_2 is in contact with incandescent coke.

Figure 4 indicates that as the temperature drops the time to make the definite reduction of CO_2 to CO is greatly increased, until 1600° F. it would acquire an infinite time to reduce all of the CO_2 to CO.

In the cupola the temperature just below the melting zone is from 2800 to 3000° F. The dotted line which has been placed on the chart is an extrapolation to indicate that to obtain 20 per cent of CO from CO_2 would require one-tenth of a second in contact with coke at 2800° F.

This indicates the vital importance of the relation of the height of bed coke to air pressure which regulates the air volume and which in turn determines the velocity of the gases through the cupola. The height of the bed coke is automatically regulated to synchronize the time of melting a charge of iron to the exact time required to burn a charge of coke. The ratio of CO_2 to CO is the governor which maintains this balance.

Applying these conclusions to the conditions existing in the bed coke within the cupola will show how regulation is accomplished.

Synchronization of Rate of Melting and Rate of Burning. One pound of coke may furnish sufficient heat to melt 7, 8, 9, 10, 11, 12, 13, 14, 15 or 16 lb. of iron. In each case the one lb. of coke must be burned in exactly the same time required to melt the charge of iron. Also, the one lb. of coke must develop the

Table 3

1 Lb. Coke Burned to Various Combinations of CO_2 and CO

CO_2, per cent	CO, per cent	Nitrogen, per cent	Net Btu. Developed	Equivalent Lb. Melted Iron	Per Cent of Potential	Not Developed—Btu.	Not Developed—Iron, lb.	Not Developed—Per Cent	Air, lb.	Gases, lb.
0.0	33.3	66.6	3915	7.2	29.9	9180	17.0	70.1	5.52	6.42
8.0	20.0	72.0	6538	12.1	49.9	6557	12.1	50.1	7.09	7.99
9.0	18.3	72.7	6938	12.8	53.0	6157	11.4	47.0	7.33	8.23
10.0	16.7	73.3	7358	13.6	56.2	5737	10.6	43.8	7.58	8.48
11.0	15.0	74.0	7799	14.4	59.6	5296	9.8	40.4	7.85	8.75
12.0	13.3	74.7	8263	15.3	63.1	4832	8.9	36.9	8.13	9.03
13.0	11.7	75.3	8753	16.2	66.8	4342	8.0	33.2	8.42	9.32
14.0	10.0	76.0	9270	17.2	70.8	3825	7.0	29.2	8.73	9.63
15.0	8.3	76.7	9816	18.2	75.0	3279	6.0	25.0	9.06	9.96
20.0	0.0	80.0	13095	24.2	100.0	0	0.0	0.0	11.03	11.93

1 lb. carbon + 12.26 lb. air develops 14,550 Btu., equivalent to heat contained in 26.9 lb. molten iron at 2700° F.
1 lb. coke = 0.90 lb. carbon.

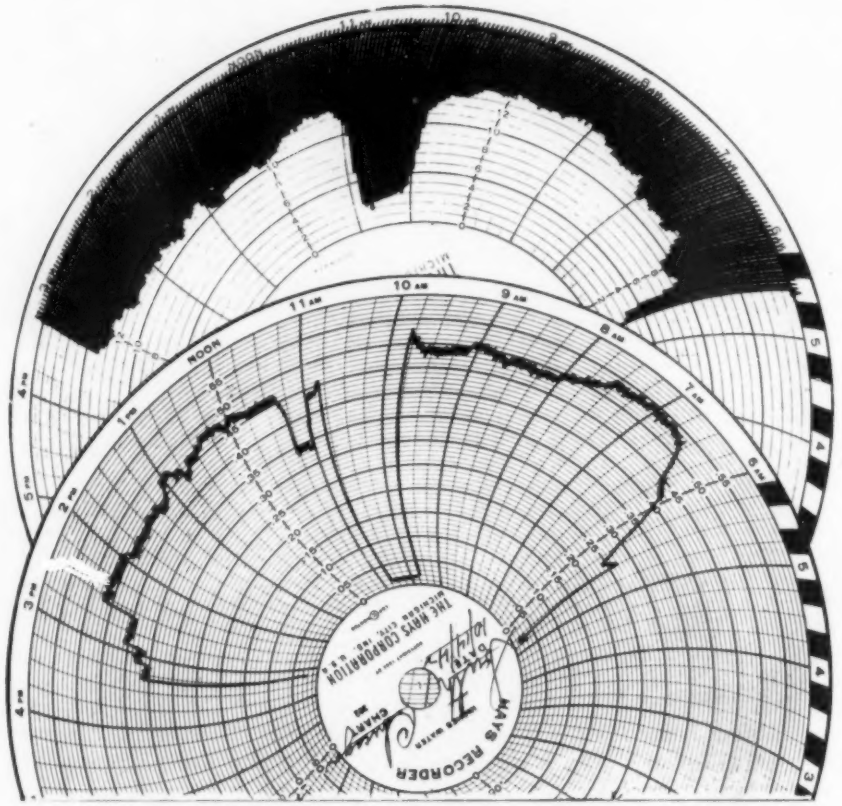


Fig. 6—Chart from a CO_2 recorder shows relationship of blast volume to CO_2 in effluent gas throughout a cupola heat.

required amount of heat at the required temperature to melt the iron in the exact time required to burn the coke.

If the iron charge melts in less time than required to burn the charge of coke then a new charge of coke comes down ahead of scheduled time and increases the height of the bed coke. If this condition should continue indefinitely, the cupola will become filled with coke, with no room for the iron charge.

If the iron charge is delayed in melting beyond the time required to burn a charge of coke, then the bed coke is burned away until a new charge of coke is released. If this condition continues, the space becomes filled with iron, at which point melting ceases.

How, then, is exact synchronization achieved? Simply by the laws of nature, over which we have no control when the coke charge and air volume are fixed. The operation in detail is easily explained.

For example, in the first case mentioned, if the iron is melted too soon, the remaining coke not burned increases the depth of the reducing zone, and a greater proportion of CO_2 is reduced to CO, thereby lowering the temperature of the coke, which has a tendency to retard the rate of melting, and also that portion of the coke which is used to reduce the CO_2 does not reach the combustion zone.

Hence there is less coke to burn and the time of burning is reduced. The result of this combined action is to increase the time of melting and reduce the time of combustion until automatically a stage of equalization is reached.

In the second case, where the time of melting a charge is retarded and the depth of reducing zone decreased, less CO_2 is reduced to CO. Consequently the temperature is increased and a less proportion of the coke is used to reduce CO_2. Therefore a greater proportion reaches the combustion zone, which increases the time required to burn the charge.

Here, again, this action continues automatically until exact synchronization is obtained, after which there is a constant fluctuation of perhaps one per cent more or less in the percentage of CO_2 by which the synchronization is maintained indefinitely. This regulation is just as positive as the governor on a Corliss engine which, when once set, controls the speed of the engine without attention of the operator.

Height of Bed Coke

To apply these principles definitely to the bed coke, Fig. 5 has been prepared for five different volumes of air per minute, namely: 150, 200, 250, 300, 350 cu. ft. per sq. ft. of area; or for a 60-in. cupola having 20 sq. ft. of area, these volumes become 3000, 4000, 5000, 6000, 7000 cu. ft. air per min.

In calculating the velocity through the cupola, it is assumed that the

Table 4

CO_2:CO RATIO WITHIN CUPOLA COMPARED WITH THAT WITHIN BLAST FURNACE

A			*B*			*C*		
Forsythe's Blast Furnace Percentage by Volume			*Prest's Cupola Observations 11-6-34 At Uniform Depth*			*Prest's Cupola Observations 11-6-34 At Various Depths*		
Depth, in ft.	*CO, per cent*	*CO_2, per cent*	*Constant Depth, in ft.*	*Time*	*CO_2, per cent*	*Depth, in ft.*	*Time*	*CO_2, per cent*
			Blast on	6.01	0.0			
Top	29.5	11.0	9.4	6.06	9.7	2	7.20	10.8
4	29.5	10.5	9.4	6.12	14.4	2½	7.25	9.6
8	27.0	8.0	9.4	6.18	4.4	3	7.30	10.0
10	32.0	7.0	9.4	6.26	3.4	4½	7.33	8.4
12	33.0	7.0	9.4	6.32	.5	5	7.36	6.8
14	31.0	6.5	9.4	6.45	1.2	7	7.42	1.0
16½	34.1	2.2	9.4	6.50	7.4	8	7.46	1.6
20	35.1	0.7	9.4	6.52	7.4	8½	7.50	1.0
39	35.0	1.1	9.4	6.55	5.6	9	7.53	5.0
52	35.2	1.5	9.4	7.00	Pipe	9½	7.57	3.8
65	35.9	0.5			Clogged	10	8.01	4.6
70½	36.6	0.0						
Tuyeres	37.7	0.8						

coke and iron occupy 60 per cent of the space, leaving 40 per cent of voids. This increases the velocity of the gases two and a half times above what it would be if the area were free.

Reducing the volumes indicated to ft. per sec. at normal atmospheric temperature, velocities become 75, 100, 125, 150 and 175 ft. respectively. However, since the gases are suddenly raised to a temperature above 3000° F. volume is increased sevenfold and velocity increased in the same proportion.

Velocity of Gas Reactions

This indicates that in less than one sec. of time the air has entered the cupola; entered into combustion liberating vast quantities of heat; has gone through a secondary reaction of reducing CO_2 to CO; has transferred its heat to melt the iron and to preheat the iron; and has passed through the stack to the atmosphere. This indicates the extreme velocity of the gaseous reactions.

The effect on the height of the bed coke is shown in Fig. 5, in which it is assumed that all the oxygen has been eliminated in 8/100ths of a sec. By calculating the distance traveled in this length of time, the depth of the combustion zone is 6-in. at the low velocity and 14-in. at the high velocity.

The diagram indicates that for the higher amounts of coke per ton of iron the bed coke has the maximum height, and as the coke per ton of iron becomes less and less, the bed coke drops to the point shown by the 11 to 1 ratio. The CO_2 percentages are shown in Fig. 1. We have no direct data on which to base these calculations, and the only object of Fig. 1 is to show the direction in which these reactions take place rather than to indicate definite dimensions.

CO_2 Recorder

Effect of velocity at which air is admitted to the cupola on the regulation of the amount of heat developed per pound of coke—in other words, the ratio of CO_2 to CO—is best shown by a CO_2 recorder in connection with an air volume gage.

For this purpose, Fig. 6 is introduced, which is a somewhat radical example for the reason that the iron

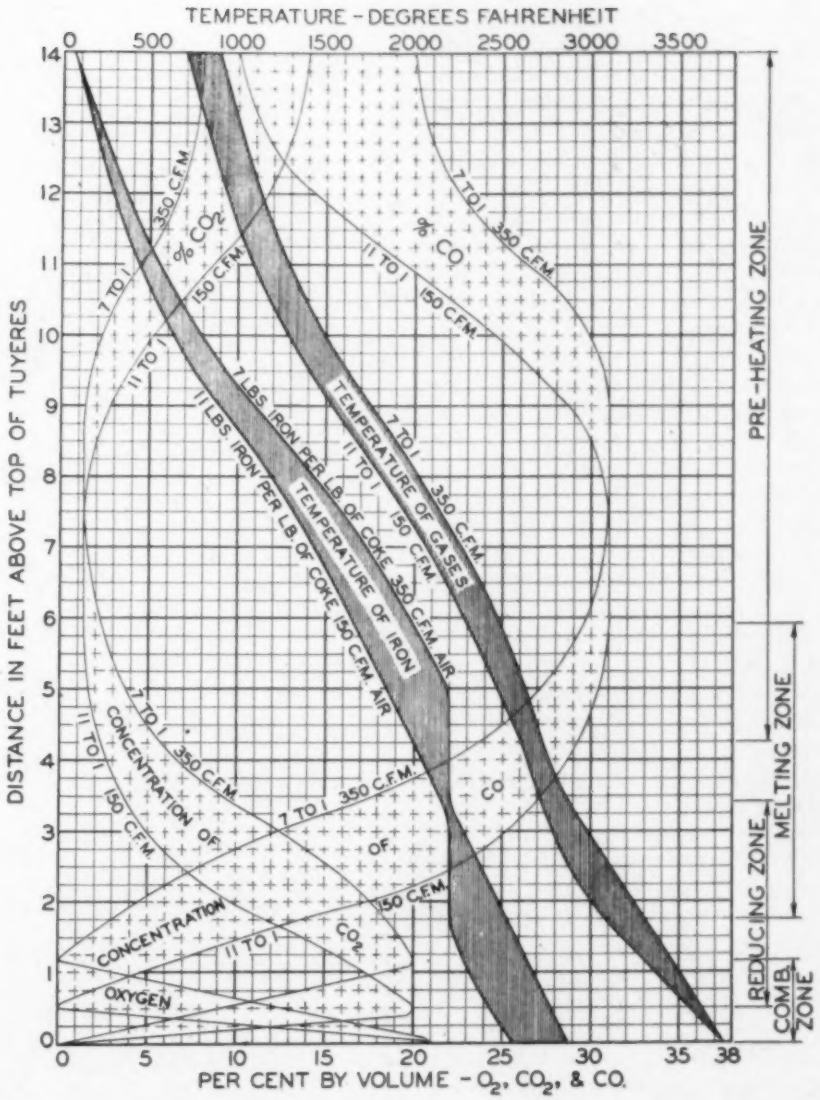


Fig. 7—Relation between gas and iron temperature and per cent oxygen, CO, and CO_2 in the various zones of the cupola for the usual range of operation.

from the cupola was tapped into an air furnace containing 30 tons of iron.

The iron from the air furnace was tapped into a receiving ladle from which the pouring ladles were filled. The air furnace was operated only through one half the heat, after which the iron from the cupola was tapped into the receiving ladle. But while the last 30 tons in the air furnace was delivered to the pouring ladles, the blast in the cupola was shut off for approximately one hour, as indicated in Fig. 6.

At the beginning, 6 A.M., bed coke had not burned down to an equilibrium, for the blast pressure was reduced to retard melting and to preheat charges, and to bring the walls of the cupola up to their required operating temperature. The low air pressure increased the time limit in which the CO_2 was in contact with the incandescent coke, resulting in a high percentage of CO and correspondingly low percentage of CO_2.

As the blast volume was increased and the bed coke burned down, there was a gradual increase in CO_2 percentage until an operational balance was reached at about 8:30 A.M. From then on a more or less constant percentage of CO_2 was maintained until the air volume was radically reduced at 9:45 A.M. to stop melting in the cupola and to permit pouring off 30 tons of metal from the air furnace.

The reduced blast, as indicated by Fig. 6, immediately produced a high reduction of CO_2 to CO, to such an extent that the temperature of the combustion and reducing zones was reduced to a point where melting practically ceased. After the lapse of 1 hr. 20 min., the air volume was again applied, with immediate decrease in the CO_2 reduction following very closely the inverse air volume record.

Near the end of the heat the blast was reduced momentarily with corresponding drop in CO_2 content, after which the volume was further reduced in response to the lowering charges from the time that charging ceased and the time that the entire metal was melted.

There is considerable reference in the literature on combustion that apparently does not agree with the direct relationship between air velocity and coke charge that is indicated by the CO_2 chart. However, wherever CO_2 recording charts are used it is evident that a change in blast pressure produces immediately a corresponding change in the CO_2-CO relation which is the basic factor in determining the amount of heat to be developed from each pound of fuel.

Every cupola should be equipped with a pressure gauge, a volume gauge and a CO_2 recording instrument. These will give continuous information regarding the various efficiencies under which the cupola is operating at each moment and furnish definite means for close regulation.

Changeable Atmospheres

CO_2-CO ratio within cupola is not in a state of equilibrium but, on the contrary, is constantly changing with great rapidity. It has already been pointed out that the gas CO_2 is unstable at high temperatures in the presence of incandescent carbon.

The reduction of CO_2 to CO continues through the melting zone to an extent far greater than the final equilibrium which is formed at the lower temperatures with iron as a catalyst. This is a condition that prevails in the upper part of the preheating zone.

This reaction is the reverse of what takes place in the reducing zone and by means of which CO is changed back to CO_2; thus $2CO = C + CO_2$. While this reaction has

Table 5

CALCULATIONS FOR HEAT DEVELOPED IN THE CUPOLA

Lb. coke burned					0.6875
Lb. coke used in reduction CO_2					0.3125
Total coke disappeared					1.0000 lb.
Heat Developed in the Combustion Zone					*Btu.*
13095 × 0.6875 =					9003.79
Sensible heat from .6875 lb. coke					870
Total heat developed in Combustion Zone					9873
Heat Developed in Reducing Zone					
Sensible heat from 0.3125 lb. coke					390
Total Heat Developed 1 Lb. Coke					10263
Not Developed, 13095 × 0.3125					4092
Total Potential heat					14355
Lb. of Gas in Reduction Zone (See Table 24 of Appendix)	*1 lb. Carbon*				*1 lb. Coke*
CO_2	1.3781	×	0.90	=	1.24029 lb.
CO	1.4625	×	0.90	=	1.31625 lb.
Nitrogen	6.4721	×	0.90	=	5.82489 lb.
Excess Oxygen	0.1161	×	0.90	=	0.10449 lb.
Total Lb. Gas	9.4288	×	0.90	=	8.48592 lb.
Deduct	0.3125	×	0.90	=	0.28125 lb.
Lb. of Gas in Combustion Zone	9.1163	×	0.90	=	8.20467 lb.
Lb. Gas in Combustion Zone				8.2	
Lb. Gas in Reduction Zone				8.5	

Transfer of Heat			*Specific*	*Approx.*
Combustion Zone		*Btu.*	*Heat*	*Temp. °F.*
Heat Developed		9872	.32	3800
Loss:				
Superheat Iron	326			
Development of Slag	480	806		−300
Heat delivered to *Reducing Zone*		9066		3500
Add sensible heat in .3125 lbs. coke		390		+330
Total heat in Reducing Zone		9456		
Loss:				
Superheat Iron	650			
Preheating Coke	235			
Superheat Slag	240			
Reducing CO_2 to CO	1225	2350		−950
Heat delivered to *Melting Zone*		7106	.29	2880
Loss:				
Latent heat of fusion, 8 lbs. iron	576			
Preheating Iron	272			
Preheating Coke	245			−400
Formation of Slag	110	1203		
Heat delivered to *Preheating Zone*		5903	.28	2480
Loss:				
Preheat Iron, 2000°	2624			
Preheat Coke, 2000°	780			
Calcining Limestone	370			
Reduce CO_2 to CO	420	4194		−1680
In effluent gases		1709	.25	800

SUMMARY

Heat Used	*Btu.*	*Per cent*
Absorbed by Iron	4448	34.0
Absorbed by Slag	830	6.34
Calcining Limestone	370	2.82
Total	5648	43.1
Heat Not Used		
Through stack	1709	13.1
CO	1645	12.6
Coke not burned	4093	31.2
Total	7447	56.9
Total Heat Used and Not Used	13095	100.0

Table 6

HEAT BALANCES FOR USUAL RANGE OF VARIATIONS IN CUPOLA PRACTICE

	Lb. Iron Per Lb. Coke	Lb. Coke Per Ton Melted	Blast Press. oz.	Air, Cu. Ft. Per Min.	Air, Cu. Ft. Per Hr.	CO2, per cent	Air Required Per Lb. Coke Cu. Ft.	Lb. of Coke Burned Per Hr.	Lb. of Iron Melted Per Hr.	Btu. Per Lb. of Coke (Carbon) (90%) Deel-oped	Btu. Per Lb. of Coke (Carbon) (90%) Unde-veloped	Effluent Gases Lb.	Effluent Gases Temp., °F.	Btu. Loss	Btu. in Slag and Other Losses	Total Loss, Btu.	Btu. in Iron	Btu. in 1 Lb. of Iron	Temp. °F. of Iron	Per cent of Standard Rate Com-bustion	Per cent of Standard Rate Melting	Heat Un-developed	Per cent Loss Sensible Heat in Gases At	Per cent Loss Sensible Heat in Gases B‡	Per cent Loss Slag and Other Losses A†	Per cent Loss Slag and Other Losses B‡	Efficiency Combustion	Efficiency Heat Absorption by Iron	Efficiency Overall
7	285.7	4	150	9,000	8.0	87.9	7.1	102.4	716.7	6538	6557	8.0	762	1524	1150	2674	3864	552	2710	73.1	51.2	50.1	11.6	23.3	8.8	17.6	49.9	59.1	29.5
	285.7	6	200	12,000	8.5	89.4	7.2	134.2	939.5	6736	6359	8.1	795	1609	1200	2809	3927	561	2750	95.9	67.1	48.6	12.3	23.9	9.2	17.8	51.4	58.3	30.0
	285.7	10	250	15,000	9.0	90.9	7.3	165.0	1155.1	6938	6157	8.2	828	1698	1250	2948	3990	570	2790	117.9	82.5	47.0	13.0	24.5	9.5	18.0	53.0	57.5	30.5
	285.7	15	300	18,000	9.5	92.4	7.5	194.8	1363.6	7145	5950	8.4	853	1792	1300	3092	4053	579	2830	139.1	97.4	45.4	13.7	25.1	9.9	18.2	54.6	56.7	31.0
	285.7	20	350	21,000	10.0	94.0	7.6	223.4	1563.8	7357	5738	8.5	887	1884	1350	3234	4123	589	2875	159.6	111.7	43.8	14.4	25.6	10.3	18.3	56.2	56.0	31.5
8	250.0	4	150	9,000	9.0	90.9	7.3	99.0	792.0	6938	6157	8.2	740	1518	1100	2618	4320	540	2660	70.7	56.6	47.0	11.6	21.9	8.4	15.9	53.0	62.3	33.0
	250.0	6	200	12,000	9.5	92.4	7.5	129.9	1039.0	7145	6950	8.4	767	1611	1150	2761	4384	548	2695	92.8	74.2	45.4	12.3	22.5	8.8	16.1	54.6	51.4	33.5
	250.0	10	250	15,000	10.0	94.0	7.6	159.6	1276.6	7357	5738	8.5	804	1709	1200	2909	4448	556	2730	114.0	91.2	43.8	13.1	23.2	9.2	16.3	56.2	60.5	34.0
	250.0	15	300	18,000	10.5	95.6	7.7	188.3	1506.2	7575	5520	8.6	840	1805	1250	3055	4520	565	2770	134.5	107.6	42.2	13.8	23.8	9.5	16.5	57.8	59.7	34.5
	250.0	20	350	21,000	11.0	97.3	7.9	215.8	1726.6	7799	5296	8.8	856	1883	1300	3183	4616	577	2820	154.2	123.3	40.4	14.4	24.1	9.9	16.7	59.6	59.2	35.3
9	222.2	4	150	9,000	10.0	94.0	7.6	95.7	861.7	7357	5738	8.5	719	1528	1050	2578	4779	531	2620	68.4	61.5	43.8	11.7	20.8	8.0	14.3	56.2	65.0	36.5
	222.2	6	200	12,000	10.5	95.6	7.7	125.5	1129.7	7575	5520	8.6	764	1642	1100	2742	4833	537	2645	89.6	80.7	42.2	12.5	21.7	8.4	14.5	57.8	63.8	36.9
	222.2	10	250	15,000	11.0	97.3	7.9	154.2	1387.4	7799	5296	8.8	793	1744	1150	2894	4905	545	2680	110.1	99.1	40.4	13.3	22.4	8.8	14.7	59.6	62.9	37.5
	222.2	15	300	18,000	11.5	99.0	8.0	181.8	1636.3	8028	5067	8.9	828	1842	1200	3042	4986	554	2720	129.9	116.9	38.7	14.1	22.9	9.2	14.9	61.3	62.1	38.1
	222.2	20	350	21,000	12.0	100.7	8.1	208.5	1876.9	8263	4832	9.0	853	1919	1250	3169	5094	566	2755	149.0	134.1	36.9	14.7	23.2	9.5	15.1	63.1	61.6	38.9
10	200.0	4	150	9,000	11.0	97.3	7.9	92.5	924.9	7799	5296	8.8	709	1559	1000	2559	5240	524	2590	66.1	66.1	40.4	11.9	20.0	7.6	12.8	59.6	67.2	40.0
	200.0	6	200	12,000	11.5	99.0	8.0	121.2	1212.1	8028	5067	8.9	754	1678	1050	2728	5300	530	2615	86.6	86.6	38.7	12.8	20.9	8.0	13.1	61.3	66.0	40.5
	200.0	10	250	15,000	12.0	100.7	8.1	149.0	1489.5	8263	4832	9.0	797	1793	1100	2893	5370	537	2645	106.4	106.4	36.9	13.7	21.7	8.4	13.3	63.1	65.0	41.0
	200.0	15	300	18,000	12.5	102.6	8.3	175.4	1754.3	8506	4589	9.2	824	1896	1150	3046	5460	546	2685	125.3	125.3	35.0	14.5	22.3	8.8	13.5	65.0	64.2	41.7
	200.0	20	350	21,000	13.0	104.4	8.4	201.1	2011.4	8752	4343	9.3	852	1982	1200	3182	5570	557	2735	143.7	143.7	33.2	15.1	22.6	9.2	13.7	66.8	63.6	42.5
11	181.8	4	150	9,000	12.0	100.7	8.1	89.4	983.1	8263	4832	9.0	696	1565	1000	2565	5698	518	2565	63.8	70.2	36.9	12.0	18.9	7.6	12.1	63.1	69.0	43.5
	181.8	6	200	12,000	12.5	102.6	8.3	117.0	1286.5	8506	4589	9.2	736	1692	1050	2742	5764	524	2590	83.5	91.9	35.0	12.9	19.9	8.0	12.3	65.0	67.8	44.6
	181.8	10	250	15,000	13.0	104.4	8.4	143.7	1580.4	8752	4343	9.3	779	1811	1100	2911	5841	531	2620	102.6	112.9	33.2	13.8	20.7	8.4	12.6	66.8	66.7	44.6
	181.8	15	300	18,000	13.5	106.3	8.6	169.3	1862.6	9008	4087	9.5	812	1929	1150	3079	5929	539	2655	121.0	130.0	31.2	14.7	21.4	8.8	12.8	68.8	65.8	45.3
	181.8	20	350	21,000	14.0	108.2	8.7	194.1	2134.9	9270	3825	9.6	846	2031	1200	3231	6039	549	2700	138.6	152.5	29.2	15.5	21.9	9.2	12.9	70.8	65.1	46.1

NOTES:
*The standard rate of combustion used for this table is 140 lb. of coke per hr.; iron to coke ratio 10 to 1. The standard melting rate is therefore 1400 lb. per hr.
†The per cent of heat loss shown in columns marked A is based on the potential heat of the coke (13,095 Btu.)
‡The per cent loss shown in columns marked B is based on the number of Btu. developed for various stages of combustion (per cent CO_2).

been known to occur in blast furnaces, it is seldom referred to by writers on Cupola Practice.

In order to determine the extent to which this reaction takes place and the conditions which are involved in obtaining an equilibrium, observations were made in 1934 under the direction of the author by S. C. Prest, principal assistant in the Cupola Department of the Griffin Wheel Company. Gas analyses at various elevations in the cupola from the charging door to the melting zone were examined in several cupolas from Boston to Los Angeles.

A previous analysis of this reaction in cupola practice was made by F. C. Thompson, Professor of Chemistry at Manchester University, England, in a paper entitled "Some Experiments on the Reaction $2CO = C + CO_2$ with Special Reference to Fuel Economy in Cupola Practice," in the *Foundry Trade Journal*, June 7, 1928, p. 405, vol. 38.

This phenomenon has also been examined and recorded by Forsythe in his book, "The Blast Furnace and the Manufacture of Pig Iron". Gas samples were taken to indicate the reaction taking place at various levels in the blast furnace.

Another series of tests is recorded by the Carnegie Bureau Book of Instructions, p. 168 and 169, under the subject, *The Making, Shaping and Treating of Steel.* The results are shown diagramatically and follow very closely the observations made by Forsythe.

The results of the observations made by Forsythe and Prest are recorded in Table 4.

The observations by Mr. Prest were taken from a Chicago cupola, under two conditions on the same day, and may be assumed as average conditions.

The first set of observations were taken at a constant elevation at the top of the bed coke, beginning immediately after the blast was applied. Samples of gas continued to be taken at intervals of a few minutes for an hour. The results are shown in Table 4B.

These observations were not made under laboratory conditions and only serve the purpose of indicating the trend of the equilibrium between CO and CO_2. By comparison of the data in Table 4 it is obvious that the results in the blast furnace and

in the cupola are similar to the extent that the cooler portions near the top are favorable to the precipitation of carbon from the CO, causing an increase in CO_2.

It remains for more delicate methods to fix the definite limits of time and temperature which govern this reverse reaction,

$$2CO = CO_2 + C$$

The greatest discrepancy between the blast furnace and the cupola is in the lower regions where the CO_2 gas in the cupola at the high temperature of the melting zone and low bed permit oxidation, while in the blast furnace CO is produced in the combustion zone, which is the condition for developing the desired reducing atmosphere.

This is easily explained by the different coke to air ratios and the desire, in one case, to develop the conditions which produce the maximum CO, and, in the other, to produce the maximum CO_2 consistent with controlled oxidation.

Table 4B indicates the powerful effect of temperature on the reverse reaction in which the samples were taken close to the melting zone just after the blast was applied, while the furnace is comparatively cold. The CO_2 will be comparatively high, but as the coke in this region becomes heated the CO_2 content nearly disappears.

At this stage, with the high bed, the chemical reactions approach those of the blast furnace. However, as the bed is burned away to supply the necessary heat to bring the temperature of the cupola walls and the preheating of charges to a normal working condition, a balance is reached where the coke to iron ratio and the rate at which the coke is fed into the combustion zone result in a constant elevation of the bed coke above the tuyeres.

When this condition is reached, the time limit of contact between air and coke is reduced, hence CO_2 is increased. The low percentage of CO_2 while the height of the bed is being reduced produces a condition of low oxidation in which the silicon loss is negligible.

In the Thompson and Becker observations, they report as evidence that the reaction does not occur at temperatures above 1300 or 1400° F. and that the reaction continues as the temperature decreases to 400° F.

Table 7

HEAT BALANCE—COLD BLAST CUPOLA

Basis: 1 lb. Coke; Varying Iron Ratios; Volume air constant, 250 cfm. per sq. ft. cupola area.

Iron Melted per lb. Coke	7	8	9	10	11
Lb. of Coke					
For melting one ton of iron	285.7	250.0	222.2	200.0	181.8
For combustion CO_2	190.0	171.9	158.1	147.4	138.8
For reduction CO_2 to CO	95.7	78.1	64.1	52.6	43.0
Rate disappearance, lb. per hr.	165.0	159.6	154.2	149.0	143.7
Rate of melting, tons per hr.	.577	.638	.694	.745	.790
Per cent CO_2 in effluent gases	9.0	10.0	11.0	12.0	13.0
Lb. of Air					
For each lb. coke charged	7.33	7.58	7.85	8.13	8.42
For each lb. coke burned	11.03	11.03	11.03	11.03	11.03
For one ton iron melted	2094	1895	1744	1626	1531
Lb. of Gases					
Per lb. coke disappeared	8.23	8.48	8.75	9.03	9.32
Per ton iron melted	2351	2120	1944	1806	1695
Disappearance 1 lb. Coke					
Percent burned to CO_2 (A+½B)	66.46	68.75	71.15	73.68	76.35
Per cent reduced, CO_2 to CO (½B)	33.54	31.25	28.85	26.32	23.65
	100.00	100.00	100.00	100.00	100.00
Btu. Developed					
From combustion (A+½B)	8704	9003	9318	9649	9998
Sensible heat (A+½B)	840	870	900	930	960
Total in Combustion Zone	9544	9873	10218	10579	10958
Sensible heat (½B) in Reducing Z.	420	390	360	330	300
Total heat below Melting Zone	9964	10263	10578	10909	11258
Temperature of Gases	3850	3850	3850	3850	3850
Btu. Loss, CO_2 to CO	1766	1645	1519	1386	1246
Net Btu. developed 1 lb. coke	6938	7357	7799	8263	8752
Potential Btu. not developed	6157	5738	5296	4832	4343
Total Potential 1 lb. coke	13095	13095	13095	13095	13095
Add sensible heat transferred	1260	1260	1260	1260	1260
Total Btu. below Melting Zone	14355	14355	14355	14355	14355
Temperature Iron					
Fluid Iron in Melting Zone	2200	2200	2200	2200	2200
Iron at Tapping Hole	2790	2730	2680	2645	2620
Degrees superheat	590	530	480	445	420
Btu. Transferred to Iron					
In solid iron at Melting Zone	2296	2624	2952	3280	3508
In fluid iron in Melting Zone	3038	3472	3906	4340	4774
In iron as tapped	3990	4448	4905	5370	5841
In superheat	952	976	999	1030	1067
Btu. Transferred from Gases in Combustion Zone					
Superheating Iron	317	326	333	350	355
Formation and Superheating Slag.	520	480	440	400	400
Total	837	806	773	750	755
Temp. Gases entering Reducing Zone..	3470	3490	3510	3540	3540
Btu. Transferred in Reducing Zone					
Superheating Iron	635	650	666	680	712
Preheating Coke	235	235	235	235	235
Superheating Slag	250	240	230	220	220
Reducing CO_2 to CO	1316	1225	1129	1036	926
Total	2436	2350	2260	2171	2093
Temp. Gases delivered to Melting Zone	2880	2940	2990	3000	3060
Btu. Transferred in Melting Zone					
Latent heat fusion	504	576	648	720	792
Preheating Iron, 200° F.	238	272	306	340	374
Preheating Coke, 500° F.	245	245	245	245	245
Formation and Superheating Slag..	110	110	110	110	110
Total	1097	1203	1309	1415	1521
Temp. Gases delivered to Preheating Z.	2490	2530	2560	2560	2610
Total Btu. Transferred Below Preheating Zone	4370	4359	4342	4336	4369
Btu. remaining	5594	5903	6236	6573	6889
Transferred in Preheating Zone					
Preheating Iron to 2000° F.	2296	2624	2952	3280	3608
Preheating Coke to 2000° F.	780	780	780	780	780
Calcining Limestone	370	370	370	370	370
Reducing CO_2 to CO	450	420	390	350	320
Total	3896	4194	4492	4780	5078
Btu. in Effluent Gases	1698	1709	1744	1793	1811
Temperature, ° F.	828	804	793	797	779

Table 8

HEAT BALANCE—COLD BLAST CUPOLA

Basis: 1 lb. Iron; Varying Coke Ratios; Volume Air constant, 250 cfm. per sq. ft. cupola area.

Lb. Iron Per 1 Lb. Coke	7	8	9	10	11
Lb. of Coke					
For melting one ton of iron	285.7	250.0	222.2	200.0	181.8
For combustion CO_2	190.0	171.9	158.1	147.4	138.8
For reduction CO_2 to CO	95.7	78.1	64.1	52.6	43.0
Per Sq. Ft. Cupola Area Per Hour					
Rate of coke disappearance, lb.	165.0	159.6	154.2	149.0	143.7
Rate of melting, tons	.577	.638	.694	.745	.790
Per cent CO_2 in effluent gases	9.0	10.0	11.0	12.0	13.0
Lb. Air					
Per Lb. Iron Melted	1.047	.947	.872	.813	.765
Per Lb. Coke Burned	11.03	11.03	11.03	11.03	11.03
Per Ton Iron Melted	2094	1895	1744	1626	1531
Lb. Gases					
Per Lb. Iron	1.176	1.060	.972	.903	.847
Per Ton Iron	2351	2120	1944	1806	1695
Disappearance Coke Per Lb. Iron					
Lb. burned to CO_2 (A+½B)	.0949	.0859	.0790	.0737	.0694
Lb. reducing CO_2 to CO (½B)	.0479	.0391	.0321	.0263	.0215
Total coke melting 1 lb. iron	.1428	.1250	.1111	.1000	.0909
Btu. Developed in Combustion Zone					
From combustion (A+½B)	1243	1125	1035	965	909
Sensible heat (A+½B)	120	109	100	93	87
Total developed in Combustion Zone	1363	1234	1135	1058	996
Sensible heat Reduction Zone (½B)	60	49	40	33	27
Total Btu. below Melting Zone	1423	1283	1175	1091	1023
Temperature Gases	3850	3850	3850	3850	3850
Btu. Loss—Reducing CO_2 to CO	252	206	169	139	113
Net Btu. developed below Melting Zone	991	920	866	826	795
Potential—not developed	879	717	589	483	395
Btu. Potential in coke for 1 Lb. Iron	1871	1637	1455	1309	1190
Btu. Sensible heat in coke at 3000° F.	180	158	140	126	114
Total Btu. below Melting Zone	2051	1795	1595	1435	1305
Temperature Iron in Melting Zone, ° F.	2200	2200	2200	2200	2200
Temperature Iron at Tapping Hole	2790	2730	2680	2645	2620
Degrees Superheat	590	530	480	445	420
Btu. Solid Iron in Melting Zone	328	328	328	328	328
Btu. Fluid Iron	434	434	434	434	434
Btu. as tapped	570	556	545	537	531
Btu. in Superheat	136	122	111	103	97
In Combustion Zone					
Superheating Iron	45.3	40.7	37.0	35.0	32.3
Formation Slag	74.3	60.0	48.9	40.0	36.3
Total	119.6	100.7	85.9	75.0	68.6
Temperature Gases—top of zone	3470	3490	3510	3540	3540
In Reducing Zone					
Superheating Iron	90.7	81.2	74.0	68.0	64.7
Preheating Coke	33.6	29.4	26.1	23.5	21.4
Superheating Slag	35.7	30.0	25.6	22.0	20.0
Reducing CO_2 to CO	188.0	153.4	125.4	103.6	83.9
Total	348.0	294.0	251.1	217.1	190.0
Temperature Gases—top of zone	2880	2940	2990	3000	3060
In Melting Zone					
Latent heat fusion	72.0	72.0	72.0	72.0	72.0
Preheating Iron—200° F.	34.0	34.0	34.0	34.0	34.0
Preheating Coke—500° F.	35.0	30.6	27.2	24.5	22.3
Formation and Superheating Slag	15.7	13.8	12.2	11.0	10.0
Total	156.7	150.4	145.4	141.5	138.3
Temperature Gases—top of zone	2490	2530	2560	2560	2610
Btu. Transferred Below					
Preheating Zone	624.3	545.1	482.4	433.6	396.9
Remaining for Preheating Zone	799.1	737.9	692.9	657.3	626.3
Total	1423.4	1283.0	1175.3	1090.9	1023.2
In Preheating Zone					
Preheating Iron—2000° F.	328.0	328.0	328.0	328.0	328.0
Preheating Coke—2000° F.	111.4	97.5	86.7	78.0	70.9
Calcining Limestone	52.9	46.2	41.1	37.0	33.6
Reducing CO_2 to CO	64.3	52.5	43.3	35.0	29.1
Total Btu. transferred	556.6	524.2	499.1	478.0	461.6
Remaining in Effluent Gases	242.5	213.6	193.8	179.3	164.7
Temperature	828	804	793	797	779

They found the reverse action to be relatively slow, hence the overall extent to which CO_2 can be increased is regulated by reducing the blast pressure, thus allowing a greater time in the cupola before the gases are discharged from the cupola as well as providing more time for the reaction to take place.

It is suggested that increasing the height of the cupola would have the same effect.

In their laboratory experiments, it was shown that iron is the most effective catalyst in promoting this reaction and that the rate varies in proportion to the surface area of the iron.

The extent to which the phenomena takes place at the various temperature levels in the cupola is shown in Fig. 7. It also shows the temperature of iron and temperature of the gases throughout the cupola for various coke ratios.

Figure 7 illustrates the care required in obtaining necessary gas samples for analysis or in taking temperature measurements due to the rapidly changing conditions which take place at various points within the cupola. Consistent results, however, can be secured by taking the samples sufficiently near the charging doors.

Correct sampling is a fundamental necessity in the calculation of thermal efficiencies which depends upon three items, namely: 1 temperature and composition of the effluent gases; 2 temperature of the iron as tapped from the cupola; 3 and temperature of the incoming blast.

Heat balances are essential for the study of fuel economy. This information is not available, except in a very general way, in the literature on cupola practice. The balance of this analysis of the conditions which exist in the cupola is devoted to a detailed study of the fundamentals which underlie fuel economy, with copious tabulations based upon actual operating conditions.

1 Lb. Coke Through the Cupola

On p. 510 1 lb. of carbon was followed through the cupola from the charging door to the combustion zone. It will now be followed on its return trip as a gas through the cupola from the combustion zone to the stack where it is discharged to the atmosphere.

Table 9

RELATION OF HEAT DEVELOPED TO HEAT USED AND NOT USED
ALSO
RELATION OF HEAT AVAILABLE TO HEAT NOT AVAILABLE FOR MELTING

Lb. Iron per lb. Coke	7		8		9		10		11	
Per cent CO_2	9		10		11		12		13	
	Btu.	Equivalent Lb. Iron	Btu.	Equivalent Lb. Iron	Btu.	Equivalent Lb. Iron	Btu.	Equivalent Lb. Iron	Btu.	Equivalent Lb. Iron
Btu. Developed—										
1-lb. Coke	6938	12.8	7358	13.6	7799	14.4	8263	15.3	8753	16.2
Not Used	1698	3.1	1709	3.2	1744	3.2	1793	3.3	1811	3.3
Net Used	5240	9.7	5649	10.5	6065	11.2	6470	12.0	6942	12.9
Not Developed	6157	11.4	5738	10.6	5296	9.8	4832	9.0	4343	8.0
Total Not Used	7855	14.5	7447	13.8	7040	13.0	6625	12.3	6154	11.3
Per Cent Not Used	60.0		56.8		53.7		50.5		47.0	
Btu. Not Available for Melting Below 2600° F.	5594	10.4	5903	10.9	6236	11.5	6573	12.1	6889	12.7
Not Used	1698	3.1	1709	3.1	1744	3.2	1793	3.3	1811	3.3
Used in Preheating	3896	7.3	4194	7.8	4492	8.3	4780	8.8	5078	9.4

For this purpose consider the stage of combustion to be represented by 10 per cent CO_2+16.67 per cent CO, substituting for the one pound of carbon, one pound of coke (0.90 per cent carbon), having a calorific value of 13,095 Btu.

In order to determine the amount of heat developed under the conditions mentioned refer to Table 28 on page 534, which forms the basis for the following calculations in Table 5.

The conspicuous observation in the summary in Table 5 is the relatively small percentage of the potential heat (43.1 per cent) that is used and the large percentage (56.9 per cent) which escapes through the stack and is invisible to those who do not visualize the Btu. involved.

Cupola operators are not called upon to account for the potential heat in the fuel which they use. This is analagous to sending an employe on a business trip and not requiring an expense account.

To illustrate, supposing each Btu. represented one cent, then $130.95 would be the amount advanced, the identical trip being repeated many times, with no accounting. Finally, however, the auditor decides to look into the details of expenditure, whereupon the employe furnishes the following information (see summary in Table 5) on page 516.

(Btu. Used) Railroad fare and hotel bill ..... $ 56.48
(Btu. Undeveloped) Disappeared before starting 40.93
(Btu. CO_2 to CO) Loaned to a friend 16.45
(Btu. Thru Stack) Cash balance, thrown away 17.09

(Btu. Potential) Total advanced..$130.95

This condition would not be allowed to continue very long. However, it is a permanent condition regarding the use of fuel.

For the purpose of comparing the relative efficiency in fuel economy for varying coke ratios and air volumes in cold blast cupolas, the following heat balances have been prepared:

Table 6. A general table of heat balances for five coke ratios and five air volumes as indicated by the twenty-five intersections in Fig. 1.

Table 7 shows five detailed heat balances for five iron ratios with constant air volume of 250 cu. ft. per min. per sq. ft. of cupola area. These correspond with the five intermediate intersections on Fig. 1.

Table 8. These heat balances are the same as in Table 7, except that 1 lb. iron with varying coke ratios is used as a basis.

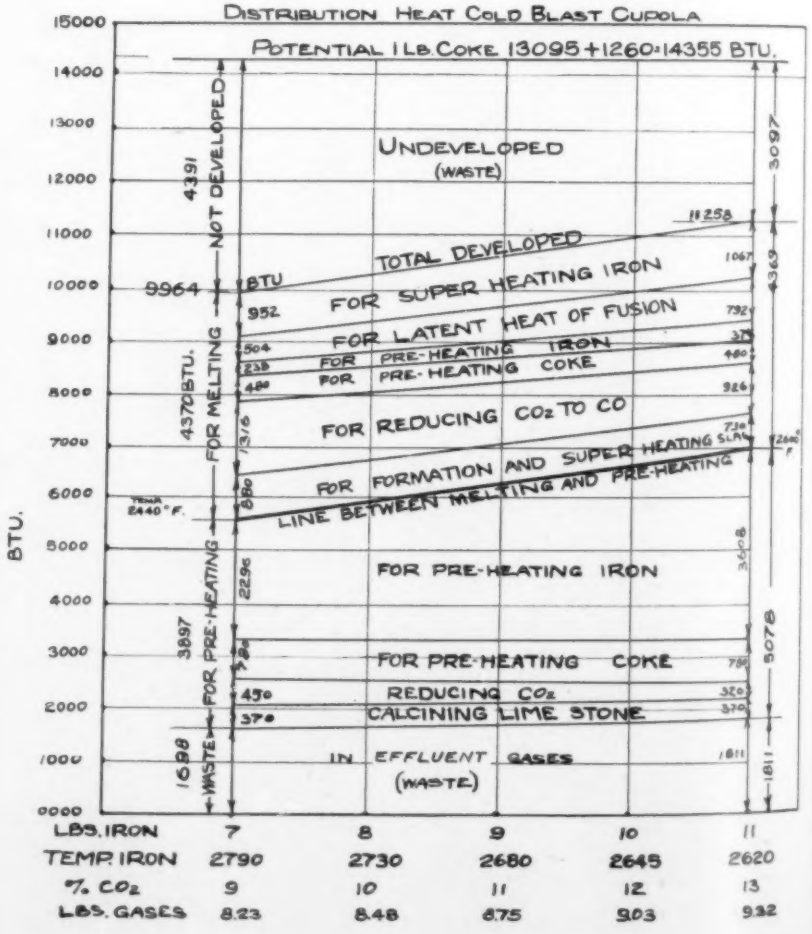


Fig. 8—Chart shows distribution of heat in a cold blast cupola.

Table 10

DISTRIBUTION OF HEAT TRANSFER THROUGH THE CUPOLA

7 LB. IRON — 1 LB. COKE

Potential Heat ..... 13,095 Btu.
9 per cent CO_2 develops..... 8,704 + 1260 = 9,964

Undeveloped ..... 4,391 Btu.

Transfer of Heat (Btu.) to various zones of cupola	*Combustion*	*Reduction*	*Melting*	*Total for Melting*	*Preheat*	*Total*
Iron—						
Preheat			238	238	2296	2534
Latent heat fusion			504	504		504
Superheat	317	635		952		952
Total	317	635	742	1694	2296	3990
Coke Preheat		235	245	480	780	1260
Slag—Formation and Superheating	520	250	110	880		880
Reduce CO_2 to CO		1316		1316	450	1766
Calcining Limestone					370	370
Total	837	2436	1097	4370	3896	8266
Effluent Gases						1698
Total Btu.						9964

8 LB. IRON — 1 LB. COKE

Potential Heat ..... 13,095 Btu.
10 per cent CO_2 develops..... 9,002 + 1260 = 10,262

Undeveloped ..... 4,093 Btu.

Transfer of Heat (Btu.) to various zones of cupola	*Combustion*	*Reduction*	*Melting*	*Total for Melting*	*Preheat*	*Total*
Iron—						
Preheat			272	272	2624	2896
Latent heat fusion			576	576		576
Superheat	326	650		976		976
Total	326	650	848	1824	2624	4448
Coke Preheat		235	245	480	780	1260
Slag—Formation and Superheating	480	240	110	830		830
Reduce CO_2 to CO		1225		1225	420	1645
Calcining Limestone					370	370
Total	806	2350	1203	4359	4194	8553
Effluent Gases						1709
Total Btu.						10262

9 LB. IRON — 1 LB. COKE

Potential Heat ..... 13,095 Btu.
11 per cent CO_2 develops..... 9,318 + 1260 = 10,578

Undeveloped ..... 3,777 Btu.

Transfer of Heat (Btu.) to various zones of cupola	*Combustion*	*Reduction*	*Melting*	*Total for Melting*	*Preheat*	*Total*
Iron—						
Preheat			306	306	2952	3258
Latent heat fusion			648	648		648
Superheat	333	666		999		999
Total	333	666	954	1953	2952	4905
Coke Preheat		235	245	480	780	1260
Slag—Formation and Superheating	440	230	110	780		780
Reduce CO_2 to CO		1129		1129	390	1519
Calcining Limestone					370	370
Total	773	2260	1309	4342	4492	8834
Effluent Gases						1744
Total Btu.						10578

10 LB. IRON — 1 LB. COKE

Potential Heat ..... 13,095 Btu.
12 per cent CO_2 develops..... 9,649 + 1260 = 10,909

Undeveloped ..... 3,446 Btu.

Transfer of Heat (Btu.) to various zones of cupola	*Combustion*	*Reduction*	*Melting*	*Total for Melting*	*Preheat*	*Total*
Iron—						
Preheat			340	340	3280	3620
Latent heat fusion			720	720		720
Superheat	350	680		1030		1030
Total	350	680	1060	2090	3280	5370
Coke Preheat		235	245	480	780	1260
Slag—Formation and Superheating	400	220	110	730		730
Reduce CO_2 to CO		1036		1036	350	1386
Calcining Limestone					370	370
Total	750	2171	1415	4336	4780	9116
Effluent Gases						1793
Total Btu.						10909

11 LB. IRON — 1 LB. COKE

Potential Heat ..... 13,095 Btu.
13 per cent CO_2 develops..... 9,998 + 1260 = 11,258

Undeveloped ..... 3,097 Btu.

Transfer of Heat (Btu.) to various zones of cupola	*Combustion*	*Reduction*	*Melting*	*Total for Melting*	*Preheat*	*Total*
Iron—						
Preheat			374	374	3608	3982
Latent heat fusion			792	792		792
Superheat	355	712		1067		1067
Total	355	712	1166	2233	3608	5841
Coke Preheat		235	245	480	780	1260
Slag—Formation and Superheating	400	220	110	730		730
Reduce CO_2 to CO		926		926	320	1246
Calcining Limestone					370	370
Total	755	2093	1521	4369	5078	9447
Effluent Gases						1811
Total Btu.						11258

Fuel Economy

Having followed 1 lb. of coke through the cupola from charging door to combustion zone where it is transformed into approximately 9 lb. of gases which travel upward, distributing the heat developed by combustion to the various zones of the cupola, finally carrying the surplus heat to the atmosphere, the results are recorded in heat balances for various coke ratios and air volumes.

The outstanding observation is the large percentage of waste heat escaping to the atmosphere, which is shown to be from 47 to 60 per cent.

Inasmuch as coke is the most essential material used in cupola practice and is used in far greater quantities than other materials, and since the unit price has advanced rapidly in recent years, it is more than ever important to consider economy in its use.

The obvious waste is expressed by Pat Dwyer in the *Foundry*, September 15, 1926: "At one time or another probably every foundryman responsible for results either to himself or to an employer has stood on the cupola charging deck watching the flame go roaring up the stack and wondering why the same flame could not be harnessed in some manner and utilized instead of going to waste."

This waste is so obvious that innumerable inventors have attempted to reduce it by a modification of cupola design.

Tuyeres of all types and shapes from continuous to a few individual tuyeres have been used. Their shapes are:

Horizontal
Vertical
Combined Horizontal and Vertical
Inverted T
Expanded
Oval
Round
Triangular
Water Cooled
Center through Bottom

They have been placed in single rows, double rows, triple rows, and diagonally around the entire preheating zone as in the Poumay cupola and the balanced draft developed by the British Cast Iron Research Institute.

Preheated blast has long been considered.

The height of the cupola has been

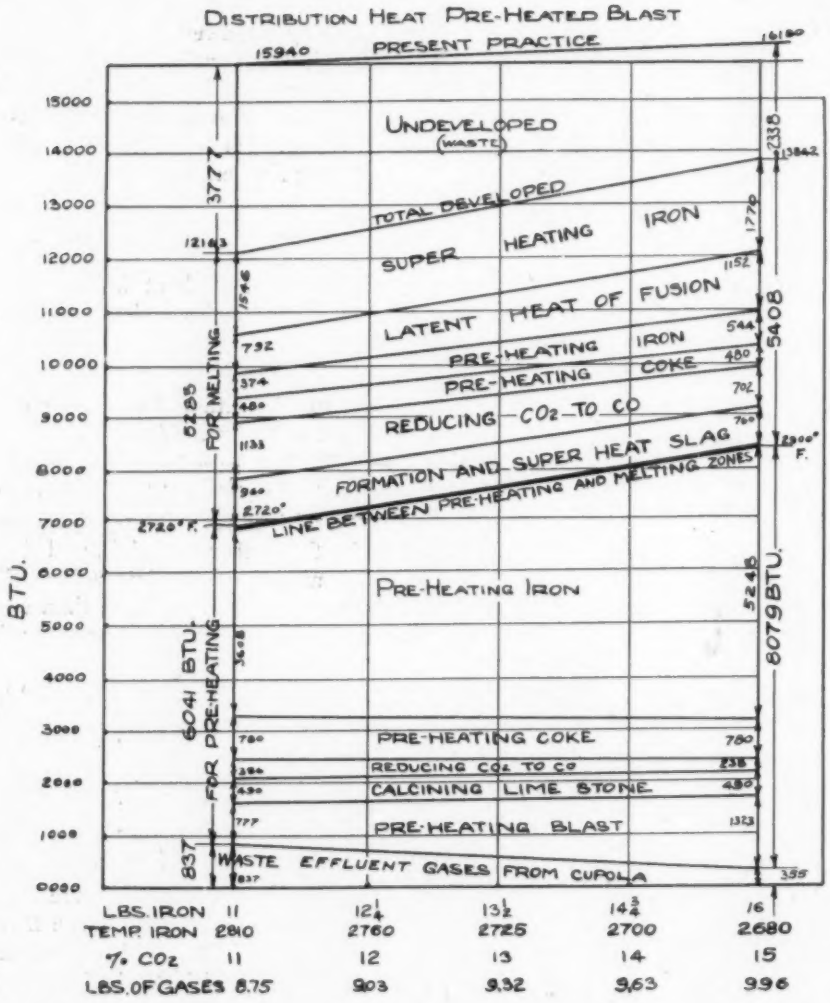


Fig. 9—Chart shows distribution of heat in a pre-heated blast furnace.

increased experimentally, hoping to reduce the temperature of the stack gases.

The effect of changing the lining from a straight cupola to extreme boshing has been observed.

Most of these ideas have been more or less disappointing, at least the improvement has not been sufficiently definite to cause their general use to the exclusion of other types. They are all worthy subjects of discussion from an operational standpoint, which is outside the scope of this paper.

Before attempting any radical departure from present standards, a clear-cut analysis should be made of the amount and reasons for present wastes and to locate definitely the bottleneck in the melting zone which fixes melting limitations.

This analysis will point out a deficiency of heat below the bottleneck and an excess above, consequently any readjustment of the heat should be made on the basis of drawing upon the excess heat above the bottleneck and transferring it to point of deficiency below the bottleneck.

The criterion for deficiency in heat for melting purposes is when the temperature of the gases is not sufficiently high to melt the iron with dispatch. It is clearly evident that all the heat in the world at a temperature below 2000° F. would not melt a pound of iron.

For the purpose of analysis it is necessary to separate the heat into two parts:

(1) First, that part which can be used for melting iron. This is represented by the heat contained in the gases of combustion above a temperature of, say, 2500° F., which is approximately the low limit for prompt melting.

(2) Second, that part which can be used for preheating purposes only. This is represented by the heat remaining in the gases after their temperature is reduced to 2500° F.

The distribution of the heat available for melting, and the part left for preheating, is shown in Tables 9 and 10, from which a graphical diagram is made to visualize more easily the critical limitation at the melting zone in cold blast cupola practice for all coke ratios. See Fig. 8.

Figure 8 is important from the standpoint of a study of the basis

Table 11

HEAT DEVELOPED PER LB. OF COKE

Lb. Iron	CO_2, per cent	CO_2, lb.	CO_2 Per Lb. Iron	Btu. Developed	Sensible Heat—Blast	Sensible Heat—Coke	Total Btu. Combustion Zone	Undeveloped	Total Potential
7	9	1.09	.15	8704	0	1260	9964	4391	14355
11	13	1.75	.15	9999	0	1260	11259	3096	14355

HEAT DISTRIBUTION PER LB. OF COKE

Lb. Iron	CO_2, per cent	For Melting	For Preheat	Sensible Heat, Effluent Gases	Btu. Not Developed	Total
7	9	4370	3896	1698	4391	14355
11	13	4369	5078	1811	3097	14355

Table 12

PROPOSED HOT BLAST PRACTICE

Lb. Iron	CO_2, per cent	CO_2, lb.	CO_2 Per Lb. Iron	Sensible Heat—Btu. Developed	Sensible Heat—In Preheated Blast	Sensible Heat—In Coke	Total in Combustion Zone	Undeveloped	Total Potential
11	11	1.40	.13	9318	1600	1260	12178	3777	15955
16	15	2.12	.13	10757	1800	1260	13817	2338	16155

for controlling cupola efficiency. For the purpose of analysis, the distribution of potential heat from a pound of coke is divided into four parts: first, that portion which is undeveloped; second, the portion used for melting purposes; third, the portion used for preheating iron, coke, limestone; and fourth, amount of waste heat in the effluent gases.

Summarizing these quantities, the following extremes are found in the range of 7 lb. to 11 lb. of iron melted per lb. of coke used. See Table 11.

In Table 11 the limited amount of heat available for melting is the outstanding feature. This is because, as previously mentioned, the only heat available for melting is that part of the heat in the gases in the melting zone having a temperature above 2500° F.

When the temperature of the gases is reduced to 2500° F. or thereabouts, its usefulness for melting ceases and can only be used for preheating iron, coke and limestone, and expelling moisture from the charges.

Table 11 indicates that there is a very liberal amount of heat left over after preheating has been accomplished, as indicated by the temperature of 800° F. in the effluent gases which are delivered to the atmosphere.

If the equivalent of this waste could be carried to the combustion zone and added to the amount of heat developed by combustion, the temperature of the gases in the combustion zone would be raised approximately 600° F. and practically the entire amount of additional heat would be used for melting purposes.

The possibilities of a decided increase in cupola efficiency is thus very obvious and strongly indicates that the following economies are not too difficult of attainment:

First, the development of 25 per cent more total heat from a pound of coke than is ordinarily produced;

Second, the reduction of the coke charge 33⅓ per cent without reducing the iron charge; or

Third, an increase in the iron charge 50 per cent without increasing the coke charge.

To accomplish this, it is evident that the amount of heat used for melting must be increased 50 per cent, or, say:

4400 Btu. + 2200 Btu. = 6600 Btu.

The question then arises, How can the additional 2200 Btu. be introduced into the Combustion Zone in addition to the amount of heat ordinarily developed?

Under ordinary practice, an amount of air almost equal to the weight of the iron is introduced into the Combustion Zone at atmospheric temperature and leaves the cupola in the effluent gases at a temperature of approximately 800° F., representing a total loss as far as this amount of heat is concerned.

If the temperature of the air of combustion could be increased 800° F. prior to entering the Combustion Zone, this entire waste would be prevented.

Referring to Table 34 on page 536, in the column under 800° F. Preheat, it will be noted that the Btu. in the air required for combustion of one lb. of coke for stages 11 per cent to 15 per cent CO_2 is 1600 Btu. and 1800 Btu. respectively, which corresponds with the heat loss in the effluent gases in Table 11.

This additional heat is produced without additional coke or air, hence

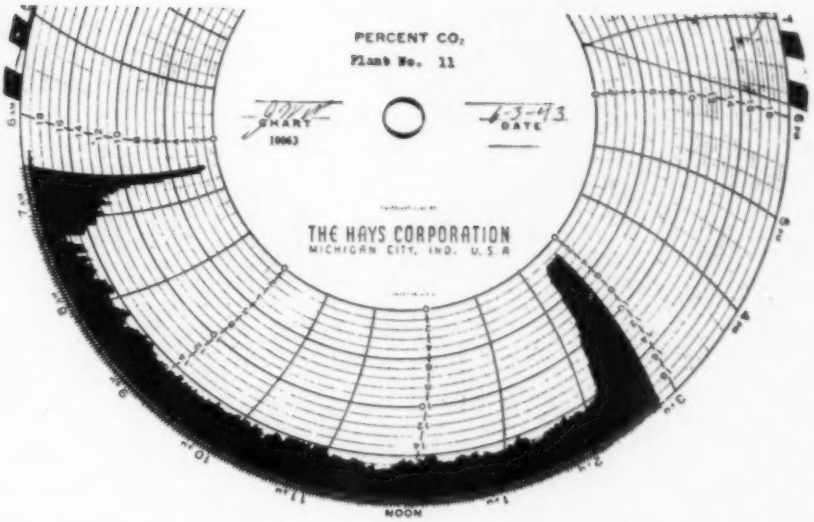
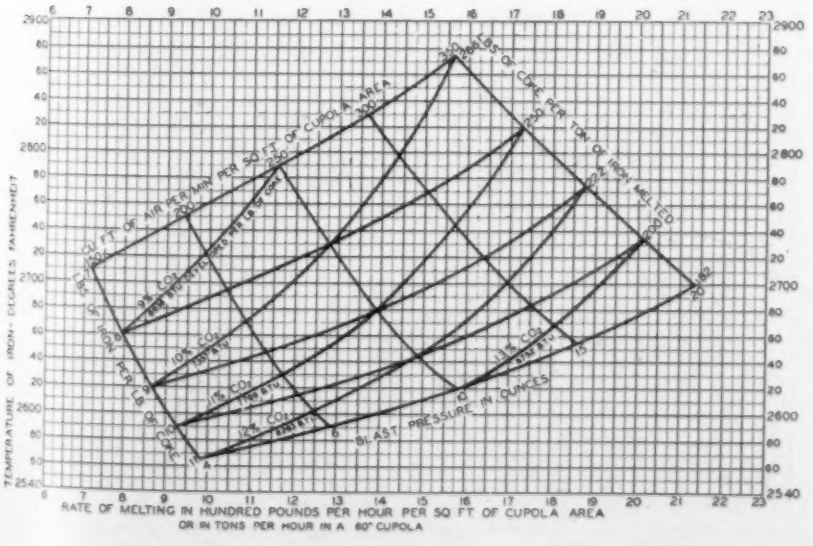


Fig. 11—Chart from an automatic CO_2 recorder showing 85 per cent combustion efficiency in pre-heated blast cupola practice.

Fig. 10—Usual range of variations in cupola practice with blast pre-heated to 800° F. Showing the relation of iron to coke ratios, varying blast pressures, and air volumes to rate of melting, temperature of the iron, efficiency, and rate of combustion.



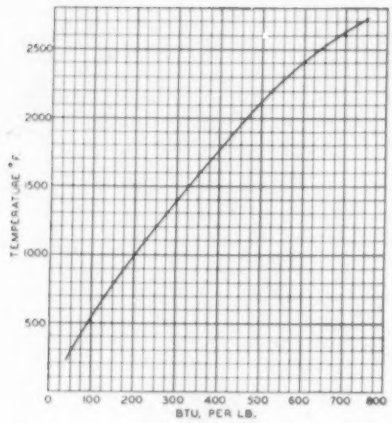


Fig. 12—Sensible heat content of slag above 32° F.

there is no deduction for forming slag from the ash of coke, for heat carried off in the effluent gases, or for loss of efficiency resulting from aqueous vapor in the combustion zone. Consequently all the additional heat is used for melting more iron and for increasing the temperature of the combustion, reducing and melting zones.

Further beneficial action of the increased heat is a decrease in the CO_2 concentration in proportion to the amount of heat developed. The oxidizing agent in melting iron is the CO_2 in the gases of combustion, therefore the greater the concentration of CO_2 the greater oxidation of silicon, manganese, iron, etc.

Inasmuch as the higher concentration of CO_2 indicates a greater development of heat the objective in securing the highest cupola efficiency is to develop the highest possible percentage of CO_2 without increasing normal oxidation.

It is shown in Table 11 that for each pound of iron melted 15/100ths of a lb. of CO_2 passes through the melting zone, and inasmuch as the problem is to increase the number of pounds melted per lb. of coke consumed it will be possible to increase the concentration of CO_2 by 2 per cent without increasing the tendency towards oxidation. In fact, the CO_2 per lb. of iron melted will be reduced to 0.13 lb.

If the percentage of CO_2 is increased from 9 per cent in the cold blast to 11 per cent in the hot blast cupola, it will be found by referring to Table 26 on page 533 that the amount of heat developed per

Table 13

HEAT BALANCE—800° F. PREHEATED BLAST

Basis: 1 Lb. Coke; Varying Iron Ratios; Volume Air Constant, 200 cfm. per sq. ft. cupola area.

Lb. Iron Melted—1 Lb. Coke	11	12¼	13½	14¾	16
Lb. Coke					
For melting 1 Ton Iron	181.8	163.3	148.1	135.6	125.0
For combustion to CO_2	129.4	120.3	113.1	107.4	102.7
For reduction CO_2 to CO	52.4	43.0	35.0	28.2	22.3
Rate of disappearance, lb. per hr.	123.4	119.1	115.0	110.9	106.9
Rate of melting, tons per hr.	.679	.730	.776	.818	.855
Per cent CO_2 in effluent gases	11.0	12.0	13.0	14.0	15.0
Lb. Air					
For each lb. coke charged	7.85	8.13	8.42	8.73	9.06
For each lb. coke burned	11.03	11.03	11.03	11.03	11.03
For one ton iron melted	1427	1328	1247	1184	1133
Lb. Gases					
Per lb. coke disappeared	8.75	9.03	9.32	9.63	9.96
Per ton iron melted	1591	1475	1380	1306	1245
Disappearance 1 Lb. Coke					
Per cent burned to CO_2 (A+½B)	71.15	73.68	76.35	79.17	82.14
Per cent reduction CO_2 to CO (½B)	28.85	26.32	23.65	20.83	17.86
Total per cent	100.00	100.00	100.00	100.00	100.00
Btu. Developed in Combustion Zone					
From combustion (A+½B)	9318	9649	9998	10367	10757
Sensible heat preheated (A+½B)	900	930	960	1000	1035
Btu. in preheated air, 800° F.	1585	1640	1700	1760	1825
Total Btu. developed in Combustion Zone	11803	12219	12658	13127	13617
Sensible heat Reduction Zone (½B)	360	330	300	260	225
Total Heat developed below Melting Zone	12163	12549	12958	13387	13842
Temperature Gases in Combustion Zone (Theoretical)	4450	4450	4450	4450	4450
Btu. Loss Converting CO_2 to CO	1519	1386	1245	1097	940
Net Btu. developed 1 Lb. Coke	7799	8263	8753	9270	9817
Potential Heat not developed	5296	4832	4342	3825	3278
Total Potential Btu. 1 Lb. Coke	13095	13095	13095	13095	13095
Add Sensible Heat	1260	1260	1260	1260	1260
Total Below Melting Zone	14355	14355	14355	14355	14355
Temperature Fluid Iron in Melting Zone, 2200° F.	2200	2200	2200	2200	2200
Temperature Iron as Tapped	2810	2760	2725	2700	2680
Degrees Superheat	610	560	525	500	480
Btu. in Iron					
In Solid Iron in Melting Zone	3608	4018	4428	4838	5248
In Fluid Iron in Melting Zone	4774	5317	5859	6402	6944
In Iron as Tapped	6320	6897	7492	8101	8714
In Superheat	1546	1580	1633	1699	1770
HEAT TRANSFER					
Btu. Transferred from Gases in Combustion Zone					
Superheating Iron	515	523	544	566	590
Formation and Superheating Slag	530	495	460	425	390
Total Btu.	1045	1018	1004	991	980
Temperature Gases Top of Zone, ° F. in Reducing Zone	3890	3900	3920	3930	3940
Superheating Iron	1031	1047	1089	1133	1180
Preheating Coke	235	235	235	235	235
Superheating Slag	270	255	240	225	210
Reducing CO_2 to CO	1133	1034	930	818	702
Total Btu. Transferred	2669	2571	2494	2411	2327
Temperature of Gases Top of Zone, ° F. in Melting Zone	3250	3300	3350	3390	3440
Latent Heat Fusion	792	882	972	1062	1152
Preheating Iron—200° F.	374	417	459	502	544
Preheating Coke—500° F.	245	245	245	245	245
Formation and Superheating Slag	160	160	160	160	160
Total Btu. Transferred	1571	1704	1836	1969	2101
Temperature of Gases Top of Zone, ° F.	2750	2780	2810	2840	2870
Total Heat Below Preheating Zone	5285	5293	5334	5371	5408
Total Btu. Remaining in Preheating Zone	6878	7256	7624	8016	8434
Btu. in Preheating Zone					
Preheating Iron to 2000° F.	3608	4018	4428	4838	5248
Preheating Coke to 2000° F.	780	780	780	780	780
Calcining Limestone	490	490	490	490	490
Reducing CO_2 to CO	386	352	315	279	238
Btu. to Preheater	777	875	994	1139	1323
Total Btu. Transferred	6041	6515	7007	7526	8079
Btu. in Effluent Gases	837	741	617	490	355
Temperature, ° F.	610	550	480	400	320

Table 14

HEAT BALANCE—800° F. PREHEATED BLAST

Basis: 1 Lb. Iron; Varying Coke Ratios;
Volume Air Constant, 200 cfm. per sq. ft. cupola area.

Lb. Iron Melted—1 Lb. Coke	11	12¼	13½	14¾	16
Lb. Coke					
For melting 1 ton iron	181.8	163.3	148.1	135.6	125.0
For melting 1 lb. iron	.0909	.0816	.0740	.0678	.0625
Per cent for combustion to CO_2	71.15	73.68	76.35	79.17	82.14
Per cent reduction CO_2 to CO	28.85	26.32	23.65	20.83	17.86
Per sq. ft. Cupola Area					
Lb. coke used per hr.	123.4	119.1	115.0	110.9	106.9
Tons melted per hr.	.679	.730	.776	.818	.853
Per cent CO_2 in effluent gases	11.0	12.0	13.0	14.0	15.0
Lb. Air					
Per lb. iron melted	.714	.664	.624	.592	.566
Per lb. coke burned	1.00	.900	.815	.796	.688
Per ton iron melted	1427	1328	1247	1184	1133
Lb. Gases					
From coke used per lb. iron	.795	.737	.690	.655	.623
From coke used per ton iron	1591	1475	1380	1306	1245
Disappearance of Coke Per Lb. Iron					
Lb. burned to CO_2 (A+½B)	.0647	.0601	.0565	.0537	.0522
Lb. reduction CO_2 to CO (½+B)	.0262	.0215	.0175	.0141	.0113
Total	.0909	.0816	.0740	.0678	.0635
Btu. Developed in Combustion Zone					
From combustion (A+½B)	847	787	741	703	672
Sensible heat coke (A+½B)	82	76	71	68	65
Sensible heat blast, 800° F.	144	134	126	119	114
Total Btu. developed in Combustion Zone	1073	997	938	890	851
Sensible heat Reducing Zone (½B)	33	27	22	18	14
Total heat developed below Melting Zone	1106	1024	960	908	865
Temperature Gases in Combustion Zone	4450	4450	4450	4450	4450
Btu. Loss Reducing CO_2 to CO	138	113	92	74	59
Net Btu. developed from coke per lb. iron	709	694	648	629	614
Potential Heat not developed:					
(Per lb. iron melted)	482	394	322	259	205
Potential Btu. in Coke per lb. iron	1191	1068	970	888	819
Add Sensible heat in coke	114	103	93	85	79
Total Btu. below Melting Zone per lb. iron	1305	1171	1063	973	898
Temperature Fluid Iron in Melting Zone	2200	2200	2200	2200	2200
Temperature Iron as Tapped	2810	2760	2725	2700	2680
Degrees of Superheat		560	525	500	480
Btu. in 1 Lb. Iron					
In Solid Iron in Melting Zone	328	328	328	328	328
In Fluid Iron in Melting Zone	434	434	434	434	434
In Iron as Tapped	574	563	555	549	545
In Superheat	140	129	121	115	111
HEAT TRANSFER TO 1 LB. IRON					
Btu. Transferred from Gases in Combustion Zone					
Superheating Iron	46	43	40	38	37
Formation and Superheating Slag	48	40	34	29	24
Total Btu.	95	83	74	67	61
Temperature Gases Top of Zone, ° F. in Reducing Zone	3890	3900	3920	3930	3940
Superheating Iron	94	85	81	76	74
Preheating Coke	22	19	17	16	15
Superheating Slag	24	21	17	15	13
Reducing CO_2 to CO	103	57	46	39	44
Total Btu. Transferred	243	183	162	146	145
Temperature Gases Top of Zone, ° F. in Melting Zone	3250	3300	3350	3390	3440
Latent Heat Fusion	72	72	72	72	72
Preheating Iron—200° F.	34	34	34	34	34
Preheating Coke—500° F.	22	20	18	16	15
Formation and Superheating Slag	15	13	12	11	10
Total Btu. Transferred	143	139	136	133	131
Temperature of Gases Top of Zone, ° F.	2750	2780	2810	2840	2570
Total Heat Below Preheating Zone	480	405	371	352	338
Total Btu. Remaining in Preheating Zone	625	592	609	553	527
Btu. in Preheating Zone					
Preheating Iron to 2000° F.	328	328	328	328	328
Preheating Coke to 2000° F.	71	64	58	52	49
Calcining Limestone	44	40	36	33	31
Reducing CO_2 to CO	35	29	23	18	15
Btu. to Preheater	70	71	74	79	82
Total Btu. Transferred	548	532	519	510	505
Btu. in Effluent Gases	74	61	46	33	22
Temperature, ° F.	610	550	480	400	320

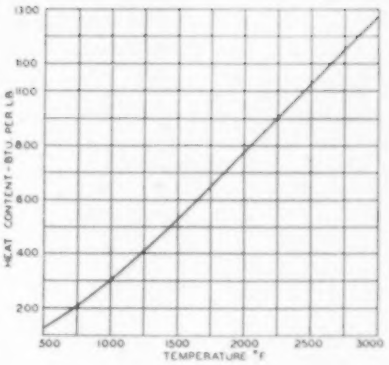


Fig. 13—Sensible heat content of carbon above 70° F.

lb. of coke will be increased 600 Btu., and if the percentage of 13 per cent in the cold blast is increased to 15 per cent in the hot blast, the amount of heat developed per lb. of coke will be increased 750 Btu. Table 12 shows data on Proposed Hot Blast Practice.

From the above data the detailed heat balances in Tables 13 and 14 are prepared.

The foregoing data in Table 15 is shown in graphic form in Fig. 9.

Heat Balances

The heat balances for preheated blast in Table 13 are shown in Fig. 10, which indicates the relationship for all coke and iron ratios and various blast volumes to combustion efficiency, temperature of iron at the tapping hole, and rate of melting.

Figure 11, a reproduction of a chart from an automatic CO_2 recorder indicates the high efficiency of combustion permissible in preheated blast practice.

In the foregoing tables the heat balances have been made up from two standpoints:

First—Basis 1 lb. Coke. It is often convenient to follow the various phases of 1 lb. of coke as it passes through the cupola in the development of heat in the combustion zone and its distribution to the various purposes for which it is used. Hence Tables 7 and 13 have been prepared on the basis of 1 lb. of coke, with varying iron ratios and constant air volumes. Other balances have been prepared for each air volume shown in Fig. 1. These, however, are too voluminous to be incorporated into this paper.

Second—Basis 1 lb. Iron. It is

instructive to note the relation of the heat absorbed by 1 lb. iron to the total heat developed, also its relationship to the heat absorbed for miscellaneous purposes. Hence Tables 8 and 14 have been prepared on the basis of 1 lb. of iron, with varying coke ratios and constant air volume.

To clarify further the heat balances recorded in Tables 7 and 13, and to make their application more definite, the quantities representing the distribution of heat have been broken down and classified in Tables 10 and 15 according to the purposes for which the heat is used. The total heat developed is separated into two parts: First, for melting, and Second, for preheating.

This division is of fundamental importance in developing higher overall thermal efficiencies. The objective throughout in the preparation of heat balances and summaries of distribution is to point definitely to the fundamentals which regulate cupola efficiencies and point out why certain cupolas are operating at a 30 per cent efficiency when 70 per cent is easily within the range of possibility.

Summary

The following summary in Table 16 drawn up from the foregoing tables has been prepared as a view of the entire field of cupola practice, showing how the various efficiencies are built up and why it is essential to concentrate as much heat as possible below the Melting Zone.

In this summary it is assumed that 2400° F. is the lowest temperature suitable for melting purposes, which occurs when the coke ratio is at the highest point or vice versa when the pounds of iron melted per lb. of coke is the lowest; and that for each additional pound of iron melted per pound of coke, the temperature of the gases should be increased 50 degrees.

This, of course, is an approximation, but seems to fit actual experience. There is no significance to the units and tens used in the Btu. quantities, but as a matter of mathematical convenience in balancing calculations they have been retained. See Table 16.

This summary is the final affirmative answer to the questions raised on p. 523:

Table 15

DISTRIBUTION OF HEAT TRANSFER THROUGH THE CUPOLA

11 Lb. Iron — 1 Lb. Coke, air 200 cfm. preheated to 800° F.

Potential heat ..... 13095 Btu.
11 per cent CO_2 develops..... 9318 + 1260 + 1585 = 12,163 Btu. for melting.
Undeveloped ..... 3777 Btu.

Transfer of Btu. in Iron	*Combustion Zone*	*Reducing*	*Melting*	*Total Below Preheat*	*Preheat*	*Total*
Preheat			374	374	3608	3982
Latent heat fusion			792	792		792
Superheat	515	1031		1546		1546
Total	515	1031	1166	2712	3608	6320
Coke Preheat		235	245	480	780	1260
Slag—Formation and Superheat	530	270	160	960		960
Reduce CO_2 to CO		1133		1133	386	1519
Calcine Limestone					490	490
To Preheat Blast					777	777
Effluent Gases					837	837
Total	1045	2269	1571	5285	6878	12163

12¼ Lb. Iron — 1 Lb. Coke, air 200 cfm. preheated to 800° F.

Potential heat ..... 13095 Btu.
12 per cent CO_2 develops..... 9649 + 1260 + 1640 = 12,549 Btu. for melting.
Undeveloped ..... 3446 Btu.

Transfer of Btu. in Iron	*Combustion Zone*	*Reducing*	*Melting*	*Total Below Preheat*	*Preheat*	*Total*
Preheat			417	417	4018	4435
Latent heat fusion			882	882		882
Superheat	523	1047		1570		1570
Total	523	1047	1299	2869	4018	6887
Coke Preheat		235	245	480	780	1260
Slag—Formation and Superheat	495	255	160	910		910
Reduce CO_2 to CO		1034		1034	352	1386
Calcine Limestone					490	490
To Preheat Blast					875	875
Effluent Gases					741	741
Total	1018	2571	1704	5293	7256	12549

13½ Lb. Iron — 1 Lb. Coke, air 200 cfm. preheated to 800° F.

Potential heat ..... 13095 Btu.
13 per cent CO_2 develops..... 9998 + 1260 + 1700 = 12,958 Btu. for melting.
Undeveloped ..... 3097 Btu.

Transfer of Btu. in Iron	*Combustion Zone*	*Reducing*	*Melting*	*Total Below Preheat*	*Preheat*	*Total*
Preheat			459	459	4428	4887
Latent heat fusion			972	972		972
Superheat	544	1089		1633		1633
Total	544	1089	1431	3064	4428	7492
Coke Preheat		235	245	480	780	1260
Slag—Formation and Superheat	460	240	160	860		860
Reduce CO_2 to CO		930		930	315	1245
Calcine Limestone					490	490
To Preheat Blast					994	994
Effluent Gases					617	617
Total	1004	2494	1836	5334	7624	12958

14¾ Lb. Iron — 1 Lb. Coke, air 200 cfm. preheated to 800° F.

Potential heat ..... 13095 Btu.
14 per cent CO_2 develops..... 10367 + 1260 + 1760 = 13,387 Btu. for melting.
Undeveloped ..... 2728 Btu.

Transfer of Btu. in Iron	*Combustion Zone*	*Reducing*	*Melting*	*Total Below Preheat*	*Preheat*	*Total*
Preheat			502	502	4838	5340
Latent heat fusion			1062	1062		1062
Superheat	566	1133		1699		1699
Total	566	1133	1564	3263	4838	8101
Coke Preheat		235	245	480	780	1260
Slag—Formation and Superheat	425	225	160	810		810
Reduce CO_2 to CO		818		818	279	1097
Calcine Limestone					490	490
To Preheat Blast					1139	1139
Effluent Gases					490	490
Total	991	2411	1969	5371	8016	13387

16 Lb. Iron — 1 Lb. Coke, air 200 cfm. preheated to 800° F.

Potential heat ..... 13095 Btu.
15 per cent CO_2 develops..... 10757 + 1260 + 1825 = 13,842 Btu. for melting.
Undeveloped ..... 2338 Btu.

Transfer of Btu. in Iron	*Combustion Zone*	*Reducing*	*Melting*	*Total Below Preheat*	*Preheat*	*Total*
Preheat			544	544	5248	5792
Latent heat fusion			1152	1152		1152
Superheat	590	1180		1770		1770
Total	590	1180	1696	3466	5248	8714
Coke Preheat		235	245	480	780	1260
Slag—Formation and Superheat	390	210	160	760		760
Reduce CO_2 to CO		702		702	238	940
Calcine Limestone					490	490
To Preheat Blast					1323	1323
Effluent Gases					355	355
Total	980	2327	2101	5468	8434	13842

Table 16

SUMMARY OF CONDITIONS WHICH CONTROL CUPOLA EFFICIENCIES

Lb. Iron Melted, per Lb. Coke	Cold Blast			Preheated Blast		
	7	9	11	11	13½	16
CO_2 in Gases, per cent	9.0	11.0	13.0	11.0	13.0	15.0
Lb. gases per lb. coke	8.23	8.75	9.33	8.75	9.33	9.96
Specific heat at 2500° F.	.281	.283	.285	.283	.285	.287
Btu. developed by combustion	8704	9318	9998	9318	9998	10757
Sensible heat, 1 lb. coke (3000° F.)	1260	1260	1260	1260	1260	1260
Sensible heat, Preheated Blast (800° F.	—	—	—	1585	1700	1825
Total heat developed	9964	10578	11258	12163	12958	13844
Temperature, ° F.	3900	3860	3800	4360	4320	4290
Btu. not developed	4391	3777	3097	3777	3097	2338
Potential Heat, 1 lb. Coke	13095	13095	13095	13095	13095	13095
Required Temperature Gas Top Melting Zone	2400	2500	2600	2650	2750	2850
Gas, lb. ° F.	19752	21875	24258	23187	25657	28386
Specific heat	.278	.282	.286	.286	.291	.295
Btu. not available for melting	5491	6190	6962	6631	7466	8374
Difference available for melting	4473	4388	4296	5532	5492	5468
Btu. shown in Tables 7 and 13	4370	4342	4369	5285	5334	5408
Btu. per lb. gas developed	1210	1210	1210	1390	1390	1390
Available for melting	543	502	462	632	589	549
Not available for melting	667	708	748	758	801	841
Distribution of Heat						
Preheat Iron	2534	3258	3982	3982	4887	5792
Latent Heat Fusion	504	648	792	792	972	1152
Superheat	952	999	1067	1546	1633	1770
Total Btu. Absorbed by Iron	3990	4905	5841	6320	7492	8714
Temperature Iron at Tapping Hole	2790	2680	2620	2810	2725	2680
Btu. for Developing Slag	880	780	730	960	860	760
Btu. for Calcining Limestone	370	370	370	490	490	490
Btu. for Reducing CO_2 to CO	1766	1519	1246	1519	1246	940
Btu. for Preheating Coke	1260	1260	1260	1260	1260	1260
Btu. for Preheating Blast	—	—	—	777	994	1323
Btu. in Effluent Gases	1698	1744	1811	837	617	355
Total Exclusive of Iron	5974	5673	5417	5843	5467	5128
Efficiency of Combustion	66.47	71.15	76.35	71.15	76.35	82.14
Efficiency of Absorption	45.83	52.64	58.41	67.83	74.93	81.00
Overall Thermal Efficiency	30.5	37.4	44.6	48.3	57.2	66.5

1. Can the heat developed from a pound of coke be increased 25 per cent?
2. Can the coke charge be reduced 33⅓ per cent without reducing the iron charge; or
3. Can the iron charge be increased 50 per cent without increasing the coke charge?

A good average in ordinary practice is 9 lb. iron melted per lb. of coke. This is to be increased to 13½ lb.

The summary indicates:

1. Btu. developed—1 lb. coke; 9 to 1 ratio..... 9,318
   Btu. developed—1 lb. coke; 13½ to 1 ratio..11,698

   Increase 2,380
   Per cent Increase..... 26.0
2. Lb. coke per ton melted; 9 to 1 ratio 222.2
   Lb. coke per ton melted; 13½ to 1 ratio 148.1

   Saving 74.1
   Per cent Saving..... 33.4
3. Iron per lb. coke—9 to 1 ratio..... 9
   Iron per lb. coke—13½ to 1 ratio..... 13½

   Increase 4½
   Per cent Increase..... 50.0

These results indicate that from a thermal standpoint there is no reason for not using the increased efficiencies.

Table 17

SOME DATA ON HOT BLAST CUPOLA OPERATION

Plant No.	*Average Daily Tonnage*	*Cupola Diameter, in.*	*Tons Melted Per Hr.*	*Lb. Coke Used for Melting One Ton**
1	251	78	24.9	150
2	111	57	13.4	147
3	128	67	18.1	148
4	222	74	23.5	156
5	56	66	17.4	146
6	197	76	22.9	155
7	143	66	18.2	153
8	74	64	21.0	146
9	157	71	18.5	156
10	116	60	14.5	149
11	123	60	16.7	134
Total:	1578 tons			

Average Saving Per Day
80 lb. coke per ton melted, at 65 cents = $1,025.

*Exclusive of Bed.

Savings Are Substantial

An objection may be raised that the foregoing analysis of the possibility of improvement in cupola efficiency is only a theory. However, if it is not practical it is not theoretical. The proof of the pudding is in the eating.

The conclusions seemed to rest on a sufficiently firm foundation to the executives of a large chilled car wheel industry, melting above 400,000 tons of iron annually, to authorize the expenditure of $100,000 for research regarding the advantages of preheated blast, on a productive basis.

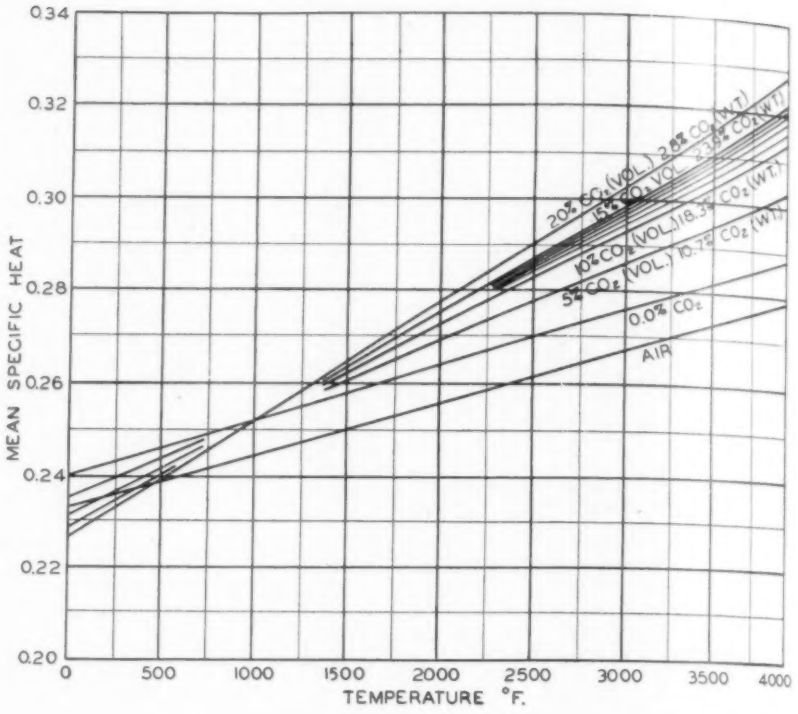
The program finally resulted in converting eleven cupolas from cold blast to preheated blast which have been in operation ten years or more. Recent results are shown in Table 17.

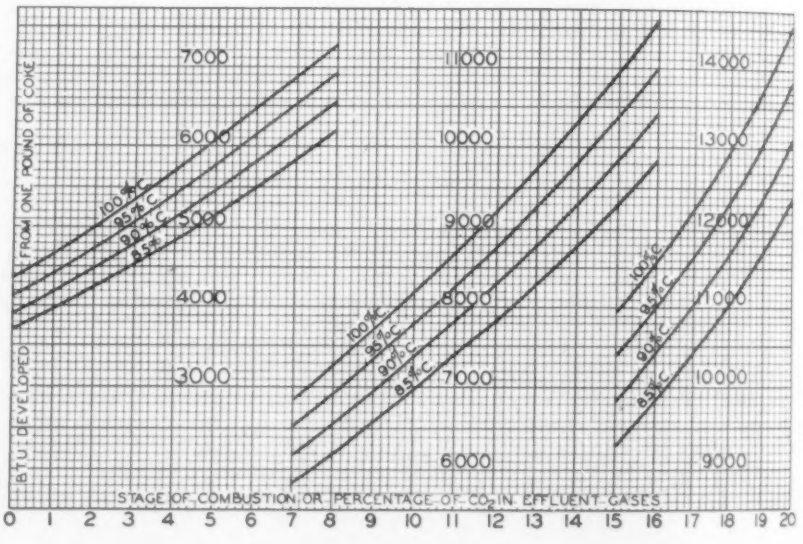
A report of the auditor of the company operating these cupolas shows that since starting the use of the preheated blast, a saving in fuel has been made of $1,500,000, after writing off $260,000 for depreciation of equipment incidental to preheating the blast. At the present rate

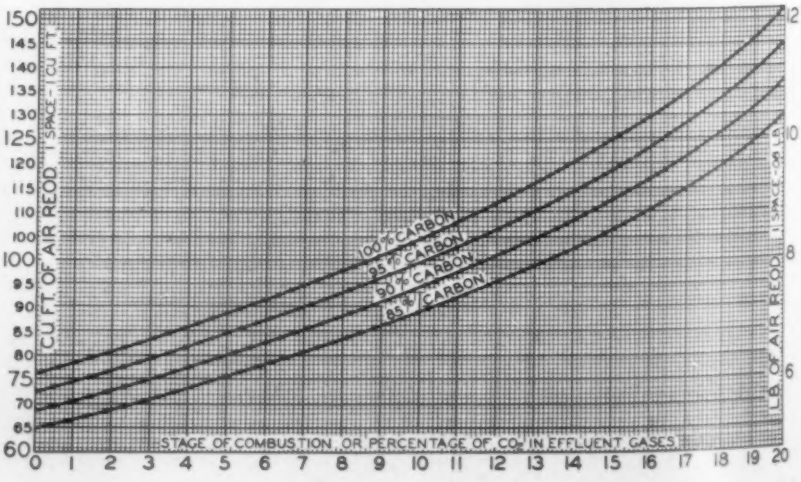
Fig. 14 (Top)—Relationship of Mean Specific Heat to the temperature of gases of combustion.

Fig. 15 (Center)—Btu. developed from combustion of 1 lb. of cokes of various carbon contents and for various stages of combustion.

Fig. 16 (Bottom)—Cu. ft. and lb. of air required for combustion of 1 lb. of cokes of various carbon contents and for various stages of combustion.







of output, 1,500 tons per day, the saving to the company from these eleven cupolas amounts to $300,000 per year.

The variations in coke ratios in the eleven cupolas reported above indicate that personal preference is still the controlling item in fixing the limits of coke ratios. The lowest ratio of 134 lb. per ton melted is not because the blast was preheated to a higher temperature, for the reverse is the case, but a higher stage of combustion is being used, which was the purpose in introducing Fig. 11, in which the CO_2 was developed at 16 per cent in comparison with Fig. 6, where the development of CO_2 was 13 to 14 per cent.

Neither the degree of preheat nor the stage of combustion alone fixes the efficiency of operations; therefore important regulation can still be made regarding the rate of combustion and the rate of melting in comparison with cupola efficiency.

Preheated Blast Used

Many cupolas with preheated blast have been installed in industries other than the chilled wheel industry, the combined capacities of which are in excess of 2,000 tons daily. The capacities of these cupolas range from two tons to 30 tons per hr.

In the piston ring industry, where the daily output is small but quality of product is the prime objective, preheated blast has been used continuously for a period of 15 years as an aid in meeting strict Government specifications.

In cupolas of larger capacity the economy in fuel becomes the most prominent factor; hence quality of product and economy in fuel are the two items which invite the use of preheated blast. These varied installations prove the essential cor-

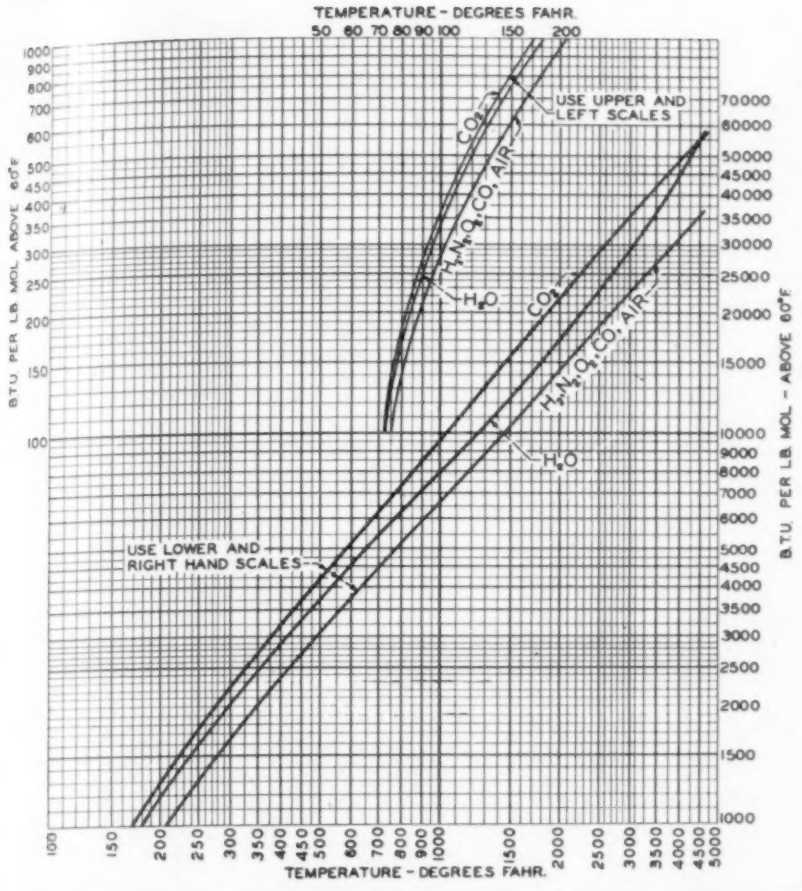


Fig. 17—Chart shows sensible heat content of gases above 60° F. in Btu. per lb. mol. To convert to Btu. per lb. divide value chart by molecular weight. Molecular weights for the various gases are as follows: CO_2—44; H_2O—18; H_2—2; CO—28; O_2—32. (After Haslam and Russell.)

rectness of the thermal balances recorded in Tables 13 and 14; and also prove that the regenerative feature is just as essential to the higher efficiencies in cupola practice as in blast furnace and open hearth practice.

No attempt has yet been made to approach the maximum efficiency in the cupola. Blast furnaces would not be constructed today without provision for preheating the blast to 1,400 or 1,600° F.; open hearth furnaces for melting steel could not be operated without the regenerative feature.

As applied to cupolas, regeneration consists in taking the potential and sensible heat in the spent, efflu-

Table 18

HEAT REQUIRED TO RAISE 1 LB. OF WATER 1° AT VARIOUS TEMPERATURE

	Temperature Range, ° F.		*Btu.*		*Calories*
Standard	Btu. from 39° to 40°	=	1.0000	=	253.4 Standard
Normal	Btu. from 60° to 61°	=	0.9945	=	252.0 Normal
Normal	Btu. from 70° to 71°	=	0.9935	=	251.7 Normal
Mean	Btu. $\frac{32° \text{ to } 212°}{180°}$	=	0.9947	=	252.0 Mean

Table 19

SPECIFIC HEAT OF SOME GASES

Gas	*Specific Heat 0° to 3600° F.*	*Specific Heat 3600° to 7200° F.*
Nitrogen	.2405 + .0000238t	.2044 + .000064t
Oxygen	.2104 + .0000208t	.1788 + .00006t
CO_2	.1900 + .0001222t	.41 + .000038t
CO	.2405 + .0000238t	.2044 + .000064t
Air	.2336 + .0000232t	
Water Vapor	.42 + .000206t	.59 + .00024t

Table 20

MEAN SPECIFIC HEAT OF SOME GASES

Gas	*Mean Specific Heat 0° to 3600° F.*	*Mean Specific Heat 3600° to 7200° F.*
Nitrogen	.2405 + .0000119t	.2044 + .000032t
Oxygen	.2104 + .0000104t	.1788 + .00003t
CO_2	.1900 + .0000611t	.41 + .000019t
CO	.2405 + .0000119t	.2044 + .000032t
Air	.2336 + .0000116t	
Water Vapor	.42 + .000103t	.59 + .00012t

Table 21

BTU. REQUIRED TO RAISE 1 LB. OF GAS FROM 0° TO t°

Gas	*0° to 3600° F.*
Nitrogen	.2405t + .0000119t²
Oxygen	.2401t + .0000104t²
CO_2	.1900t + .00006111t²
CO	.2405t + .0000119t²
Air	.2336t + .0000116t²

Table 22

BTU. REQUIRED TO RAISE 1 LB. OF GAS FROM t^1 TO t

Gas	*0° to 3600° F.*
Nitrogen	$.2405t^1\text{-}t + .0000119\ (t^{1(2)}\text{-}t^2)$
Oxygen	$.2104t^1\text{-}t + .0000104\ (t^{1(2)}\text{-}t^2)$
CO_2	$.1900t^1\text{-}t + .0000611\ (t^{1(2)}\text{-}t^2)$
CO	$.2405t^1\text{-}t + .0000119\ (t^{1(2)}\text{-}t^2)$
Air	$.2336t^1\text{-}t + .0000116\ (t^{1(2)}\text{-}t^2)$

Table 23

MEAN SPECIFIC HEAT OF SOME GASES BETWEEN TWO SPECIFIED TEMPERATURES

Gas	*0° to 3600° F.*
Nitrogen	$.2405 + .0000119\ (t^1+t)$
Oxygen	$.2104 + .0000104\ (t^1+t)$
CO_2	$.1900 + .0000611\ (t^1+t)$
CO	$.2405 + .0000119\ (t^1+t)$
Air	$.2336 + .0000116\ (t^1+t)$

ent gases and using them for preheating the blast prior to combustion. By this means there is, in addition to the preheat of 1260 Btu. in each pound of coke as it arrives at the Combustion Zone, 1600 or 1800 Btu. in the preheated air of combustion for each lb. of coke.

There is thus a continuous circulation of approximately 3000 Btu. for each lb. of coke that is taken from the comparatively low temperature at the upper part of the preheating zone and carried to the combustion zone through the media of preheated coke and air, where it reappears at the highest temperature in the combustion zone.

The marked increase in brilliancy and intensity of combustion produces a saving in fuel entirely disproportionate to the amount of extra heat introduced by the warm blast.

This is because, at the higher temperature, the chemical reactions and absorption take place with greater rapidity and the additional heat carries with it no additional gases to carry away heat through the stack; there is no additional ash to be converted into slag; also the dilution of CO_2 in proportion to the heat developed permits a higher stage of combustion, which is another source of increasing the heat developed.

It is therefore quite evident that the regenerative feature results in saving in greater proportion than indicated by the additional heat developed.

It is not the purpose of this paper to recommend any type of cupola. Its sole objective is to establish the thermal principles involved in the development and transfer of heat which fix the foundation for the various efficiencies. Much can be done to improve the adaptation of these principles to daily routine.

Whether it will be advisable to preheat the blast in any particular case will depend upon an analysis of the conditions existing at that cupola and the possible daily savings in fuel, or improvement in quality of iron that may be secured, compared with the cost of remodelling the cupola for preheated practice.

This analysis is the function of the executives of the plant. Management has a far greater responsibility than is ordinarily assumed in

Table 24

COMPOSITION OF GASES AND HEAT DEVELOPED FROM THE COMBUSTION OF 1 LB.

1		2	3	4	5	6	7	8	9	10	11	12
% CO_2 Stage of Combustion		Lb. Carbon Burned	Cu. Ft. Gases Developed by Combustion A & B: CO_2 30A	CO 30B	Nitrogen 120A	Nitrogen 60B	Total Cu. Ft. 5 & 6	Total Cu. Ft. and Lb. of Gas	Percent by Weight and Percent by Volume: CO_2	CO	N	Extra Oxygen
0	A	0.00000000 } Cu. ft.	0.00000	30.00000	0.00000	60.00000	60.00000	90.00000	0.0000	33.3333	66.6667	.9995
	B	1.00000000 } Lb.	.....	2.34000	.....	4.70700	4.70700	7.04700	0.0000	33.2056	66.7944	.0000
1	A	0.03061224 } Cu. ft.	0.91836	29.08164	3.67344	58.16328	61.83672	91.83672	1.0000	31.6667	67.3333	.9680
	B	0.96938776 } Lb.	.11250	2.26837	.28818	4.56291	4.85109	7.23196	1.5557	31.3659	67.0784	.0837
2	A	0.06250000 } Cu. ft.	1.875000	28.12500	7.50000	56.25000	63.75000	93.75000	2.0000	30.0000	68.0000	.9933
	B	0.93750000 } Lb.	.22969	2.19375	.58838	4.41281	5.00119	7.42465	3.0945	29.5468	67.3592	.0686
3	A	0.09574468 } Cu. ft.	2.87234	27.12766	11.48936	54.25532	65.74468	95.74468	3.0000	28.3333	68.6667	1.0250
	B	0.90425532 } Lb.	.35186	2.11596	.90134	4.25632	5.15767	7.62549	4.6143	27.7485	67.6372	.0915
4	A	0.13043478 } Cu. ft.	3.91304	26.08696	15.65217	52.17391	67.82608	97.82608	4.0000	26.6667	69.3333	1.0564
	B	0.86956522 } Lb.	.47934	2.03478	1.22791	4.09304	5.32096	7.83508	6.1179	25.9701	67.9120	.0945
5	A	0.16666667 } Cu. ft.	5.00000	25.00000	20.00000	50.00000	70.00000	100.00000	5.0000	25.0000	70.0000	1.0942
	B	0.83333333 } Lb.	.61250	1.95000	1.56900	3.92250	5.49150	8.05400	7.6050	24.2115	68.1835	.0976
6	A	0.20454545 } Cu. ft.	6.13636	23.86364	24.54545	47.72727	72.27272	102.27272	6.0000	23.3333	70.6667	1.1325
	B	0.79545455 } Lb.	.75170	1.86136	1.92559	3.74420	5.66980	8.28286	9.0754	22.4724	68.4522	.1010
7	A	0.24418604 } Cu. ft.	7.32558	22.67442	29.30232	45.34884	74.65116	104.65116	7.0000	21.6747	71.3333	1.1715
	B	0.75581396 } Lb.	.89738	1.76860	2.29877	3.55761	5.85638	8.52236	10.5298	20.7524	68.7178	.1045
8	A	0.28571428 } Cu. ft.	8.57143	21.42857	34.28571	42.85714	77.14285	107.14285	8.0000	20.0000	72.0000	1.2130
	B	0.71428572 } Lb.	1.05000	1.67143	2.68971	3.36214	6.05186	8.77329	11.9682	19.0513	68.9805	.1082
9	A	0.32926829 } Cu. ft.	9.87805	20.121951	39.51219	40.24390	79.75609	109.75609	9.0000	18.3333	72.6667	1.2556
	B	0.67073170 } Lb.	1.21006	1.56951	3.09973	3.15713	6.25687	9.03644	13.3910	17.3686	69.2404	.1120
10	A	0.37500000 } Cu. ft.	11.25000	18.750000	45.00000	37.50000	82.50000	112.50000	10.0000	16.6667	73.3333	1.3010
	B	0.62500000 } Lb.	1.37812	1.46250	3.53025	2.94187	6.47212	9.31274	14.7983	15.7042	69.4975	.1161
11	A	0.42307692 } Cu. ft.	12.69231	17.30769	50.76923	34.61538	85.38461	115.38461	11.0000	15.0000	74.0000	1.3472
	B	0.57692308 } Lb.	1.55480	1.35000	3.98285	2.71558	6.69842	9.60322	16.1905	14.0577	69.7518	.1202
12	A	0.47368421 } Cu. ft.	14.21053	15.78947	56.84210	31.57894	88.42105	118.42105	12.0000	13.3333	74.6667	1.3991
	B	0.52631579 } Lb.	1.74079	1.23157	4.45926	2.47737	6.93663	9.90899	17.5678	12.4288	70.0034	.1246
13	A	0.52702702 } Cu. ft.	15.81081	14.18919	63.24324	28.37838	91.62162	121.62162	13.0000	11.6667	75.3333	1.4507
	B	0.47297298 } Lb.	1.93682	1.10676	4.96143	2.22628	7.18772	10.23130	18.9303	10.8174	70.2523	.1294
14	A	0.58333333 } Cu. ft.	17.50000	12.50000	70.00000	25.00000	95.00000	125.00000	14.0000	10.0000	76.0000	1.5063
	B	0.41666667 } Lb.	2.14375	.97500	5.49150	1.96125	7.45275	10.57150	20.2786	9.2229	70.4985	.1344
15	A	0.64285714 } Cu. ft.	19.28571	10.71429	77.14286	21.42857	98.57143	128.57143	15.0000	8.3333	76.6667	1.5630
	B	0.35714286 } Lb.	2.36250	.83571	6.05186	1.68107	7.73292	10.93113	21.6127	7.6452	70.7421	.1396
16	A	0.70588235 } Cu. ft.	21.17647	8.82353	84.70588	17.64706	102.35294	132.35294	16.0000	6.6667	77.3333	1.6267
	B	0.29411765 } Lb.	2.59412	.68823	6.64518	1.38441	8.02959	11.31194	22.9326	6.0841	70.9833	.1451
17	A	0.77272727 } Cu. ft.	23.18181	6.81818	92.72727	13.63636	106.36363	136.36363	17.0000	5.0000	78.0000	1.6920
	B	0.22727273 } Lb.	2.83977	.53182	7.27445	1.06977	8.34422	11.71581	24.2388	4.5393	71.2219	.1510
18	A	0.84375000 } Cu. ft.	25.31250	4.68750	101.25000	9.37500	110.62500	140.62500	18.0000	3.3333	78.6667	1.7630
	B	0.15625000 } Lb.	3.10078	.36563	7.94306	.73547	8.67853	12.14494	25.5315	3.0105	71.4580	.1572
19	A	0.91935483 } Cu. ft.	27.58064	2.41936	110.32258	4.83871	115.16129	145.16129	19.0000	1.6667	79.3333	1.8374
	B	0.08064517 } Lb.	3.37863	.18871	8.65481	.37960	9.03440	12.60174	26.8110	1.4974	71.6916	.1639
20	A	0.10000000 } Cu. ft.	30.00000	0.00000	120.00000	.00000	120.00000	150.00000	20.0000	0.0000	80.0000	1.8789
	B	0.00000000 } Lb.	3.67840	.00000	9.41400	.00000	9.41400	13.08900	28.0957	0.0000	71.9043	.1676

A—Carbon burned to CO_2.
B—Burned to CO.

effecting melting economy. It is the function of the management to review the various costs of labor and material used per ton of iron melted. The objective, of course, is to furnish a ton of molten iron of the desired quality to the foundry at the lowest cost. There are a large number of variables to be considered.

It is the responsibility of the management to furnish the facilities that will permit the best possible operating conditions, with minimum material and labor costs, for producing the highest possible quality of product. The industry would be greatly benefitted by complete analysis of all materials which are used by the foundry such as coke, brick, fireclay and other patching materials. At present experience is the only guide to the choice of these materials.

A great responsibility rests upon the cupola operator. It is his function to get the best out of the cupola under fixed routine conditions established by the management.

It is his responsibility to operate the cupola smoothly and uniformly, without trouble at the tapping hole or slag hole, to prevent scaffolding with its tendencies toward hang-ups, and to guard against irregularities of many kinds such as burning through the shell, run-outs at the bottom, freeze-ups, and all other ills which at times baffle the operator.

The skillful operator avoids these troubles, and at the same time produces the hottest iron, free from oxidation, with the least amount of fuel.

Much depends upon the make-up of the cupola; the chipping out, patching, preparation of sand bottom, placing of bed coke, lighting and burning the bed, placing the charges; limestone and purite ratios; size and reactance of coke, blast pressure which regulates volume of air per minute, constant review of information furnished by the volume meter, CO_2 record, humidity record, moisture content in coke, making the necessary adjustments to insure uniformity from tap to tap.

All of these items are of importance and worthy of detailed analysis by competent technicians.

The metallurgical staff has the important responsibility of preparing specifications for all materials which enter the cupola or are used in its construction and maintenance. Inasmuch as the air used for combustion is practically of the same weight as the iron charged and is the most conspicuous item in cupola control, it is important that the quality of the air should be contaminated as little as possible.

The main adulterant is aqueous vapor which is indicated on the humidity recording chart. An excess of humidity is definitely objectionable on account of its tendency toward hardening the iron by increasing the silicon loss and reducing the total carbon, in addition to the definitely undesirable effect of hydrogen as an occluded gas in the iron.

Studies should be made to determine ways and means of eliminating, or correcting, the effect of excess moisture carried to the Combustion Zone by the blast.

There is a special opportunity for improving the specifications for coke, brick for cupola lining and various other types of patching materials, with special reference to economy in their use.

The Cupola Handbook

Much has been done along these lines and appears in THE CUPOLA HANDBOOK published by the AMERICAN FOUNDRYMEN'S ASSOCIATION. This book constitutes a very valuable contribution as a reference encyclopedia for information relating to all these problems. There is always an opportunity for further research into all details of cupola efficiency. The writer hopes that these studies on fuel economy furnish an incentive for preparation of more complete reference tables

OF CARBON IN A CUPOLA

13	14	15	16	17	18	19	20	21	22	23
	Cu. Ft. and Lb. Air for Combustion			British Thermal Units						% CO_2 Stage of Combustion
				Developed			Not Developed	Spec. Heat	Temp. Gases	
12.26A	6.13B	Total	14550A	4350B	Total	1-Lb.				
0.0000	75.9509	75.9509								
0.0000	6.1300	6.1300	0000	4350	4350	617.3	10200	.2679	2300	0
4.6499	73.6253	78.2753	445							
.3753	5.9423	6.3176		4217	4662	644.6	9888	.2700	2390	1
9.4944	71.2043	80.6987	909							
.7663	5.7469	6.5132		4078	4987	671.7	9563	.2720	2470	2
14.5397	68.6791	83.2226	1393							
1.1735	5.5431	6.7169		3934	5327	698.6	9223	.2741	2550	3
19.8129	66.0438	85.8567	1898							
1.5991	5.3304	6.9295		3783	5681	725.1	8869	.2763	2625	4
25.3165	63.2548	88.6085	2425							
2.0433	5.1053	7.1516		3625	6050	751.2	8500	.2786	2695	5
31.0704	60.4150	91.4855	2976							
2.5077	4.8761	7.3838		3460	6436	777.0	8114	.2810	2765	6
37.0920	57.4042	94.4962	3553							
2.9937	4.6331	7.6268		3288	6841	802.7	7709	.2834	2830	7
43.4010	54.2497	97.6507	4157							
3.5029	4.3785	7.8814		3107	7264	828.0	7286	.2859	2895	8
50.0160	50.9428	100.9589	4791							
4.0368	4.1116	8.1484		2918	7709	853.1	6841	.2884	2960	9
56.9631	47.4699	104.4330	5456							
4.5975	3.8313	8.4288		2719	8175	877.8	6375	.2909	3020	10
64.2658	43.8173	108.0831	6156							
5.1869	3.5365	8.7234		2510	8666	902.4	5884	.2934	3075	11
71.9538	39.9739	111.9278	6892							
5.8074	3.2263	9.0337		2289	9181	926.5	5369	.2959	3130	12
80.0569	35.9224	115.9793	7668							
6.4614	2.8993	9.3607		2057	9725	950.5	4825	.2984	3185	13
88.6097	31.6466	120.2564	8487							
7.1517	2.5542	9.7059		1813	10300	974.3	4250	.3012	3235	14
97.6507	27.1254	124.7762	9354							
7.8814	2.1893	10.0707		1554	10908	997.9	3642	.3036	3285	15
107.2245	22.3379	129.5625	10271							
8.6541	1.8029	10.4570		1279	11550	1021.0	3000	.3062	3335	16
117.3781	17.2617	134.6399	11243							
9.4736	1.3932	10.8668		989	12232	1044.1	2318	.3088	3380	17
128.1674	11.8671	140.0346	12277							
10.3444	.9578	11.3022		680	12957	1066.9	1593	.3114	3425	18
139.6505	6.1256	145.7761	13377							
11.2712	.4944	11.7656		351	13728	1089.4	822	.3140	3470	19
151.9017	.0000	151.9017	14550							
12.2600	.0000	12.2600		000	14550	1111.6	000	.3164	3515	20

Table 25

CUBIC FEET AND POUNDS OF AIR REQUIRED FOR COMBUSTION OF 1 LB. OF COKE OF VARIOUS CARBON CONTENTS, AND AT VARIOUS STAGES OF COMBUSTION; ALSO RESULTANT HEAT DEVELOPED EXPRESSED IN BTU.

Per Cent CO_2	Percentage Carbon — 100 % Air			95 % Air			90 % Air			85 % Air		
	Cu. Ft.	Lb.	Btu.	Cu. Ft.	Lb.	Btu.	Cu. Ft.	Lb.	Btu.	Cu. Ft.	Lb.	Btu.
0.0	75.95	6.13	4350	72.15	5.82	4132	68.35	5.52	3915	64.56	5.21	3697
0.5	77.11	6.22	4505	73.25	5.91	4280	69.40	5.60	4054	65.54	5.29	3829
1.0	78.27	6.31	4662	74.36	5.99	4429	70.44	5.68	4196	66.53	5.36	3963
1.5	79.48	6.41	4824	75.51	6.09	4583	71.53	5.77	4342	67.55	5.45	4100
2.0	80.69	6.51	4987	76.66	6.18	4738	72.62	5.86	4488	68.59	5.53	4239
2.5	81.95	6.61	5156	77.85	6.28	4898	73.75	5.95	4640	69.66	5.62	4383
3.0	83.22	6.71	5327	79.06	6.37	5061	74.90	6.04	4794	70.74	5.70	4528
3.5	84.50	6.81	5502	80.27	6.47	5227	76.05	6.13	4952	71.82	5.79	4667
4.0	85.85	6.92	5681	81.56	6.57	5397	77.26	6.23	5113	72.97	5.88	4829
4.5	87.22	7.03	5864	82.86	6.68	5571	78.50	6.33	5278	74.14	5.98	4984
5.0	88.60	7.15	6050	84.17	6.79	5747	79.74	6.43	5445	75.31	6.08	5142
5.5	90.04	7.26	6241	85.54	6.90	5929	81.04	6.53	5617	76.53	6.17	5305
6.0	91.48	7.38	6436	86.91	7.01	6114	82.33	6.64	5792	77.76	6.27	5471
6.5	92.98	7.50	6636	88.33	7.12	6304	83.68	6.75	5972	79.03	6.37	5641
7.0	94.49	7.62	6841	89.77	7.24	6499	85.04	6.86	6157	80.32	6.48	5819
7.5	96.07	7.75	7050	91.27	7.36	6697	86.46	6.97	6345	81.66	6.59	5992
8.0	97.65	7.88	7264	92.77	7.49	6901	87.88	7.09	6538	83.00	6.70	6174
8.5	99.30	8.01	7484	94.33	7.61	7110	89.37	7.21	6736	84.40	6.81	6361
9.0	100.95	8.14	7709	95.90	7.73	7324	90.85	7.33	6938	85.81	6.92	6553
9.5	102.69	8.28	7939	97.56	7.87	7542	92.42	7.45	7145	87.29	7.04	6748
10.0	104.43	8.42	8175	99.20	8.00	7766	93.99	7.58	7357	88.77	7.16	6949
10.5	106.25	8.57	8417	100.94	8.14	7996	95.62	7.71	7575	90.31	7.28	7154
11.0	108.08	8.72	8666	102.68	8.28	8233	97.27	7.85	7799	91.87	7.41	7366
11.5	110.00	8.87	8920	104.50	8.43	8474	99.00	7.98	8028	93.50	7.54	7582
12.0	111.92	9.03	9181	106.32	8.58	8722	100.72	8.13	8263	95.13	7.68	7804
12.5	113.94	9.19	9450	108.24	8.73	8978	102.55	8.27	8505	96.85	7.81	8033
13.0	115.97	9.36	9725	110.17	8.89	9239	104.37	8.42	8752	98.57	7.96	8266
13.5	118.11	9.53	10009	112.20	9.05	9509	106.30	8.58	9008	100.39	8.10	8508
14.0	120.25	9.70	10300	114.24	9.21	9785	108.22	8.73	9270	102.21	8.24	8755
14.5	122.51	9.88	10599	116.38	9.39	10069	110.26	8.89	9539	104.13	8.40	9009
15.0	124.77	10.07	10908	118.53	9.57	10363	112.29	9.06	9817	106.05	8.56	9272
15.5	127.16	10.26	11224	120.80	9.75	10663	114.44	9.23	10101	108.09	8.72	9540
16.0	129.56	10.45	11550	123.08	9.93	10972	116.60	9.40	10395	110.13	8.88	9817
16.5	132.09	10.65	11886	125.49	10.12	11291	118.88	9.58	10697	112.28	9.05	10103
17.0	134.63	10.86	12232	127.90	10.32	11620	121.17	9.77	11009	114.44	9.23	10397
17.5	137.33	11.08	12588	130.46	10.53	11959	123.60	9.97	11329	116.73	9.42	10700
18.0	140.03	11.30	12957	133.03	10.73	12309	126.03	10.17	11661	119.03	9.60	11013
18.5	142.90	11.53	13336	135.75	10.95	12669	128.61	10.38	12002	121.46	9.80	11336
19.0	145.77	11.76	13728	138.48	11.17	13042	131.19	10.58	12355	123.90	10.00	11669
19.5	148.83	12.01	14132	141.39	11.41	13425	133.95	10.81	12719	126.51	10.21	12012
20.0	151.90	12.26	14550	144.30	11.65	13822	136.71	11.03	13095	129.11	10.42	12367

for cupola operators. These tables represent only a preliminary study, hence, should be carefully revised.

APPENDIX

This Appendix contains Tables relating to the combustion of coke in cupolas which have been compiled by the author at various times as the necessity for such tabulations arose. When put to practical use, large economies in coke requirements have resulted.

It seems proper to attach an Appendix to the foregoing paper for the purpose of recording a few of the more important tables, in convenient form for reference, when calculating problems relating to the conservation of heat in cupolas.

The tables have been prepared from time to time by the author as the necessity for their use arose. They are intended only for the solution of ordinary problems relating to cupola economics and are not intended for the physical chemist in his studies of abstract problems.

It also seems proper that there should be an explanation of the origin of the constants on which the various tables are based. This involves the sources of heat, the mechanics of combustion, and the fundamental laws which underlie the specific heat of the elements entering into, and the gases resulting from, combustion.

Sources of Heat. The principal heat producing reaction is the combustion of carbon contained in the coke. In addition to this, there is a small amount of heat produced by the oxidation of miscellaneous elements such as hydrogen, sulphur, silicon, manganese, iron, etc., which augment, to a small degree, the total heat produced. The amount of heat that is developed in a chemical reaction is often expressed in calories, or Btu. per mole volume.

The mole volume in the metric system is the volume in cc., or liters, occupied by the number of grams of gas indicated by the molecular weight. The gram mole = 22.414 liters.

The mole volume, according to the British system of weights and measures, is the volume in cu. ft.

occupied by the number of lb. of gas indicated by the molecular weight. The lb. mole = 359 cu. ft.

In the preparation of heat balances the heat developed per mole is broken down to the heat developed per lb. of the material under consideration, and since the mole volume is based upon the weight in grams, or pounds, represented by the molecular weight, it is convenient to have available an abbreviated table of Atomic Weights, Table 37 on page 538.

In Table 37 the heat of formation is recorded for various combinations that occur in cupola practice, also the heat losses in the decomposition of limestone ($CaCO_3$) and the reduction of CO_2 to CO ($CO_2 + C = 2CO$); also the temperature of ignition for various forms of carbon and hydrocarbons is shown.

It has been customary by many authors to consider 14,550 Btu. as the heat developed from 1 lb. carbon obtained from coke. In certain recent papers on cupola practice this amount has been reduced to as low as 14,093 Btu.

These variations are explained by the studies made by the Carnegie Institute of Technology Coal Research Laboratory in collaboration with the National Bureau of Standards. (Contributions 71 of the Coal Research Laboratory.)

The conclusions of these tests were that the heat of formation of CO_2 from graphite varied from 14,085 Btu. to 14,110 Btu., which is in close agreement with the more reliable results obtained from the combustion of diamonds.

The results obtained from coke carbon were consistently higher largely on account of the small amount of hydrogen present. The amount of hydrogen varied according to the temperature of the preparation of the coke; the lower the temperature the higher the hydrogen content.

The variations from low to high were 0.80 to 2.78 per cent, the resultant Btu. per lb. of carbon being 14,535 to 14,866 Btu. Inasmuch as cokes are prepared at the higher temperature it would seem that the usual 14,550 Btu. rests upon a sound basis and is used as the potential

Table 26

CALORIFIC VALUE 1 LB. COKE OF VARIOUS CARBON CONTENTS FOR VARIOUS CHEMICAL COMBINATIONS

Carbon, per cent	$C + O_2 = CO_2$	CO	$2CO + O_2 = 2CO_2$	$CO_2 + C = 2CO$	$CO_2 + C = 2CO$ *Loss*
100	14550.0	4350.0	10200.0	8700.0	5850.0
99	14404.5	4306.5	10098.0	8613.0	5791.5
98	14259.0	4263.0	9996.0	8526.0	5733.0
97	14113.5	4219.5	9894.0	8439.0	5674.5
96	13968.0	4176.0	9792.0	8352.0	5616.0
95	13822.5	4132.5	9690.0	8265.0	5557.5
94	13677.0	4089.0	9588.0	8178.0	5499.0
93	13531.5	4045.5	9486.0	8091.0	5440.5
92	13386.0	4002.0	9384.0	8004.0	5382.0
91	13240.5	3958.2	9282.0	7917.0	5323.5
90	13095.0	3915.0	9180.0	7830.0	5265.0
89	12949.5	3871.5	9078.0	7743.0	5206.5
88	12804.0	3828.0	8976.0	7656.0	5148.0
87	12658.5	3784.5	8874.0	7569.0	5089.5
86	12513.0	3741.0	8772.0	7482.0	5031.0
85	12367.5	3697.5	8670.0	7395.0	4972.5
84	12222.0	3654.0	8568.0	7308.0	4914.0
83	12076.5	3610.5	8466.0	7221.0	4855.5
82	11931.0	3567.0	8364.0	7134.0	4797.0
81	11785.5	3523.5	8262.0	7047.0	4738.5
80	11640.0	3480.0	8160.0	6960.0	4680.0

Table 27

COMBUSTION OF 1 LB. OF 100 PER CENT CARBON

	Portion of 1 Lb. Carbon				*Btu.*		
Stage of Combustion ($\%CO_2$)	*In Effluent Gases*		*Burned to CO_2 in Combustion Zone*	*For Reducing CO_2 to CO*	*Developed in Combustion Zone*	*Lost in Reducing CO_2 to CO*	*Net Developed*
	in CO_2 A	*in CO* B	$A + \frac{1}{2}B$	$\frac{1}{2}B$	$A + \frac{1}{2}B$	$\frac{1}{2}B$	
6.0	.204545	.795455	.602273	.397727	8763.1	2326.7	6436.4
6.5	.224138	.775862	.612069	.387931	8905.6	2269.4	6636.2
7.0	.244186	.755814	.622093	.377907	9051.5	2210.8	6840.7
7.5	.264706	.735294	.632353	.367647	9200.7	2150.7	7050.0
8.0	.285714	.714286	.642857	.357143	9353.6	2089.3	7264.3
8.5	.307229	.692771	.653614	.346386	9510.1	2026.4	7483.7
9.0	.329268	.670732	.664634	.335366	9670.4	1961.9	7708.5
9.5	.351852	.648148	.675926	.324074	9834.7	1895.8	7938.9
10.0	.375000	.625000	.687500	.312500	10003.1	1828.1	8175.0
10.5	.398734	.601266	.699367	.300633	10175.8	1758.7	8417.1
11.0	.423077	.576923	.711539	.288461	10352.9	1687.5	8665.4
11.5	.448052	.551948	.724026	.275974	10534.6	1614.4	8920.2
12.0	.473684	.526316	.736842	.263158	10721.1	1539.5	9181.6
12.5	.500000	.500000	.750000	.250000	10912.5	1462.5	9450.0
13.0	.527027	.472973	.763513	.236487	11109.1	1383.4	9725.7
13.5	.554795	.445205	.777397	.222603	11311.1	1302.2	10008.9
14.0	.583333	.416667	.791667	.208333	11518.8	1218.7	10300.1
14.5	.612676	.387324	.806338	.193662	11732.2	1132.9	10559.3
15.0	.642857	.357143	.821429	.178571	11951.8	1044.6	10907.2
15.5	.673913	.326087	.836957	.163043	12177.7	953.8	11223.9
16.0	.705882	.294118	.852941	.147059	12410.3	860.3	11550.0

Table 28

COMBUSTION OF 1 LB. OF COKE (90 PER CENT CARBON)

Calorific Value: 13,095 Btu. Loss CO_2 Reduction: 5,265 Btu.

Stage of Combustion ($\%CO_2$)	Portion of 1 Lb. Coke — In Effluent Gases — in CO_2 A	in CO B	Burned to CO_2 $A + \frac{1}{2}B$ Combustion Efficiency	For Reducing CO_2 to CO $\frac{1}{2}B$	Btu. — Developed in Combustion Zone $A + \frac{1}{2}B$	Lost in Reducing CO_2 to CO $\frac{1}{2}B$	Net Developed
6.0	.204545	.795455	.602273	.397727	7886.79	2094.03	5792.76
6.5	.224138	.775862	.612069	.387931	8015.00	2042.50	5972.50
7.0	.244186	.755814	.622093	.377907	8146.35	1989.72	6156.63
7.5	.264706	.735294	.632353	.367647	8280.60	1935.60	6345.00
8.0	.285714	.714286	.642857	.357143	8418.24	1880.37	6537.87
8.5	.307229	.692771	.653614	.346386	8559.10	1823.90	6735.20
9.0	.329268	.670732	.664634	.335366	8703.36	1765.71	6937.65
9.5	.351852	.648148	.675926	.324074	8851.20	1706.20	7145.00
10.0	.375000	.625000	.687500	.312500	9002.79	1645.29	7357.50
10.5	.398734	.601266	.699367	.300633	9158.20	1582.80	7575.40
11.0	.423077	.576923	.711539	.288461	9317.61	1518.75	7798.86
11.5	.448052	.551948	.724026	.275974	9481.10	1453.00	8028.10
12.0	.473684	.526316	.736842	.263158	9648.99	1385.55	8263.44
12.5	.500000	.500000	.750000	.250000	9821.20	1316.20	8505.00
13.0	.527027	.472973	.763513	.236487	9998.19	1245.06	8753.13
13.5	.554795	.445205	.777397	.222603	10180.00	1172.00	9008.00
14.0	.583333	.416667	.791667	.208333	10366.92	1096.83	9270.09
14.5	.612676	.387324	.806338	.193662	10559.00	1019.60	9539.40
15.0	.642857	.357143	.821429	.178571	10756.62	940.14	9816.48
15.5	.673913	.326087	.836957	.163043	10959.90	858.40	10101.50
16.0	.705882	.294118	.852941	.147059	11169.27	774.27	10395.00

heat which is developed by complete combustion of coke, resulting in CO_2 and nitrogen.

In the case of decomposition of CO_2 resulting in CO, the net Btu. developed per lb. of carbon is only 4,350.

The quantity of heat developed from the combustion of one lb. of carbon varies somewhat with the conditions of pressure and temperature under which the reaction is carried out.

In the cupola the carbon is burned under conditions of approximate constant pressure, variable volume and at a high temperature whereas the Btu. developed in this reaction are determined in the laboratory by means of a bomb calorimeter in which the volume of the gases is constant, the pressure variable, and the temperature of the materials entering into combustion at the laboratory temperature of, say, 60 to 70° F.

Effect of Temperature at Which Reaction Takes Place. There is a slight increase in the heat developed by combustion when the temperature at which the reaction takes place is high because it requires less heat to preheat the O_2, which is a diatomic molecule as compared with the CO_2 resulting from combustion,

Table 29

HEAT CAPACITY OF IRON: OBSERVED DATA

	Temperature ° C.	° F.	Calories	Btu.
Solid	660	1220	96.1	173
Change in State	730	1346	5.3	10
Change in State	900	1620	6.0	11
Solid	1535	2795	256.0	461
Latent Heat of Fusion	1535	2795	66.0	120
Total Liquid	1535	2795	322.0	581
Liquid	2450	4440	533.0	959
Latent Heat of Vaporization	2450	4440	1224.0	2203
Total Vapor	2450	4440	1757.0	3162

(Richards' Metallurgical Tables)

Table 30

BTU. IN 1 LB. OF IRON

Temp., ° F.	Btu. in Solid	Specific Heat of Iron	Carbon Content (Per cent) 0	.50	1.0	1.5	2.0	2.5	3.0	3.5	4.0
			Latent Heat of Fusion 120	115	107	100	93	86	79	72	65
2100	345										$\overline{410}$
2190	360	.167	Lines indicate melting point: Solid above line,							$\overline{432}$	431
2280	375	.161	Liquid below line.						$\overline{454}$	453	451
2370	389	.161						$\overline{475}$	475	473	472
2460	404	.172					$\overline{497}$	496	495	494	493
2550	420	.172				$\overline{520}$	518	516	516	515	513
2640	435	.167			$\overline{542}$	541	538	537	537	535	534
2730	450	.168		$\overline{565}$	563	561	559	558	557	556	555
2800	462	.168	$\overline{582}$	581	579	577	575	574	573	572	571
2900	Aver.	.167	605	604	602	600	598	597	596	595	594

Table 31

HEAT CAPACITY OF 1 LB. IRON AT VARIOUS TEMPERATURES IN BTU.

Deg. F.	0	1	2	3	4	5	6	7	8	9
2550	515	515	515	515	516	516	516	516	517	517
2560	517	517	517	518	518	518	518	519	519	519
2570	519	520	520	520	520	520	521	521	521	521
2580	522	522	522	522	523	523	523	523	523	524
2590	524	524	524	525	525	525	525	526	526	526
2600	526	526	527	527	527	527	528	528	528	528
2610	528	529	529	529	529	530	530	530	530	531
2620	531	531	531	531	532	532	532	532	533	533
2630	533	533	534	534	534	534	534	535	535	535
2640	535	536	536	536	536	537	537	537	537	537
2650	538	538	538	538	539	539	539	539	540	540
2660	540	540	540	541	541	541	541	542	542	542
2670	542	543	543	543	543	543	544	544	544	544
2680	545	545	545	545	546	546	546	546	546	547
2690	547	547	547	548	548	548	548	549	549	549
2700	549	549	550	550	550	550	551	551	551	551
2710	552	552	552	552	552	553	553	553	553	554
2720	554	554	554	554	555	555	555	555	556	556
2730	556	556	557	557	557	557	557	558	558	558
2740	558	559	559	559	559	560	560	560	560	560
2750	561	561	561	561	562	562	562	562	563	563
2760	563	563	563	564	564	564	564	565	565	565
2770	565	566	566	566	566	566	567	567	567	567
2780	568	568	568	568	569	569	569	569	569	570
2790	570	570	570	571	571	571	571	572	572	572
2800	572	572	573	573	573	573	574	574	574	574
2810	575	575	575	575	575	576	576	576	576	577
2820	577	577	577	577	578	578	578	578	579	579
2830	579	579	580	580	580	580	580	581	581	581
2840	581	582	582	582	582	583	583	583	583	583
2850	584	584	584	584	585	585	585	585	586	586
2860	586	586	586	587	587	587	587	588	588	588
2870	588	589	589	589	589	589	590	590	590	590
2880	591	591	591	591	592	592	592	592	592	593
2890	593	593	593	594	594	594	594	595	595	595
2900	595	595	596	596	596	596	597	597	597	597

Table 32

MEAN SPECIFIC HEAT OF CUPOLA GASES BETWEEN ZERO AND 3,000° F.

Stage of Combustion — CO_2 per cent — By Vol.	By Wt.	N and CO By Wt., per cent	Temp. of Gases, ° F. 0	500	1000	1500	2000	2500	3000	Difference 100°
0	0	100.00	.2405	.246	.252	.258	.263	.270	.276	.00119
1	1.56	98.44	.2397	.246	.252	.258	.265	.271	.278	.00127
2	3.09	96.91	.2389	.246	.252	.259	.266	.272	.279	.00134
3	4.61	95.39	.2382	.245	.252	.259	.266	.274	.281	.00142
4	6.12	93.88	.2374	.245	.252	.260	.267	.275	.282	.00149
5	7.60	92.40	.2367	.244	.252	.260	.268	.276	.284	.00156
6	9.08	90.92	.2359	.244	.252	.260	.269	.277	.285	.00164
7	10.53	89.47	.2352	.244	.252	.261	.269	.278	.286	.00171
8	11.97	88.03	.2345	.243	.252	.261	.270	.279	.288	.00178
9	13.39	86.61	.2337	.243	.252	.261	.271	.280	.289	.00185
10	14.80	85.20	.2330	.243	.252	.262	.271	.281	.290	.00192
11	16.19	83.81	.2323	.242	.252	.262	.272	.282	.292	.00199
12	17.57	82.43	.2316	.242	.252	.262	.273	.283	.293	.00205
13	18.93	81.07	.2309	.241	.252	.263	.273	.284	.295	.00212
14	20.28	79.72	.2303	.241	.252	.263	.274	.285	.296	.00219
15	21.61	78.39	.2296	.241	.252	.263	.275	.286	.297	.00225
16	22.93	77.07	.2289	.240	.252	.264	.275	.287	.298	.00232
17	24.24	75.76	.2283	.240	.252	.264	.276	.288	.300	.00238
18	25.53	74.47	.2276	.240	.252	.264	.276	.289	.301	.00245
19	26.81	73.19	.2270	.239	.252	.265	.277	.290	.302	.00251
20	28.00	72.00	.2264	.239	.252	.265	.278	.291	.303	.00257
Air — Oxygen 23.2		Nitrogen 76.80	.2336	.245	.257	.268	.280	.292	.303	.00232
Water Vapor — Hydrogen 11.1		Oxygen 88.90	.4200	.471	.523	.574	.626	.677	.729	.01030

which is a triatomic molecule. Consequently this requires approximately 50 per cent more heat to increase its temperature over the given range as compared with O_2.

At a temperature of 1000° F., the increase is only 33 Btu., or a total for one lb. of carbon of 14,550 + 33 = 14,583 Btu. Consequently this variation is negligible as far as cupola problems are concerned.

Effect of Pressure

The conditions of temperatures and pressure on the amount of heat developed per lb. of fuel have been analyzed by Haslam and Russell in *Fuels and Their Combustion,* pp. 138 to 140, in which they developed the equation for constant pressure vs. constant volume as follows:

$$Qp = Qv - (n)\,RT$$

in which Qp equals the heat of reaction at constant pressure, Qv equals the heat of reaction at constant volume and n equals the increase in the number of molecules in the gas present after the reaction occurs.

R, the gas constant, in terms of Btu. per degree of temperature variation combined with pressure variation, equals 1.9885, or approximately 2.0 per mole, or 0.166 per lb. carbon; T equals the temperature at which the reaction is carried out in degrees Fahrenheit absolute (° F. + 460).

For the reaction $2CO + O = 2CO_2$ + 20,400 Btu., n equals -1. and the heat of combustion in a closed bomb at constant volume is Qv = 20,400 + (-1) (0.166) (60 + 460) = 20,400 -87 = 20,313 Btu. approx.

The heat evolved by this reaction at constant volume is 87 Btu. less than if the same amount of CO were burned under conditions of constant pressure. This correction amounts to but one-half of 1 per cent.

Laboratory Determinations of Calorific Values

In reporting the Btu. in a lb. of combustible for practical application to furnace problems, the laboratory results are reduced to the condition of uniform pressure and variable volumes. On this basis the quantity of heat evolved from 1 lb. of carbon is the following:

Table 33

BTU. IN PREHEATED BLAST FOR COMBUSTION OF 1 LB. CARBON

Temp. °F.....		200	400	600	700	800	900	1000	1100	1200	1300	1400	1500
Spec. Heat		.2382	.2429	.2475	.2498	.2522	.2545	.2568	.2591	.2614	.2638	.2661	.2684
Btu., 1 lb. Air		47.64	97.16	148.50	174.86	201.76	229.05	256.80	285.01	313.68	342.94	372.54	402.60
% CO_2 Stage of Combustion	*Lb. Air*												
0	6.13	292	596	910	1072	1237	1404	1574	1747	1923	2102	2284	2468
1	6.32	301	614	939	1105	1275	1448	1623	1801	1982	2167	2354	2544
2	6.51	310	633	967	1138	1313	1491	1672	1855	2042	2233	2425	2621
3	6.72	320	653	998	1175	1356	1539	1726	1915	2108	2305	2503	2705
4	6.93	330	673	1029	1212	1398	1587	1780	1975	2174	2377	2582	2790
5	7.15	341	695	1062	1250	1443	1638	1836	2038	2243	2452	2664	2879
6	7.38	352	717	1096	1290	1489	1690	1895	2103	2315	2531	2749	2971
7	7.63	363	741	1133	1334	1539	1748	1959	2175	2394	2617	2842	3072
8	7.88	375	766	1170	1378	1590	1805	2024	2246	2472	2702	2936	3172
9	8.15	388	792	1210	1425	1644	1867	2093	2323	2556	2795	3036	3281
10	8.43	402	819	1252	1474	1701	1931	2165	2403	2664	2891	3141	3394
11	8.72	415	847	1295	1525	1759	1997	2239	2485	2735	2990	3249	3511
12	9.03	430	877	1341	1579	1822	2068	2319	2574	2833	3097	3364	3635
13	9.36	446	909	1390	1637	1888	2144	2404	2668	2936	3210	3487	3768
14	9.71	463	943	1442	1698	1959	2224	2494	2767	3046	3330	3617	3909
15	10.07	480	978	1495	1761	2032	2307	2586	2870	3159	3453	3751	4054
16	10.46	498	1016	1553	1829	2110	2396	2686	2981	3281	3587	3897	4211
17	10.87	518	1056	1614	1901	2193	2490	2791	3098	3410	3728	4050	4376
18	11.30	538	1098	1678	1976	2280	2588	2902	3221	3534	3875	4210	4549
19	11.77	561	1144	1748	2058	2375	2696	3023	3355	3692	4036	4385	4739
20	12.26	584	1191	1821	2144	2474	2808	3148	3494	3846	4204	4567	4936

Reaction		*1 lb. of Combustible*	*Btu.*
$C + O_2$	=	CO_2	14,550
$C + \frac{1}{2} O_2$	=	CO	4,350
$2CO + O_2$	=	2CO (20,400)	
$C + CO_2$	=	2CO	−5,850*
$2H_2 + O_2$	=	$2H_2O$	61,550
$C + H_2O$	=	$CO + H_2$	−5,908*

*Loss.

CO_2 and CO Ratios (Stage of Combustion). From the established constants for potential heat energy in one lb. of carbon, or its equivalent in coke, it is clearly evident that the net heat developed is a variable depending upon the ratio of CO_2 to CO in the effluent gases.

It is convenient to consider these ratios in steps of 1 per cent in CO_2 and to call each step a stage of combustion. In each stage of combustion a definite part of each lb. of carbon, or coke, is used to reduce

Table 34

BTU. IN PREHEATED BLAST FOR COMBUSTION OF 1 LB. COKE (90 PER CENT)

Temp. °F.....		200	400	600	700	800	900	1000	1100	1200	1300	1400	1500
Spec. Heat		.2382	.2429	.2475	.2498	.2522	.2545	.2568	.2591	.2614	.2638	.2661	.2684
Btu., 1 lb. Air		47.64	97.16	148.50	174.86	201.76	229.05	256.80	285.01	313.68	342.94	372.54	402.60
% CO_2 Stage of Combustion	*Lb. Air*												
0	5.52	263	536	819	965	1113	1264	1417	1572	1731	1892	2056	2221
1	5.68	271	553	845	994	1147	1303	1461	1621	1784	1950	2119	2290
2	5.86	279	570	870	1024	1182	1342	1505	1669	1838	2010	2182	2359
3	6.04	288	588	898	1057	1220	1385	1553	1723	1897	2074	2253	2434
4	6.23	297	606	926	1091	1258	1428	1602	1777	1957	2139	2324	2511
5	6.43	307	625	956	1125	1299	1474	1652	1834	2019	2207	2398	2591
6	6.64	317	645	986	1161	1340	1521	1705	1893	2083	2278	2474	2674
7	6.86	327	667	1020	1201	1385	1573	1763	1957	2155	2355	2558	2765
8	7.09	337	689	1053	1240	1431	1624	1822	2021	2225	2432	2642	2855
9	7.33	349	713	1089	1282	1480	1680	1884	2091	2300	2515	2732	2953
10	7.58	362	737	1127	1327	1531	1738	1948	2163	2398	2602	2837	3055
11	7.85	373	762	1165	1372	1583	1797	2015	2236	2461	2691	2924	3160
12	8.13	387	789	1207	1421	1640	1861	2087	2317	2550	2787	3028	3271
13	8.42	401	818	1251	1473	1699	1930	2164	2401	2642	2889	3138	3391
14	8.73	417	849	1298	1528	1763	2002	2245	2490	2741	2997	3255	3518
15	9.06	432	880	1345	1585	1829	2076	2327	2583	2843	3108	3376	3649
16	9.40	448	914	1398	1646	1899	2156	2417	2683	2953	3228	3507	3790
17	9.77	466	950	1453	1711	1974	2241	2512	2788	3069	3355	3645	3938
18	10.17	484	988	1510	1778	2052	2329	2612	2899	3190	3487	3789	4094
19	10.58	505	1030	1573	1852	2137	2426	2721	3019	3323	3632	3946	4266
20	11.03	526	1072	1639	1930	2227	2527	2833	3145	3461	3784	4110	4442

Table 35

PHYSICAL CONSTANTS OF VARIOUS GASES INVOLVED IN COMBUSTION[1]

32° F. and Pressure One Atmosphere

	1	2	3	4	5	6
	—Specific Gravity—		*Weight*	*Cubic*	*Mean*	*Cubic*
	Air 1.00	*Oxygen 1.00*	*per Cu. Ft.*	*Feet per Lb.*	—*Specific Heat*— *0° to 3000° F.*	*Expansion*[2] *1° F.*
Oxygen	1.10544	1.00000	.089210	11.209506	.2104 + .0000104t	.00270
Pure Nitrogen	.96732	.87506	.078064	12.809951	.2405 + .0000119t	.002+
Air	1.00000	.90462	.080701	12.391432	.2336 + .0000116t	.002042
CO	.96716	.87491	.078051	12.812067	.2405 + .0000119t	.0020382
CO_2	1.52913	1.38328	.123402	8.103599	.1900 + .0000611t	.002061
Atmospheric Nitrogen	.97198	.87927	.078440	12.748597	.2405 + .0000119t	.002039
Water Vapor	.62200	.56270	.05020	19.92	.4200 + .000103t	.00248

RELATIONSHIP OF CONSTANTS

Col. 1—Specific gravity: Air = 1.00 = Weight of Gas ÷ Weight of Air
Col. 2— Oxygen = 1.00 = Weight of Gas ÷ Weight of Oxygen
Col. 3—Weight of one cu. ft. gas = Weight Cubic Foot Air × Specific Gravity of Gas, Col. 1
= Weight Cubic Foot Oxygen × Specific Gravity of Gas, Col. 2
= Molecular Weight ÷ Mole Volume
Col. 4—Cu. ft. per lb. = 1.00000 ÷ Weight of Gas per cu. ft.
= Mole Volume ÷ Molecular Weight
Col. 5—Richards' Metallurgical Calculations
Col. 6—Metallurgical Tables—Liddell

[1]Based on National Bureau of Standards (accepted standards).
[2]Absolute zero −273.18° C.; −459.72° F.; below zero Cent. −491.72° F.

Table 36

CONVERSION FACTORS

Length

1 m. = 3.2808333333 ft.
1 ft. = 0.3048006096 m.

Volume

1 cu. ft. = 28.31701660 cu. decimeters
1 cu. ft. = 28.31625206 liters
1 liter = 1.000027 decimeters

Weight

1 kg. = 2.204622341 lb.
1 lb. = 453.5924277 grams

$$\text{Cu. ft. mole} = \frac{453.5924277}{28.3170166} = 16.01837 \text{ CM}^3 \text{ gram moles}$$

$$\text{Cu. ft. mole} = \frac{453.5924277}{28.31625206} = 16.01880 \text{ liter gram moles}$$

Heat Units

1.0 Calorie = 3.968 Btu. = 2.2046 lb. calories

$$\text{1.0 Btu.} = \frac{3.968}{2.2046} = 1.7953 \text{ lb. calories}$$

1.0 Btu. = 0.252 calories = 777.97 ft. lb.
1.0 Lb. Cal. = 9/5 Btu. = .4536 calories

Mole Volumes[1]

	Molecular Weight	*Factor*	$\frac{CC}{CM^3}$ *Gram Mole*	*Liters*	*Cu. Ft. Lb. Mole*
Ideal Gas		100.00000	22414.6	22414.0	359.0454
O_2	32.000	0.99905	22393.3	22392.7	358.7042
N_2	28.016	0.99955	22404.5	22403.9	358.8836
Air	28.959	0.99944	22402.0	22401.4	358.8435
CO	28.010	0.99950	22403.4	22402.8	358.8660
CO_2	44.010	0.99330	22264.4	22263.8	356.6394
N_2 Atmospheric	28.150	0.999617	22403.8	22328.2	358.8728

[1]Based on data furnished by National Bureau of Standards.

a portion of the CO_2 to CO. This portion does not reach the combustion zone and hence cannot be considered in the calculation of heat produced by combustion.

When considering combustion, it is only that portion which reaches the combustion zone that develops CO_2 and producing heat at the rate of 14,550 Btu. for each lb. of carbon consumed, or assuming the coke to contain 90 per cent carbon, the heat devoloped from each lb. of coke that reaches the combustion zone is 13,095 Btu.

There are two steps in arriving at the final CO_2 and CO ratio; first, the complete combustion to CO_2 and, second, the reduction of a portion of the CO_2 to CO. In preparing reference tables of the various elements entering into each stage of combustion, the portion of the lb. of carbon contained in the final CO_2 is designated as A and the part in the CO is designated as B, and A + B equal 1 lb.

The reduction of CO_2 to CO is represented in the formula $CO_2 + C = 2CO$.

This represents the use of 2 lb. of carbon in the CO, one-half of which was taken from the CO_2 and one-half directly from the carbon of the coke. By this conception, the amount of carbon that reaches the

combustion zone and is burned to CO_2 is $A + \frac{B}{2}$, and the remaining $\frac{B}{2}$ represents the carbon used in the reduction of CO_2 and constitutes a loss at the rate of 5,850 Btu. for each lb. of carbon used in the reduction.

For example, in referring to Table 24 (p. 530) at the tenth stage of combustion, which represents 10 per cent of CO_2 in the final division of one lb. of carbon between CO_2 and CO, it is found that A equals 0.375 lb. of carbon contained in the CO_2; B equals 0.625 lb. of carbon contained in CO.

However, in the original combustion, only that portion of the 1 lb. of carbon that entered into combustion is represented by $A + \frac{B}{2}$ equals 0.6875 lb. and the remainder $\frac{B}{2}$ equals 0.3125 lb. was reduced without reduction of CO_2.

From this standpoint the heat developed from a lb. of carbon is

$A + \frac{B}{2}$ develops 0.6875 x 14,550	= 10,003.125 Btu.
$\frac{B}{2}$ absorbs 0.3125 x 5850	= -1,828.125 Btu.
Net Btu. developed	= 8,175.000 Btu.

There are other conceptions of the method by which heat is developed as indicated by the CO_2 to CO ratio, but regardless of the intermediate steps that may have taken place from the initial CO_2 to the final CO_2 to CO ratio, the result is always the same. We may consider that the heat is developed directly by the complete combustion of the "A" portion of the carbon and by the incomplete CO portion, thus:

A develops 0.375 x 14,550	= 5,456.25 Btu.
B develops 0.625 x 4,350	= 2,718.75 Btu.
Total heat developed	= 8,175.00 Btu.

It may be assumed that the total lb. of carbon is developed into CO_2 and the "B" portion causes a loss in the ratio of 10,200 Btu. to the original lb. of carbon, thus:

A + B Potential,	1.0000 x 14,550	= 14,550 Btu.
B Loss,	0.625 x 10,200	= 6,375 Btu.
Total heat developed		= 8,175 Btu.

A still further application may be

Table 37

ATOMIC WEIGHT, HEAT OF FORMATION AND TEMPERATURE OF IGNITION

A

ATOMIC WEIGHT

Element	Symbol	Atomic Weight—Exact	Atomic Weight—Ordinarily Used
Calcium	Ca	40.08	40
Carbon	C	12.01	12
Chromium	Cr	52.01	52
Hydrogen	H	1.008	1
Iron	Fe	55.84	56
Manganese	Mn	54.93	55
Nitrogen, Pure	N	14.008	14
Nitrogen, Atmospheric	Na	14.075	14
Oxygen	O	16.00	16
Phosphorus	P	30.98	31
Silicon	Si	28.06	28
Sulphur	S	32.06	32
Tellurium	Te	127.61	127.6
Titanium	Ti	47.90	48

Heat for Formation Slag

Temperature calcining $CaCO_2$	above 1475
Heat to calcine 1 lb. $CaCO_2$	771 Btu.
Heat to melt and superheat slag	900 Btu.

B

HEAT OF FORMATION

Molecular Symbol	Molecular Weight	Btu. per Lb.—Mole	Btu. per Lb.—Compounds	Btu. per Lb.—Element
CO_2	44	174,600	3954	14,550
CO	28	52,200	1864	4,350
$CO+\frac{1}{2}O_2$	44	122,400	2782	10,200
CaO	56	275,060	4876	6,826
FeO	72	115,740	1608	2,066
Fe_2O_3	160	375,300	2233	3,190
H_2O	18	104,036	5780	52,018
MnO	71	173,700	2446	3,158
SiO_2	60	361,800	6030	12,921
SO_2	64	127,656	1994	3,989
TeO_2	80	392,400	4905	8,175
Fe_3C	180	33,840	188	2,820
$CaOSiO_2$	116	680,220	5864	12,146
Heat Loss in Decomposition				
$CaCO_3$	100	−78,786	−788	−1,970
CO_2+C	28	−70,200	−2507	−5,850

C

TEMPERATURE OF IGNITION

Substance	Temperature, ° F.
Coke	1300
Anthracite Coal	1300
Bituminous Coal	620
Lignite Dust	300
Hydrogen	1075 to 1090
Illuminating Gas	1075 to 1090
CO	1190 to 1215
Water Gas	1190 to 1215
Producer Gas	1190 to 1215

that if the original pound of carbon had entered into complete combustion there would have been 14,550 Btu. However, the $\frac{B}{2}$ portion which enters into the reduction of CO_2 would represent a loss at the rate of 5,850 Btu. per lb., and the second $\frac{B}{2}$ which did not reach the combustion zone represents a total loss, thus:

A + B Potential,	1.0000 x 14,550	= 14,550 Btu.
$\frac{B}{2}$ Loss,	0.3125 x 5800	= -1828.125
$\frac{B}{2}$ not burned,	0.3125 x 14550	= -4546.875
		-6,375 Btu.
Total heat developed		= 8,175 Btu.

Miscellaneous Sources of Heat. Many minor sources of heat may be

Table 38

WEIGHT AND VOLUME OF THE ELEMENTS ENTERING INTO AND RESULTING FROM THE COMBUSTION OF 1 LB. OF CARBON

Complete Combustion = air + carbon = CO_2 + N
Incomplete Combustion = air + carbon = CO + N

Composition of Air:	*Weight*	*Volume*
Oxygen	23.20%	20.99%
Nitrogen	76.80%	79.01%

One Cubic Foot of Air Contains:

Oxygen	1.000 × 20.99% =	.2099 cu. ft.	× .08921 =	.018725 lb.
Nitrogen	1.000 × 79.01% =	.7901 cu. ft.	× .07844 =	.061976 lb.
		1.0000 cu. ft.		.080701 lb.

COMPLETE COMBUSTION

air + carbon = CO_2 + N + 14,550 Btu.

Chemical Formula:

Molecular Weight ..... $\frac{C + O_2}{12.01 + 32} = \frac{CO_2}{44.01}$

Combining Weights ..... 1 lb. C + 2.664446 O_2 = 3.6644 lb.

	Cu. ft.	*Wt. per cu. ft.*	*Weight, lb.*
Air = $\frac{2.664446}{.0187251}$ =	142.293 ×	.080701 =	11.483
Nitrogen = 142.29275 × 79.01 =	112.425 ×	.07844 =	8.819
Oxygen = 142.29275 × 20.99 =	29.867 ×	.089210 =	2.664
Carbon			1.000
CO_2 $\frac{3.66446}{.123402}$ =	29.695 ×	.123402 =	3.664
Total — CO_2 + N	142.120		12.483
Per cent CO_2	20.89%		29.35%
Per cent N	79.11%		70.65%

INCOMPLETE COMBUSTION

air + carbon = 2CO + N; or CO_2 + C = 2CO

Chemical Formula:

Molecular Weight (44.01 + 12.01) .. $\frac{2C + O_2}{24.02 + 32} = 2CO$ = 56.02

Combining Weights ..... 1 lb. C + 1.3322 O_2 = 2.3322 lb. CO

	Cu. ft.	*Weight per cu. ft.*	*Weight, lb.*
Air = $\frac{1.332223}{.0187251}$	71.146 ×	.080701 =	5.741
Nitrogen = 71.14637 × 79.01 =	56.213 ×	.07844 =	4.409
Oxygen = 71.14637 × 20.99 =	14.934 ×	.089210 =	1.332
Carbon =			1.000
CO = $\frac{2.33223}{.078051}$ =	29.881 ×	.078051 =	2.332
Total: CO + N	86.094		6.741
Per cent CO	34.71%		34.59%
Per cent N	65.29%		65.41%

divided into two classes:

First, the sensible heat contained in the materials which make up the charges;

Second, the heat resulting from the exothermic reactions in the formation of FeO, MnO, SiO_2, P_2O_5, SO_2, $CaSiO_2$, etc.

In a series of tests made in Germany and published in *Stahl und Eisen* (1927), p. 1389, the following thermal input in percentage is given:

	Per cent	
Sensible Heat—Atmospheric Temp.		
In iron charge	0.27	
In coke	0.05	
In moisture of coke charge	0.02	
In limestone	0.08	
In oxygen	0.12	
In nitrogen	0.65	
In moisture in the air	0.01	1.20
Combustion of Carbon:		
CO_2	76.80	
CO	14.70	91.50
Miscellaneous Oxidation:		
FeO	0.43	
MnO	0.38	
SiO_2	5.84	
P_2O_5	0.58	
SO_2	0.07	7.30
		100.00

In the tests made by Piwowarsky and Broglio, well-known experts in the preparation of cupola heat balances, they report the following heat input:

Heat Source	*Per cent*
Heat from coke	93.95
Oxidation of iron	4.96
Sensible heat in the charge	1.06
Other exothermic reactions	.03
	100.00

In the above reports the oxidation of the iron and other exothermic reactions are much larger than permitted in good practice. However, these quantities vary according to the length of the heat.

It is common experience that the metal loss in the last charge melted is very much greater than in the intermediate charges, hence when this loss is spread over the entire heat the length of the heat has a large bearing on the percentage of oxidation.

The object in presenting the result of these tests is to show the variety of sources for the development of heat rather than to indicate the quantity.

Effluent Cupola Gases. The gases emitted from a cupola stack are not only the pure gases resulting from the combustion of carbon, but also contain an appreciable volume of water vapor originating from the 0.04 to 0.10 lb. of water in each pound of coke, which, when expelled, results in from one to two cu. ft. of water vapor.

In addition to this, there is a variable amount of water vapor entering the tuyeres depending upon the relative humidity of the air blast, and a further increment arises from the combustion of hydrogen.

Altogether, the water vapor may increase the theoretical volume of gases resulting from the combustion of a pound of carbon by 3 or 4 cu. ft.

Other miscellaneous gases such as SO_2, FeO, MnO and SiO_2 are present as indicated by the color of the flame.

These reactions require an amount of oxygen in addition to that used for the combustion of the coke. The amount of excess oxygen used for this purpose is about 4 per cent, as indicated in the German

reports recorded under the subject, "Miscellaneous Sources of Heat", p. 538.

This is probably a little excessive for heats extending over several hours.

The oxygen used for this purpose is not taken into account when making gas analyses. However, the nitrogen which accompanies the oxygen passes on to increase the volume of nitrogen above the theoretical ratio.

The increase in CO_2 originating from the limestone which is used as a flux is a small but definite amount.

Assuming that 1 lb. of limestone is used to each 10 lb. of coke, and that one-half of the weight of the limestone consists of CO_2, then each pound of carbon consumed would release from 0.4 cu. ft. to 0.6 cu. ft. of CO_2.

In all, there may be 4 or 5 cu. ft. of miscellaneous gases in addition to the theoretical amount.

In the analysis of the effluent gases for commercial purposes in connection with cupola control and cupola efficiency, the only gases considered are CO, CO_2 and N. The percentages of CO and CO_2 present are first determined and the remainder recorded as nitrogen.

Consequently the percentage of CO is smaller than the theoretical 34.71 per cent and the CO_2 is not affected to as great an extent on account of the increase from the limestone. The balance which is called nitrogen, however, is decidedly increased on account of the dilution with other gases.

For simplicity in the preparation of tables recording the air requirements for combustion and resultant gases, and at the same time to more closely approach the real facts, the air requirements are based upon a resultant gas containing 33⅓ per cent CO and 66⅔ per cent nitrogen in the case of incomplete combustion, and 20 per cent CO_2 and 80 per cent nitrogen in the case of perfect combustion.

There always will be some excess air used because there is always a slip between the theoretical equation and the actual performance.

To avoid the use of fractional cu. ft. of CO and CO_2 developed by the combustion of one lb. of carbon, 30 cu. ft. will be used, which is within one-half of 1 per cent in excess of the theoretical for the combustion of carbon. In addition to this, 5 per cent excess air is used in Table 24 on page 530 to allow for miscellaneous oxidation.

On this basis the amount of carbon that is used for the reduction of CO_2 to CO and the amount burned to CO_2 in each pound of carbon consumed is readily determined mathematically, as shown in Tables 27 and 28 (pp. 533 and 534).

As explained, the quantities in Table 24 therefore are not the theoretical quantities developed by the combustion of pure carbon, as shown in Table 38. They are, however, closer approximations to true condition as verified by experience.

It is always found that the weight of the air required for combustion is greater than the theoretical amount. The quantities cannot be fixed by discussion. They are simply matters of agreement.

When calculations are made for blower capacity, the air quantity shown in the table should be increased by at least 25 per cent. This will allow for the expansion of the air up to a temperature of 100 degrees at the sea level. For high altitudes, a further consideration should be given to the increased expansion on account of reduced pressure.

For the sake of those mathematically inclined, the quantities in the tables have been calculated to several decimal places beyond any practical utility. The final result in any calculation cannot be of greater accuracy than the approximate data used at the beginning. For ordinary calculations, one or two decimal places are far within the accuracy required in any cupola problem.

The temperatures of the escaping gases from the cupola are variable according to the point at which the observation is made. The temperature is higher at the outer circumference where the resistance to the flow of gases is least and their velocity greatest.

At the center of the cupola, the velocity of the gases is impeded by the more closely packed iron. Because of the greater time involved and the large area of surface contact the temperatures are lower.

The weight of the gases fluctuate continuously to synchronize the time of melting a charge of iron to the time of burning a charge of coke. Hence the gas analysis is a variable throughout the range of fluctuations. Furthermore, the weight of the iron and the weight and analysis of the coke are not taken with great precision, neither is the temperature of the iron at the time of tapping known with mathematical precision.

In making the calculations for heat balances where all these variables are involved, the best that can be done is to work within the limits of the observed data. Mathematical precision of a theoretical formula does not add to the exactness of the solution.

Thermal Capacity

Thermal capacity of a substance is the amount of heat required to raise a unit of weight a unit of temperature.

In the English system of weights and measures the unit of weight is 1 lb. and the unit of temperature is 1° F.

The unit of quantity of heat is designated as a British Thermal Unit, which is the quantity of heat required to raise 1 lb. of water from 39° to 40° F. (maximum density). As the temperature of water is elevated, its density decreases.

The amount of heat required to raise 1 lb. of water one degree at various temperatures is shown in Table 18 on page 529.

Heat capacities are determined in the laboratory by means of a calorimeter, hence most determinations are made at standard room temperatures of about 68° F. The quantity of heat is measured by a change in temperature of the product.

Specific Heat

Specific heat of any substance is the ratio of its thermal capacity to that of water at 60° F., hence specific heat and thermal capacity are the same quantity when considering a unit of weight and one degree temperature change.

Specific heat is an important factor in the calculation of temperature and heat transfer in the various parts of the cupola. When the quantity of heat in any substance is known its temperature is readily calculated and vice versa if the temperature is known, the quantity of heat becomes known.

Inasmuch as the specific heat is not constant at various temperatures, ready access to tables made for the purpose greatly facilitates the calculation of temperature gradients and heat transfer.

The principal items to be considered from the standpoint of heat capacity are the gases entering into, and resulting from, combustion, the iron, the coke, and the slag.

Heat Capacity of Gases

Complications arise in regard to gases of combustion on account of variable percentages of CO_2, CO and nitrogen having variable specific heats which do not follow the same law of increase as the temperature increases; and, further, the total weight of the gases varies with the stage of combustion.

There are, therefore, four variables to contend with:

First, the specific heat of the individual gases at zero degrees temperature.

Second, the rate at which the specific heat increases as temperature increases.

Third, the percentage by weight of each of the gases.

Fourth, the total weight of the gases developed from the various stages of combustion of 1 lb. of carbon.

There are many formulae recommended for the various gases by various investigators, any one of which is sufficiently accurate for ordinary use.

The formulae recommended by the National Bureau of Standards for the various gases entering into combustion have an allowable error of plus or minus 2 per cent within the range of temperatures developed in the cupola.

For the sake of simplicity, and still to be within this allowable range of error, the formulae taken from Richards' *Metallurgical Calculations* are chosen as the basis for Table 32 (p. 535), in which mean specific heats are recorded for various combinations of the gases of combustion that are developed within the cupola corresponding with the various stages of combustion.

The formulae are made up of a constant at zero temperature plus a uniform rise for each degree from zero to 3600° F., also a formula applying between 3600° and 7200° F. See Table 19 on page 529.

Mean Specific Heat

The above formulae determine the specific heat at any particular temperature. They do not indicate, however, the amount of heat required to raise the substance from one specified temperature to another.

For this purpose, the mean specific heat between the limits of the temperatures under consideration becomes a more usable factor. In this case where the increase is uniform the mean specific heat is one-half of the sum of the specific heats for the temperatures under consideration, consequently the constants at zero plus one-half the increase per degree will represent the mean specific heat between zero and the desired temperature. See Table 20.

The quantity of heat in Btu. required to raise 1 lb. of gas from 0° to a specified temperature (0 to t) is shown in Table 21 (p. 529).

The quantity of heat in Btu. required to raise 1 lb. of gas from one specified temperature to another (t' to t) is shown in Table 22 (p. 529).

The mean specific heat between two specified temperatures is shown in Table 23, page 529.

The foregoing formulae are self-explanatory.

Table 32 records the average mean specific heat from zero to any specified temperature for all temperatures and all combinations of gases developed in cupola practice.

The mean specific heats at various temperatures and at the various stages of combustion shown in Table 32 represent the average Btu. in each pound of gas for each degree of temperature rise from zero to the specified temperature.

The quantity of heat in terms of Btu. contained in 1 lb. of gas for any stage of combustion is found by multiplying the mean specific heat by the temperature of the gases in degrees Fahrenheit. These quantities are shown graphically in Fig. 14 on page 528.

The tables and charts are not intended for scientific laboratories engaged in the development of great accuracy in heat constants. They are, however, far within the degree of accuracy required for use in developing heat balances based upon observed temperatures, gas analyses, and estimated weights of air entering into combustion, and the gases resulting from combustion.

Flame Temperature

Having the weight of the gases, their mean specific heat and quantity of heat developed from combustion of 1 lb. of carbon preheated to approximately 3,000° F. is easily calculated by the following formula:

$$\frac{\text{Btu. Developed}}{\text{Lb. Gases}} \times \frac{1}{\text{Mean Specific Heat}} = \text{Temperature of Gases in Combustion Zone}$$

Heat Capacity of 1 Lb. of Iron

Efficiency of heat transfer and the overall thermal efficiency of the cupola is based upon the quantity of heat in one pound of iron at the time of tapping.

The specific heat of iron varies with increasing temperature and is modified by a change in state at several critical points, also by the latent heat of fusion when in the molten state. On account of these variables a convenient table of reference (Table 6) was included in a paper prepared by the author under the subject, "Conservation of Heat Energy in Cupolas," and presented to the World's Power Conference held in Stockholm in 1933.

This table includes the entire iron carbon group, giving the heat of recalescence, melting temperatures, specific heat, latent heat of fusion, latent heat of vaporization, and heat capacity of one pound of iron at various temperatures in the solid, liquid and gaseous state. These quantities are based mainly upon Richards' *Metallurgical Calculations,* with the exception that the specific heat of liquid iron is placed at 0.23 instead of 0.20, which agrees more nearly with the more recent determination.

There is less reliable data on the latent heat of fusion for iron of various carbon contents than for the other items tabulated. There is, however, a close relation between the temperature of fusion and the latent heat of fusion. The quantities given are sufficiently accurate for use in the calculation of problems relating to cupola efficiencies. The heat capacities of one pound of iron have been determined experimentally and are recorded in Tables 29, 30, 31 (pp. 534-535).

SKILL AND LEADERSHIP THROUGH APPRENTICE TRAINING

An apprenticeship system, to be successful, must be considered by management as an essential part of the business. If apprentice training is included in company activities, it must be given due importance if it is to produce satisfactory results.

J. E. Goss
Brown & Sharpe Mfg. Co.
Providence, R. I.

A MEETING of foundry owners and managers was held in a large Eastern city for the purpose of discussing apprenticeship training problems. During the discussion one man asked—"How many of us are prepared for the day when younger men must take our places"? Not one man at that meeting could say that he was ready for such a time.

The question referred to young men with good training and journeyman experience in foundry practice, plus good training and experience in foundry management. It has been proved that the best way to prepare foundry management personnel is by means of apprenticeship. The men at that meeting seemingly had not done a good job in this respect; in fact, several of them so admitted.

Purpose of Apprenticeship

Primarily, of course, apprenticeship is set up to train young men in molding and coremaking. Any other main objective would remove such training from the apprenticeship category. We shall continue to need more than management personnel. Try as we may to specialize our processes, there still will remain many foundry jobs calling for a variety of skills and a general knowledge of foundry casting practice.

If we subordinate the training for the work of journeymen to the training for management duties, the result will be that we shall produce neither good journeymen nor good managers. A practical training background is a definite asset to management. Also, management can function to better advantage if it has producers with the same kind of background.

If we agree with such a concept of foundry apprenticeship, we will predicate our plans upon a system which should give us both journeymen and management personnel. The required ratio of one to the other can be estimated roughly, and our system can be set up to come close to meeting these requirements.

Public Relations

Item number one in a foundry apprenticeship system should be a form of public relations. The definition commonly given to public relations will be recalled. It is the accomplishment of something good, followed by giving information about it to all who may be directly or indirectly interested.

Our first concern, then, is to place ourselves, if we are not already there, in a position where we can conscientiously tell about the foundry as a good place in which to work. This will not be easy. The mere word "foundry" seems, to many people, to be synonymous with dirt and danger, heat and heavy lifting.

Our public comprises parents of high school boys, school teachers, vocational advisors, parent-teacher groups, and the boys themselves. In planning to acquaint these people with opportunities in the foundry, we should bear in mind what seems to be an almost universal practice. It is the steering of the more intelligent boys toward college. The value of a college education need not be discounted in pointing out the opportunities for such boys in the foundry industry.

Source of Supply

What we are leading toward, in this so-called public relations work, is an adequate and satisfactory source of supply. By this time, probably, we all have recognized that concern about source of supply and careful testing of raw material should not be confined to pig iron, sands, binders, facings, and other such items.

The high schools and trade schools offer the most fertile fields, but in many high schools will be found the tendency on the part of principals and advisors to work toward a college education for the boys with good scholastic records.

Acquainting the key people in the secondary schools with possibilities in the foundry, particularly for such boys, may be done by talks to groups, at the schools, and by literature describing the training facilities and pointing out the kinds of work to which such training may lead.

To be effective, this publicity must not paint a rosy picture or present claims instead of facts. The new apprentice, if he is the type of boy we want, will be quick to compare what he has been told with the actualities he finds in the foundry. A large number of beginners and a large turnover will mean waste and, perhaps, discouragement over an ap-

Presented at an Apprentice Training Session of the Fiftieth Annual Meeting, American Foundrymen's Association, at Cleveland, May 9, 1946.

prenticeship system. It is much better to attract fewer boys and to have them carry on their training.

When we test our incoming raw materials we do so with their definite uses in mind. When we test our raw apprenticeship material we are not usually testing competently selected material, as is generally the case with our irons and sands, nor can we know exactly to what uses this material eventually may be assigned. We can, however, come somewhere near the mark in determining our needs for both journeymen and supervisors during the next few years. Also, we can be mindful of what our management requirements may be some years thereafter.

Pre-employment tests for apprenticeship should be a distinct help to the foundryman, provided that he chooses the tests carefully, administers them properly, and then continues to use his judgment of people as an important factor in the hiring process.

System Planning

Selling foundry apprenticeship to individuals and groups requires the planning and carrying out of a system which can be explained in detail and with justifiable pride. We would not start our sons in a school or college on the mere assurance that they would be given an education. We would want to know what courses were offered and something about the content of each. We would inquire into extra-curricula activities, library facilities, and other means by which a predetermined objective could be attained.

A carefully thought out foundry program for apprentices is an advantage to all concerned. We want to produce journeymen molders with all-around training and experience; therefore, we must see to it that our apprentices are given, in an orderly, systematic way, what we believe to be the proper proportionate period of time on each phase of the molder's job.

Perhaps the best way to accomplish this is by centralization of apprentice training. The foundry superintendent and his foremen are interested principally in the production of castings. Any such duty as moving apprentices from one kind of work to another would be secondary and, therefore, sometimes neglected. But, foundry superintendents and foremen should never be left out of an apprenticeship system. It is important that they be a part of it, although it should not be expected that they will set aside their urgent production problems in favor of apprenticeship routine.

Apprenticeship Department

Centralization of apprenticeship training should mean better job instruction and better selection and presentation of class work. Moreover, it can be of assistance in the hiring of apprentices, if such hiring is done by a personnel department. It definitely can help in the placement of graduate apprentices and in the selection of foremen.

If the size of a foundry does not warrant the setting up of an apprenticeship department, much can be done by the appointment of a competent individual whose time may be devoted entirely to apprentice training. If the number of apprentices is too small to warrant the full time of a man, an arrangement might be made to assign a man, competent to supervise apprentices, to a job from which he can take whatever time is necessary for his training duties.

The importance of having the right man as a supervisor of apprentices can not be too greatly stressed. He should have much more than a thorough knowledge of the way in which good castings are produced. He must be able to impart that knowledge in readily understood terms and analyze jobs so that they can be taught in the order of their learning difficulties. He must decide on the variety of work that will give his apprentices the best experience and have the good will of the foreman so that such variety will be made available to the boys.

Supervisory Qualities

In addition to his techniques and their application to his work, he must be patient, forgiving to a reasonable degree, interested in youth and possess an understanding of people sufficient not only to enable him to grade apprentices but also to sense their potentialities.

Probably no one program of foundry assignments would fit any two apprenticeship systems, yet in determining our assignments and the length of time to be spent on each, we should bear in mind that our apprentices are to be taught a trade. They rightly should expect that an apprenticeship in one foundry will qualify them to readily adapt themselves to the work in another foundry producing castings of the same general classification.

Whether an apprentice is taught machine molding before he is taught bench molding may not be essential, but is is essential that he learn both molding methods during his apprenticeship time, since they are methods of making castings and, therefore, are parts of the molding trade.

Some foundrymen may hesitate to take apprentices because their regular facilities would not permit full training. If there are two or more foundries in a locality, an arrangement might be made whereby training which could not be given in one place would be available in another. Such a plan has proved to be quite practicable in at least one locality, in the machinist trade.

Supplemental Training

If an exchange arrangement is not possible or feasible, there is another way to give the apprentice a comprehensive program. This may be done by setting up, somewhere away from the manufacturing departments, whatever equipment is needed to make the training complete. Short, intensive instruction could be given and, although the values that we attach to a reasonable amount of repetitive work would be lacking, we should be able to feel that we had made up, at least in part, for whatever training our regular production practices did not permit.

An arrangement such as this has been working successfully in a large factory for many years, where it was found that it was not practicable to give certain instructions connected with the machinist trade in the manufacturing departments.

If we are to carry on an apprenticeship system that will produce competent journeymen, we must go further than establishing work assignments. Each assignment, if it is to be thoroughly covered, should be analyzed. Only in this way can we be certain that nothing with learning value has been omitted. This does not mean anything more than a little careful thought, the printing or mimeographing of the items on a pocket-size card, and then the

checking of these items when the apprentice has received instruction and practice in what they represent.

Usually, in apprenticeship, there are wide differences of opinion as to what should be taught the boys aside from actual molding practices; also, as to how and where this instruction should be given. These questions, seemingly difficult to answer satisfactorily, no doubt constitute the reason why some foundrymen, especially those operating the smaller places, are reluctant to train apprentices.

The trend appears to be in the direction of letting the public school system assume the responsibility for this so-called related work, even to the point of deciding the subjects to be taught.

Related Work

It must be admitted that circumstances make a considerable difference as to whether the foundry itself or the public school is selected as the place of training, but it would seem that the selection of subjects should be made on the basis of needs in the molding trade rather than on the general educational level so often found in cooperative programs.

Over the past two or three decades there has been a trend toward liberalization of subject matter. It may be true that a journeyman molder is better off, generally, if he knows something about physics and chemistry, but might we not just as reasonably include many other subjects more or less remotely connected with the educational needs of the average man?

This may be the appropriate place to express a word of caution about getting too far away from the fundamental principles of apprenticeship. For one thing, we are quite apt to use the word loosely, applying it to various forms of foundry training not designed or expected to produce journeymen molders.

Short-Term Programs

Today, in our efforts to cooperate with the Veterans Administration and other service organizations, we may incline still further toward short-term programs and think of them as apprenticeships. Specialized courses have their place in the foundry, but they do not supplant apprenticeship.

Apprenticeship calls for an agreement between the employer, the apprentice, and his parent or legal guardian. Otherwise, the training is not really apprenticeship, nor does it present either to the employer or his apprentice the obligation which both rightly should assume.

Unless a written agreement is made a definite part of our apprenticeship system, we are not conforming to accepted apprenticeship standards. In addition to the legal aspect, there is a certain psychological effect. It probably could be shown that employers who use an agreement fare far better than do others in the matter of keeping their apprentices.

Every right thinking foundry worker should understand that management must choose for its supervisory personnel the best among available candidates, regardless of whether or not those chosen are graduate apprentices. If we do a good job of selecting and training our apprentices, it is likely that at least some of them eventually will become good leadership material.

It would be a distressing commentary on an apprenticeship system that showed no supervisors and foremen rising out of the ranks of its own graduates. A poor management job would be indicated, and it would grow increasingly difficult to interest prospective apprentices.

Management Interest

To make an apprenticeship system work, management must continue to see it as an essential part of the business. If management leaves apprentice training entirely to an individual or to an administrative group and shows no further interest, it will fail eventually.

If a survey were made of foundries in which apprenticeship has bogged down or been discontinued, it probably would be found that top management was largely at fault. Management itself can best decide, of course, whether the nature of the work done in its foundry warrants the training of apprentices, but if such training is included in company activities it must be given due importance if it is to produce satisfactory results.

Perhaps some foundry managements hesitate about starting an apprenticeship system because of their unfamiliarity with this kind of training. Men experienced in this work have heard others say that apprentice training presented such problems and obligations, particularly to the smaller foundry, that it had not been incorporated as a means of skilled worker supply.

If such is the feeling today of any foundry management, it is only necessary to write the A.F.A. National Office for advice as to how to start and carry on the apprentice form of training. The A.F.A. Committee on Apprentice Training is the second oldest committee in the Association. Although its members are widely separated, it transacts much business throughout the year by using the mails. Fifty years have served not to make it old and inactive, but rather to vitalize and have it become recognized as an agency alive to its opportunties.

Twenty years ago apprenticeship often was referred to as dead. Today we note among industrialists an awareness of the real value of apprenticeship, and it is such work as has been done by A.F.A. that has contributed largely toward this change.

GRAY CAST IRON

Tensile Strength, Brinell Hardness and Composition Relationships

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• Tensile strength, as one measure of gray cast iron quality, and Brinell hardness and chemical composition as alternate specifications for gray iron engineering applications do not necessarily result in irons produced under various conditions being of the same quality and possessing the same properties. The need for a simple method of correlating and interpreting these properties of gray cast iron to permit of more accurate applications is apparent.

STANDARD methods for measuring the tensile strength of gray cast iron have been established for some years and are accepted by both foundrymen and designing engineers as one measure of the quality of gray iron castings. Furthermore, specifications on gray cast iron used for engineering application are usually either based on or include the values for the tensile strength ranging from 20,000 to 80,000 psi., depending upon the application.

Therefore, any discussion of tensile strength of gray cast iron, at the present time, would seem to be unnecessary and entirely repetitious. However, there are still some misunderstandings regarding the proper use of the tensile strength value for the purpose of interpreting the adaptability of gray iron castings and the interpretation of test bar results into engineering design.

The use of a tensile strength specification for gray cast iron has been criticized from time to time, on the basis that gray iron castings are not normally subjected to a tensile load, or at least to a load in any way commensurate with the sections involved. For this reason, both chemical composition and Brinell hardness have been used as alternate specifications by mutual agreement between the producer and the consumer.

Here again, however, misunderstandings may occur because of the fact that such specifications do not necessarily result in the production of equal quality irons from different foundries. Unfortunately, two irons of the same Brinell hardness, or even of the same chemical analysis, do not always exhibit the same mechanical properties, structure, or satisfactory combination of properties when produced under various conditions.

Mechanical Properties

It is believed that the reasons for observed variations in the quality of irons of either identical tensile strength, Brinell hardness number, or chemical composition require some discussion. There is an equal or greater need for some reasonably simple method of correlating and interpreting these properties of iron in order to facilitate more nearly correct applications of irons of different characteristics.

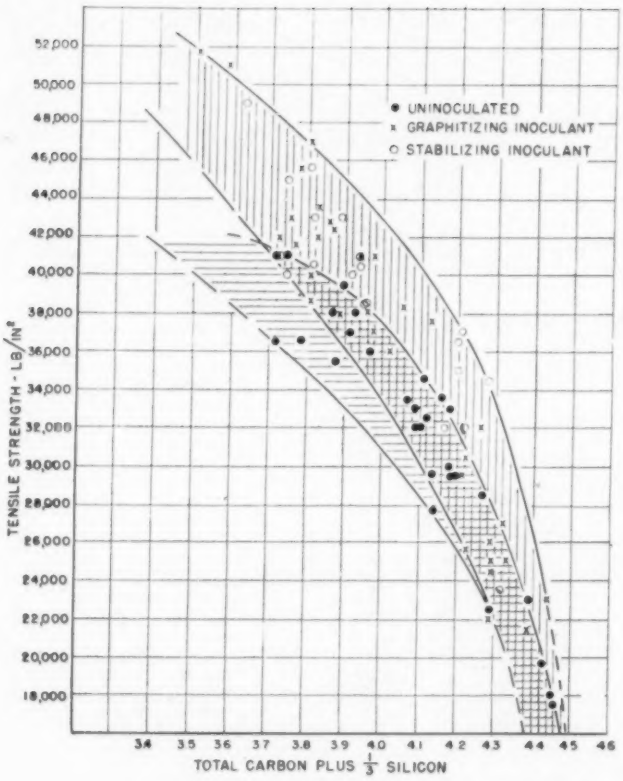
Basically — and practically — gray cast iron is not a material; it is a generic name for a "family" of materials possessing certain related characteristics.

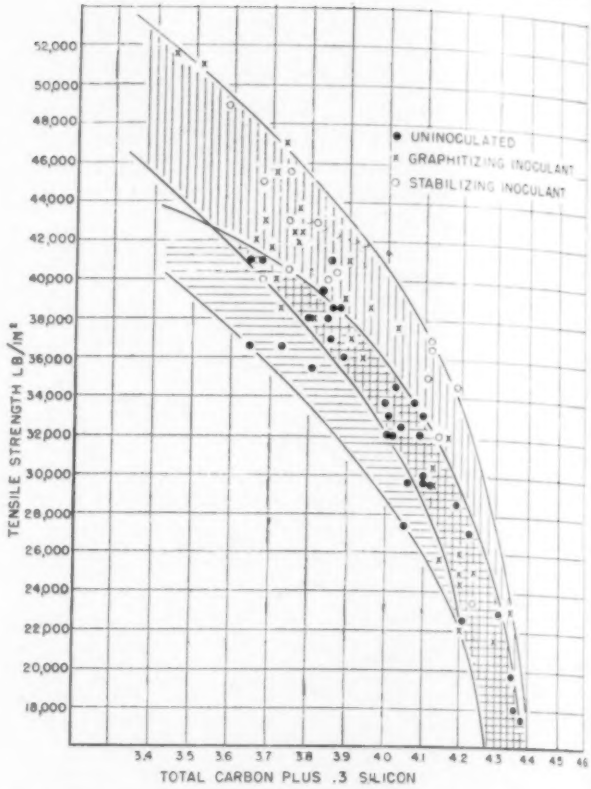
In most cases, the tensile strength specification is not utilized as an indication of the ability of a given casting to resist a tensile stress. Its greatest value is as an indication of the other qualities of the gray cast iron which are required in any specific application.

For example, a casting consumer may rightly feel that a gray iron casting of 50,000 psi. tensile strength will be more satisfactory than one having a tensile strength of 20,000 or 25,000 psi. in an application in which the casting is submitted to various types of stresses, but in which no tensile stress is involved.

Unfortunately, the use of the tensile test as a measure of quality of gray cast irons does not take into consideration the fact that irons of a given tensile strength can and do vary widely in quality and behavior under other conditions, depending upon the chemical composition and the method of manufacture. For example, an iron produced by good melting technique and with a given "balanced" chemical analysis may exhibit a tensile strength of 45,000 psi. with a low Brinell hardness

Presented at a Symposium on Engineering Properties of Gray Iron Session of the Fiftieth Annual Meeting, American Foundrymen's Association, May 8, 1946, at Cleveland.





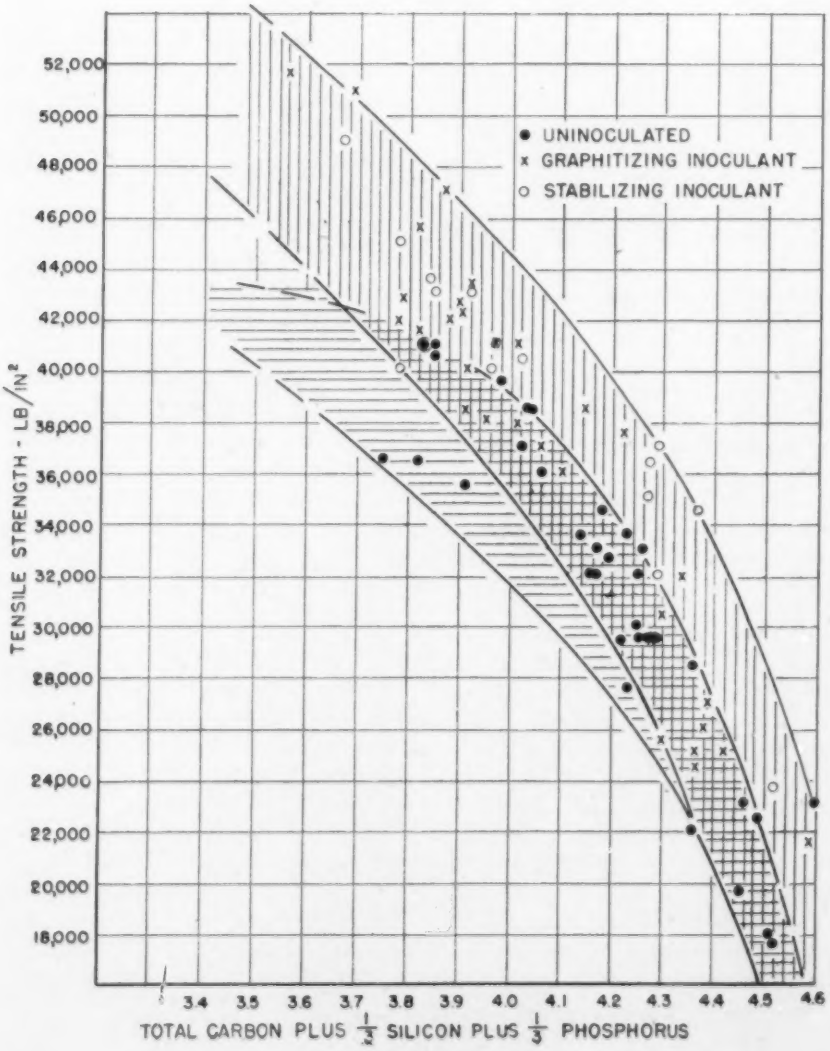


Fig. 1 (Top left)—Relationship between tensile strength and carbon equivalent (carbon plus 1/3 silicon) of gray cast iron.

Fig. 2 (Above)—Tensile strength and carbon equivalent (carbon plus 0.3 silicon) relationship. **Fig. 3 (Bottom left)**—Carbon equivalent (carbon plus 1/3 silicon plus 1/3 phosphorus) and tensile strength relation.

number and be relatively easy to machine.

On the other hand, another iron produced by a faulty melting technique, but of a composition which should permit of a tensile strength of 60,000 to 70,000 psi. may readily have the same tensile strength of 45,000 psi., but possess a high Brinell hardness number, and relatively poor machining characteristics, and be unsatisfactory in wear-resistance.

The use of the Brinell hardness number as a measure of quality is frequently unsatisfactory for the same reasons described previously. Such a test could only be entirely satisfactory as a measure of the quality of iron, if all the other factors of production were known and could be interpreted correctly. It is no news to most engineers that Brinell hardness number does not necessarily indicate the degree of machinability of gray cast irons made by different foundries and under dissimilar conditions.

The use of chemical composition

alone, for the purpose of defining the essential characteristics of gray cast irons, fails for reasons similar to those applied to tensile strength and Brinell hardness.

Two or more identical gray iron castings of identical composition may differ in physical and mechanical properties to the extent that an individual foundry may meet a chemical specification and, at the same time, inadvertently produce an unsatisfactory product from the standpoint of the engineer or consumer.

Relationship Between Tensile Strength and Chemical Composition

In an effort to clarify relationship between the tensile strength and other properties of gray iron castings, the authors have made an attempt to show the extent of correlation between tensile strength, Brinell hardness, and composition with the aid of graphs plotted from information obtained on commercial gray cast irons produced by members of the Gray Iron Research Institute, Inc.

Perhaps the most important relationship is that between tensile strength and composition, since it offers a convenient measuring stick "for the foundryman, the engineer, or both."

Cast Iron Test Bars

For the purpose of this study, 82 commercially produced cast iron test bars were chosen at random from the hundreds of bars made available through a cooperative research effort being conducted at Battelle Memorial Institute. The test results and the chemical analysis of the irons used were obtained by the individual foundry in the course of routine control without previous knowledge of the fact that the results would subsequently be used for special study like the one undertaken.

After the relationships which follow were established, their accuracy was demonstrated by utilizing information obtained from experimental gray cast iron melts, from other commercially produced gray cast iron melts, and from the data contained in the A.F.A. publications, Cast Metals Handbook and Alloy Cast Irons, as well as in other handbooks.

Carbon Equivalent Concept

The most logical approach toward establishing the relationship between tensile strength and chemical composition involved the use of what is known as the carbon equivalent concept. The method of presentation utilizes both the total carbon and the silicon contents of gray cast iron in the form of a single value.

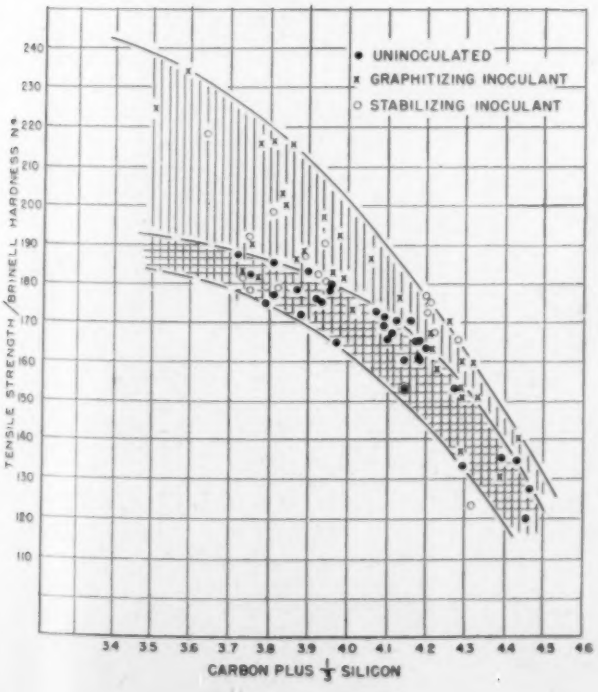
Although the percentages of total carbon, of silicon, or of carbon plus silicon have no significance in themselves in demonstrating a relationship between strength and composition, these values can, when utilized in accordance with the carbon equivalent concept, give a value, expressed in percentage, which may be plotted against the values for tensile strength or for other physical properties to show a simple relationship between composition and a property of the iron.

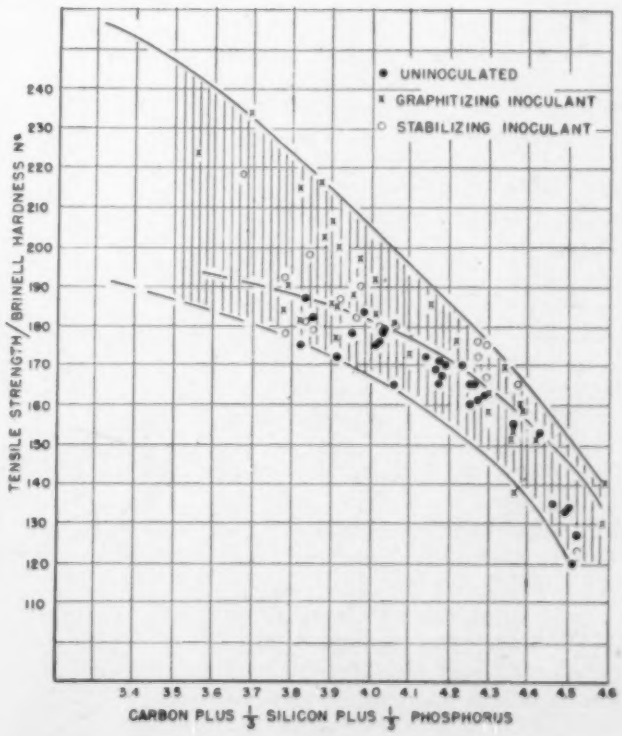
Alloying Elements

In the application of the carbon equivalent concept, normal manganese and sulphur contents must be assumed and the iron must be substantially free from alloying elements such as nickel, copper, chromium, molybdenum, and vanadium. However, similar relationships may be established for the alloyed irons when the amount and type of alloy is known and when its effect for any given carbon equivalent has been determined.

The published information on carbon equivalents indicated that there are several ways in which they may be calculated from the total carbon and silicon contents. In this work, five of these formulas have been

Fig. 4 (Below)—Relationship of tensile strength-Brinell hardness ratio to carbon equivalent (carbon plus 1/3 silicon). Fig. 5 (Right)—Tensile strength-Brinell hardness ratio and carbon equivalent (carbon plus 1/3 silicon plus 1/3 phosphorus) relationship.





utilized in order that the degree by which they differ, one from the other, could be determined. It was anticipated that one formula might prove more consistent than any of the other four in providing a relationship between composition and strength.

The gray cast irons used are listed in Table 1 and cover the range of total carbon content from 2.70 per cent to 3.70 per cent with silicon from 1.3 per cent to 3.0 per cent, and phosphorus from 0.06 per cent to 0.80 per cent. The carbon equivalents calculated from these data range from 3.45 per cent to as high as 4.64 per cent. This is considered normal for unalloyed gray cast irons within the tensile strength limits of 20,000 to 50,000 psi.

Although the calculation of the carbon equivalent of a given iron by the five methods permitted variations as great as 0.30 per cent in the ultimate value, it was observed that three of the five methods covered the range of such variations adequately, and that the other two were, to all intents, duplications of one or another of the three.

Carbon Equivalent Calculation

Therefore, in presenting the relationship between the tensile strength of gray cast iron and its carbon equivalent in the form of graphs, only three methods for the calculation of the carbon equivalent were utilized. The resulting graphs are shown as Figs. 1, 2 and 3.

Fig. 1 includes the data on tensile strength plotted against carbon equivalent derived from the formula carbon plus 1/3 silicon (C + 1/3 Si). Fig. 2 was prepared using the formula carbon plus 0.3 per cent silicon (C + 0.3 Si) for carbon equivalent, while Fig. 3 was prepared with the carbon equivalent derived from the formula total carbon plus 1/3 silicon plus 1/3 phosphorus (C + 1/3 Si + 1/3 P). The fields included by the shaded areas in Figs. 1, 2 and 3 are sufficient to include the regions covered by the five formulas for carbon equivalent.

Regardless of the method used for calculating the carbon equivalent, a definite relationship between the tensile strength and the carbon equivalent is obvious from the character of the shaded areas in Figs. 1, 2 and 3. As one would anticipate from the nature of the information, the data fall within zones or bands rather than on lines, even when the data are further divided by differentiating between those irons poured directly into test bar molds and those receiving, just prior to pouring, some type of ladle treatment referred to as inoculation.

It was not considered necessary to differentiate between those irons treated or inoculated with the so-called graphitizing type materials from those treated with the "stabilizing" inoculating treatment, since the difference is readily observed in the graphs.

In order to avoid the possibility that references to inoculated irons may not be understood, the following passage is quoted from the A.F.A. ALLOY CAST IRONS to describe the agents used in the inoculation treatment:

"Inoculants or inoculating alloys are those materials which are added to molten cast irons in the ladle for the purpose of altering or modifying the structure and thereby improving the physical and mechanical properties to a degree not explainable on the basis of the change in composition resulting from their use."

The stabilizing inoculants referred to in Figs. 1, 2 and 3 introduce small

Table 1

CARBON EQUIVALENTS OF 82 GRAY CAST IRONS

Tensile Strength, psi.	*Ratio of Tensile Strength to Brinell Hardness*	*Total Carbon + ⅓ Si*	*Total Carbon + .3 Si*	*Total Carbon + ⅓ (Si+P)*	*Total Carbon + .3 (Si+P)*	*4.3+C −(4.3-.33 Si) · (1-.1 P)*
51,500[1]	224	3.51	3.45	3.56	3.49	3.56
51,000[1]	234	3.59	3.52	3.69	3.61	3.65
49,000[2]	218	3.64	3.58	3.67	3.61	3.68
47,000[1]	216	3.81	3.74	3.87	3.80	3.85
45,500[2]	198	3.81	3.75	3.84	3.78	3.83
45,500[1]	215	3.78	3.71	3.82	3.75	3.82
45,000[2]	192	3.75	3.68	3.78	3.71	3.79
43,500[1]	200	3.84	3.77	3.92	3.85	3.89
43,000[2]	179	3.82	3.75	3.85	3.78	3.86
43,000[2]	187	3.89	3.82	3.92	3.85	3.91
43,000[1]	190	3.76	3.69	3.79	3.72	3.80
42,500[1]	216	3.86	3.78	3.90	3.82	3.92
42,500[1]	186	3.87	3.78	3.90	3.81	3.89
42,000[1]	184	3.73	3.66	3.78	3.70	3.75
42,000[2]	203	3.83	3.77	3.88	3.81	3.88
41,500[1]	182	3.77	3.70	3.82	3.74	3.82
41,000	182	3.75	3.68	3.85	3.77	3.85
41,000	187	3.72	3.65	3.83	3.74	3.82
41,000[2]	181	3.73	3.66	3.83	3.75	3.84
41,000[1]	192	3.98	3.90	4.01	3.93	4.01
41,000[1]	197	3.94	3.86	3.97	3.89	3.96
41,000[2]	190	3.94	3.86	3.97	3.89	3.96
40,500[2]	189	3.82	3.75	3.85	3.78	3.86
40,500[2]	180	3.94	3.87	4.02	3.94	4.01
40,000[2]	182	3.92	3.85	3.96	3.89	3.97
40,000[2]	178	3.75	3.68	3.78	3.71	3.78
40,000[1]	185	3.81	3.72	3.91	3.82	3.90
39,500	183	3.90	3.84	3.98	3.91	4.00
38,500[1]	186	4.06	3.96	4.15	4.04	4.16
38,500	178	3.95	3.88	4.03	3.95	4.03
38,500	179	3.95	3.88	4.03	3.95	4.03
38,500[1]	177	3.81	3.73	3.91	3.82	3.91
38,000[2]	183	3.96	3.90	4.01	3.94	4.01
38,000[1]	188	3.89	3.81	3.95	3.86	3.95
38,000	178	3.87	3.80	3.95	3.88	3.96
38,000	175	3.93	3.85	4.01	3.92	4.01
37,500[1]	176	4.13	4.03	4.22	4.11	4.22
37,000[1]	181	3.98	3.91	4.06	3.98	4.07
37,000	176	3.92	3.86	4.02	3.95	4.02
37,000[2]	175	4.21	4.12	4.29	4.14	4.28

(Table 1 continued on following page)

[1]Treated with graphitizing inoculant.
[2]Treated with stabilizing inoculant.

percentages of the alloying element chromium. Otherwise, there is no change in composition, other than that already indicated by the carbon equivalent. The lowest line in each of Figs. 1, 2 and 3 defines the range of minimum values obtained for the uninoculated commercial irons of the series studies, whereas the second line from the bottom defines the range of minimum values for all of the inoculated irons of the same series.

The top line in each of Figs. 1, 2 and 3 defines the range of maximum tensile strength obtained from the inoculated irons, while the second line from the top defines the range of maximum tensile strength obtained for the uninoculated cast irons of this series. In order to test the accuracy of location of the shaded bands of Figs. 1, 2 and 3, the values for tensile strength and carbon equivalent, contained in numerous publications together with other test bar results, were found to fall within the areas included in Figs. 1, 2 and 3.

Inasmuch as the data utilized in this paper cover a wide range of commercial melting techniques and foundry practices, it is reasonably safe to assume that the relationships indicated apply to most unalloyed gray cast irons being produced at the present time within the composition range indicated. It is possible that lower minimum values are obtainable under some conditions, but such occasions should be rare.

Although a value for phosphorus may be included as one of the variables of the carbon equivalent formula, on the basis of the theoretical effect of phosphorus on the eutectic composition of cast iron, the evidence in Figs. 1, 2 and 3 would indicate that using phosphorus in the formula does not improve the compatibility of the carbon equivalent-tensile strength relation.

The shaded zones shown in Figs. 1 and 2 (without the phosphorus) are narrower than the comparable zones of Fig. 3 (with the phosphorus). The only "high" phosphorus irons in the series studied are among those above 4.1 per cent in carbon equivalent.

As previously mentioned, the information presented graphically in Figs. 1, 2 and 3 was obtained from production heats of gray cast iron made under the widest possible range of cupola operation and foundry technique. Furthermore, all testing and chemical analysis was done using routine methods and in plant control laboratories.

A foundry, therefore, utilizing consistent melting, testing, and analyzing procedures may be expected to produce irons which will fall in a narrower band than those shown in Figs. 1, 2 and 3. For example, less scatter would probably have been obtained if standard ASTM methods or extremely accurate research methods of chemical analysis were employed instead of the routine check methods actually used.

Permissible Analytical Error

According to the ASTM standards for chemical analysis, the permissible error in analysis would account for a spread of 0.10 per cent in carbon equivalent. Inasmuch as highly accurate technique is not required in routine check methods, not all the analytical data utilized in obtaining Figs. 1, 2 and 3 can be assumed to be within the accuracy of 0.10 per cent carbon equivalent.

Hence, while one cannot question but that some of the spread in tensile strength shown in the graphs may be accounted for by variations

Table 1 (Continued)

Carbon Equivalents of 82 Gray Cast Irons

Tensile Strength, psi.	Ratio of Tensile Strength to Brinell Hardness	Total Carbon + ⅓ Si	Total Carbon + .3 Si	Total Carbon + ⅓ (Si+P)	Total Carbon + .3 (Si+P)	4.3+C −(4.3-.33 Si) · (1-.1 P)
36,500	175	3.79	3.73	3.82	3.72	3.83
36,500	162	3.72	3.65	3.75	3.68	3.75
36,500[2]	176	4.20	4.12	4.27	4.18	4.26
36,000	165	3.97	3.89	4.06	3.97	4.03
36,000[1]	173	4.02	3.94	4.10	4.01	4.10
35,500	172	3.88	3.81	3.91	3.83	3.91
35,000[2]	172	4.20	4.11	4.27	4.17	4.26
34,500[2]	165	4.28	4.19	4.37	4.27	4.36
34,500	167	4.11	4.03	4.18	4.09	4.18
33,500	170	4.16	4.08	4.23	4.14	4.22
33,500	172	4.07	4.00	4.14	4.06	4.13
33,000	165	4.18	4.10	4.26	4.17	4.25
33,000	171	4.09	4.01	4.17	4.08	4.15
32,500	170	4.12	4.04	4.19	4.10	4.19
32,000	165	4.17	4.09	4.25	4.15	4.24
32,000[1]	170	4.26	4.17	4.34	4.23	4.33
32,000	169	4.09	4.02	4.16	4.08	4.16
32,000[2]	167	4.22	4.14	4.29	4.20	4.32
32,000	165	4.10	4.02	4.17	4.08	4.16
30,500[2]	163	4.22	4.13	4.30	4.15	4.31
30,000	160	4.18	4.10	4.25	4.16	4.24
29,500	163	4.20	4.12	4.29	4.20	4.28
29,500	161	4.18	4.10	4.27	4.18	4.26
29,500[1]	167	4.21	4.13	4.28	4.19	4.26
29,500	160	4.14	4.06	4.22	4.12	4.21
28,500	155	4.27	4.19	4.36	4.27	4.35
27,500	153	4.14	4.05	4.23	4.13	4.22
27,000[1]	159	4.32	4.23	4.39	4.29	4.37
26,000[1]	160	4.29	4.20	4.38	4.28	4.37
25,500[1]	158	4.23	4.15	4.30	4.21	4.30
25,000[1]	151	4.33	4.24	4.42	4.32	4.40
25,000[1]	153	4.29	4.20	4.36	4.26	4.36
24,500[1]	151	4.29	4.20	4.36	4.26	4.36
23,500[2]	143	4.32	4.24	4.52	4.42	4.52
23,000	135	4.39	4.31	4.46	4.37	4.45
23,000[1]	140	4.44	4.35	4.64	4.53	4.60
22,500	133	4.29	4.21	4.49	4.39	4.48
22,000[1]	137	4.29	4.20	4.36	4.26	4.36
21,500[1]	130	4.39	4.30	4.59	4.48	4.58
19,500	134	4.43	4.35	4.50	4.40	4.48
18,000	120	4.45	4.36	4.51	4.41	4.49
17,500	127	4.46	4.38	4.52	4.43	4.51

[1]Treated with graphitizing inoculant.
[2]Treated with stabilizing inoculant.

Table 2

RELATION OF STRUCTURE TO TENSILE STRENGTH-BRINELL HARDNESS RATIO*

Tensile Strength Divided by Brinell Hardness	*Structure*
210 and up	Normal graphite, smallest cell
190 to 210	Normal graphite, small cell
180 to 190	Normal or low percentage "Type D" graphite, medium cell size
160 to 180	"Type D" graphite, variable cell size
Below 160	Completely "Type D" graphite, large cells, or free ferrite

*Written discussion by T. E. Barlow and R. G. McElwee, A.F.A. TRANSACTIONS, vol. 50, p. 1099 (1942).

in structure of the metal which may also profoundly affect the tensile properties, still part of the spread could arise from inaccuracies in the analytical and physical data.

Since the relationship between tensile strength and carbon equivalent takes the form of a band or zone, rather than a single line, relative position within the zone may be a direct measure of relative quality. For example, irons defined by the upper limit of the shaded area in Fig. 1 may be considered to be "premium" irons.

The structure of such "premium" irons would indicate maximum resistance to wear, maximum transverse properties (including deflection), maximum machinability for the strength obtained, and maximum impact properties for a given carbon equivalent.

On the other hand, gray cast irons described by the lower line of the tensile strength-carbon equivalent band may be expected to exhibit excessive chilling tendencies, lower impact properties, less desirable machining characteristics, and lower resistance to galling and scratching for a given carbon equivalent.

Tensile Strength to Brinell Hardness Ratio. A common method of specifying the physical and mechanical properties of gray cast iron is by Brinell hardness number; the assumption being that the tensile strength is roughly proportional to Brinell hardness and can be calculated by simply multiplying the Brinell hardness by an empirical factor. In most cases this factor is 200, although there are instances where factors of lower values are being utilized.

The ratio of tensile strength to Brinell hardness for gray cast iron, unfortunately, is not a constant but is affected by variables associated with microstructure, etc., that generally reflect more upon one property than upon the other. The ratios for the irons studied in this series are included in Table 1. They vary from as low as 120 to 1 to as high as 234 to 1.

Apparently, the value for the ratio varies with chemical composition, becoming smaller as the carbon equivalent increases. In Fig. 4, the relation between the ratio and carbon equivalent is presented graphically, using as the carbon equivalent, carbon plus 1/3 silicon (C + 1/3 Si).

Figure 5 contains the same information, but expressing the carbon equivalent as carbon plus 1/3 silicon plus 1/3 phosphorus (C + 1/3 Si + 1/3 P). Since the use of a value for phosphorus in the carbon equivalent formula appears to increase the scatter of results rather than to narrow it, the simple expression C + 1/3 Si, shown in Fig. 4, is used subsequently in this paper.

The same technique used in plotting the curves of Figs. 1, 2 and 3 is employed in Figs. 4 and 5 for designating treatments of the irons. In these figures, however, the lower boundary of the zone for inoculated irons is the same as for the uninoculated irons. Therefore, the center line defines the maximum for the uninoculated irons rather than the minimum values for the inoculated irons.

Although the graphs of Fig. 4 indicate that the tensile strength-Brinell hardness ratio is a function of composition (C + 1/3 Si), a wide range of ratio values exist for the same given carbon equivalent. This is particularly true in the lower carbon equivalent or "hard iron" range. The spread in ratio values for the lower carbon equivalent irons is significant as an indication of structure and physical properties other than tensile strength and Brinell hardness.

The relationship between the tensile strength, Brinell hardness, and

Table 3

RELATIONSHIP BETWEEN STRUCTURE AND TENSILE STRENGTH TO BRINELL HARDNESS RATIO—LISTED BY COMPOSITION RANGE

Carbon Equivalent, per cent	*Tensile Strength Divided by Brinell Hardness*	*Structure*
3.45-3.65	210 and over	Smallest cell, normal graphite
	190-210	Small cell, normal graphite
	180-190	Medium cell, some "Type D"* graphite
	170-180	Large cell, some "Type D"—medium cell, completely "Type D"
	160-170	Large cell, partial "Type D"
	160 and below	Large cell, complete "Type D"
3.65-3.85	210 and over**	Smallest cell, normal graphite
	190-210	Smallest cell, normal graphite
	180-190	Medium cell, normal graphite or small cell, partial "Type D"
	170-180	Large to medium cell with partial "Type D" graphite
	160-170	Large cell, "Type D" graphite or free ferrite
3.85-4.20	190-210	Medium cell, normal graphite
	180-190	Medium cell, large normal graphite
	170-180	Medium or large cell, some "Type D"
	160-170	Large cell, "Type D" graphite
	160 or below	Free ferrite, "Type D" graphite

*"Type D" graphite—A.F.A.-A.S.T.M. graphite flake size chart—also called modified, eutectiform, dendritic, pseudo-eutectic, etc. Undesirable for wear resistance; low deflection, low toughness values, and poor transverse properties.

**Very few irons in this range.

the structure of gray cast iron was described by Adams[1] in 1942. Adams presented information on the cell size, graphite size and distribution, and composition of irons of various physical and mechanical properties. He attempted to classify these irons on the basis of their cell size, graphite size, graphite type, and graphite distribution.

A written discussion, presented by R. G. McElwee and T. E. Barlow,[1] attempted to correlate the same data on the basis of the tensile strength-Brinell hardness ratio. The discussion included the information shown in Table 2.

This table may be modified by the inclusion of the effect of chemical composition as indicated in Fig. 4. On this basis, the relationship between the tensile strength-Brinell hardness ratio and the microstructure is tabulated in Table 3.

Structure Type

The relationship indicated in Fig. 4 confirms the experimental information presented by Adams[1] in spite of the fact that the approach to the problem was different. Furthermore, an interesting agreement exists between the shaded areas of Fig. 4 and the change in the type of structure as indicated in Table 3.

For example, the center line of Fig. 4 passes through the points listed in Table 3 as the limiting ratios for completely normal graphite distribution. Thus, in the lower carbon equivalent irons (3.45-3.65 per cent), those irons having a ratio of 190 or greater show normal graphite distribution with variation only in cell size and graphite size (the higher the ratio the smaller the cell size and the more refined the graphite flake).

Of those irons having a ratio of less than 190, all contained some "Type D" or modified graphite in their microstructures. Therefore, the value for the tensile strength-Brinell hardness ratio within a given range of carbon equivalent is a function of cell size, graphite size, and the percentages of "Type D" graphite structures present in the test bar; the lower values indicate the larger cell size and the greater percentage of the undesirable "Type D" graphite structures present.

Within the composition range 3.65 to 3.80 per cent carbon equivalent, relatively few irons have a tensile strength-Brinell hardness ratio of 210 or greater. Most of the irons containing normal graphite structures and extremely small cell size have a ratio in the range of 190 to 210. Therefore, in this composition range, the maximum ratio obtained is limited by composition as well as structural considerations.

Of the higher carbon equivalent irons studied, the difference between maximum and minimum ratios obtained are a function of structure. Although the variations are less pronounced in this range, cell size is restricted by composition and cooling rate, and the amount of "Type D" graphite is reflected in lower ratio values.

However, ratio values as low as 120 in high carbon equivalent irons do not necessarily indicate that they contain "Type D" graphite, since in this case the high carbon and silicon content promotes the formation of free ferrite in the 1.2-in. test bar. The presence of ferrite accounts for the low ratios obtained in the irons of high carbon equivalent values, the effect of graphite distribution and cell size merely controlling the relative position within the shaded area of Fig. 4.

Conclusions

Since both published and unpublished information confirms the relationships indicated in this paper, the following observations can be made:

1. In the absence of alloying elements, the tensile strength of gray cast iron in a 1.2-in. arbitration bar is a function of composition expressed as total carbon plus 1/3 silicon or the carbon equivalent.

2. Graphical presentation of the relationship between tensile strength and carbon equivalent may be utilized to advantage to indicate the relative quality of gray cast iron and to provide the engineer with an indication of machinability, wear resistance, and other essential engineering characteristics.

3. Dividing the value for tensile strength by the Brinell hardness number gives a value which is a function of the carbon equivalent (carbon plus 1/3 silicon).

4. The ratio of tensile strength to Brinell hardness for any given chemical composition is also a function of the structure of cast iron as indicated by the cell size, the graphite structure, distribution of the graphite, and the character of the matrix.

5. If the value for the tensile strength divided by Brinell hardness number is known for an iron of any given carbon equivalent, the relative quality of that iron is indicated by its position on a curve relating the ratio and the carbon equivalent.

6. High values for the tensile strength-Brinell hardness ratio indicate small cell size, refined graphite particles, normal graphite distribution, and relative freedom from ferrite. Conversely, low ratios with a given carbon equivalent indicate large cell size, large graphite flakes, modified or "Type D" graphite structure, or the presence of ferrite resulting either from modification or structure or from composition. Such latter irons can be expected to provide mediocre wear resistance, low deflection values, low impact strength, and low transverse properties.

7. The Brinell hardness number of gray cast iron is an indication of machinability only when the carbon equivalent is known, since a low tensile strength-Brinell hardness ratio may be the limiting factor.

8. In cases where both the tensile strength and the Brinell hardness number must be specified, their ratio should be compatible with respect to chemical composition and the type of structure obtainable.

Acknowledgment

The authors wish to acknowledge the cooperation of the Gray Iron Research Institute, Inc., which has so generously permitted the use of information obtained from foundries within its organization.

Reference

[1] Robert R. Adams, "Cast Iron Strength vs. Structure," Transactions, American Foundrymen's Association, vol. 50, pp. 1063-1103 (1942).

DISCUSSION

Chairman: R. G. McElwee, Vanadium Corp. of America, Detroit.

Co-Chairman: W. E. Mahin, Armour Research Foundation, Illinois Institute of Technology, Chicago.

R. Schneidewind[1]: I would like to ask Dr. Lorig whether, when the relationship $\frac{\text{Tensile Strength}}{\text{Brinell Hardness}}$ is less than 190, he has not also found considerable ferrite in the structure.

Dr. Lorig: The amount of ferrite which forms, particularly when carbon equivalent values become high is an im-

portant factor as far as the hardness and strength are concerned.

A. J. DUBLO[2]: Was an effort made to control the pouring temperature? Were the bars selected?

DR. LORIG: No attempt was made to control the casting conditions. These bars were randomly chosen from production melts. They were chosen from a group of bars made available to us and we know little about the actual conditions under which they were cast.

MR. DUBLO: Would it not be possible that the chilling effect would interfere with the composition?

DR. LORIG: Yes indeed. Then, of course, there are the additional factors of the variations in foundry practices and variations that come through chemical analysis including such elements as manganese and phosphorus and residuals.

DR. R. A. FLINN[3]: We were interested in Dr. Lorig's statistical study of these different heats. Some people have begun to think of carbon equivalent as a fundamental property. Dr. Lorig indicated that it is an empirical factor. We wonder if any approach has been made by the excellent statistical staff at Battelle Institute to correlate these properties more fundamentally?

Tensile strength is a function of the state of the carbon and the state of the matrix. We have been able to approach a quantitative evaluation of structure by this method that is more fundamental.

Complete analyses of the specimens would provide a valuable addition to this paper. There were two Class 50 irons shown on the curve. It would be interesting to include some Class 60 and 70 irons also.

DR. LORIG: In answer to the first part of the question, we have done nothing in regard to further correlating properties with structural characteristics of the iron. I think it is a very interesting approach and I am glad that you are doing it.

In regard to including in the tests some values of Classes 60 and 70 irons, I don't know whether that would be possible. In order to obtain such tensile values it would be necessary to alloy. We have purposely omitted the alloyed irons from the analysis.

DR. SCHNEIDEWIND: The so-called carbon equivalent is not necessarily an arbitrary or empirical expression. It is based on the composition of iron-carbon-silicon eutectics. The expression $C+\frac{1}{3}$ Si $=4.3$ shows the relationship between C and Si at eutectic compositions and is accurate up to about 6 per cent silicon.

Alloys naturally modify the expression. $C+\frac{1}{3}$ Si is therefore the carbon equivalent and is a measure of the amount of graphite one normally would expect in a gray iron which does not contain massive cementite.

H. C. AMTSBERG[4]: I would like to ask Dr. Lorig if his field covers all gray iron?

DR. LORIG: They are all gray irons.

CO-CHAIRMAN MAHIN: Were these bars all the same in as-cast size?

DR. LORIG: Yes.

CO-CHAIRMAN MAHIN: So that matter of chill that Mr. Dublo asked about would at least be a constant factor.

C. H. JUNGE[5]: I was wondering if the following approach to the strength of gray iron would lead anywhere. In the 20th Campbell Memorial Lecture, ASM, Maxwell Gensamer indicated that the strength of iron-carbon alloys can be predicted by the dispersion of the carbide phase. The logarithm of the mean length of ferrite path between carbide particles is found to be linearly related to the stress at 0.2 per cent effective strain, regardless of the shape of the carbide particles in the aggregate.

Could the length of the path between graphite particles in grey iron be related to the strength properties in a similar manner? Of course, we have the additional variable of possible changes in the strength of the matrix itself, but would the approach have merit?

DR. LORIG: That angle has not been considered in this work, but in examinations of malleable iron we did notice some relationship between the distance of the various particles.

CHAIRMAN MCELWEE: I would like to call Dr. Flinn's attention to the fact that all of the changes in physical properties can not be attributed to changes in structure.

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[1]University of Michigan, Ann Arbor, Mich.
[2]Sterling Foundry Co., Wellington, Ohio.
[3]American Brake Shoe Co., Mahwah, N. J.
[4]Westinghouse Electric Corp., E. Pittsburgh, Penna.
[5]Clevealnd Graphite Bronze Co., Cleveland

STEEL FOUNDRY CONTROL

This paper deals with control in steel melting, sand processing and testing, heat treatment and inspection. Also discussed are determination of heat analysis, slag control, temperature control, fluidity testing, deoxidation control, inclusions in steel, metallographic examination and physical testing of steel.

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CONTROL in the steel foundry for practical purposes can be broken down into four classifications. They are:

1. Control in Melting
2. Control in Sand Processing
3. Control in Heat Treatment
4. Control in Inspection

Melting Control

Need for control in any process is universally recognized, for only by maintaining control can a process be run with a minimum of time and expenditure. The task of producing castings of uniform quality and properties requires the use of effective control at the origin of the casting, the melting process.

Control of the melting process should begin with the inspection of incoming raw materials. This means analyzing all materials used in the heat, such as scrap, pig iron, fuel, lime and ferro-alloys and immediately rejecting any unsuitable for use.

The mechanical features of all types of furnaces have also been greatly improved over the years to aid the operator in closer control of the melting operation. Modern furnaces are equipped with the necessary controls for draft, air supply, fuel and reversal. A modern open hearth instrument panel is shown in Fig. 1.

In the actual melting process the primary objective of the melter is to finish the heat with the desired chemical analysis at the correct temperature for pouring. The melter has a number of methods to aid him in controlling the chemical analysis.

Determining Carbon Content. An experienced melter can estimate the carbon content of a steel bath within a few points by examining the fracture of a sample block which has been broken in half. However, several methods less subject to the human element have been developed for fast and accurate carbon determinations.

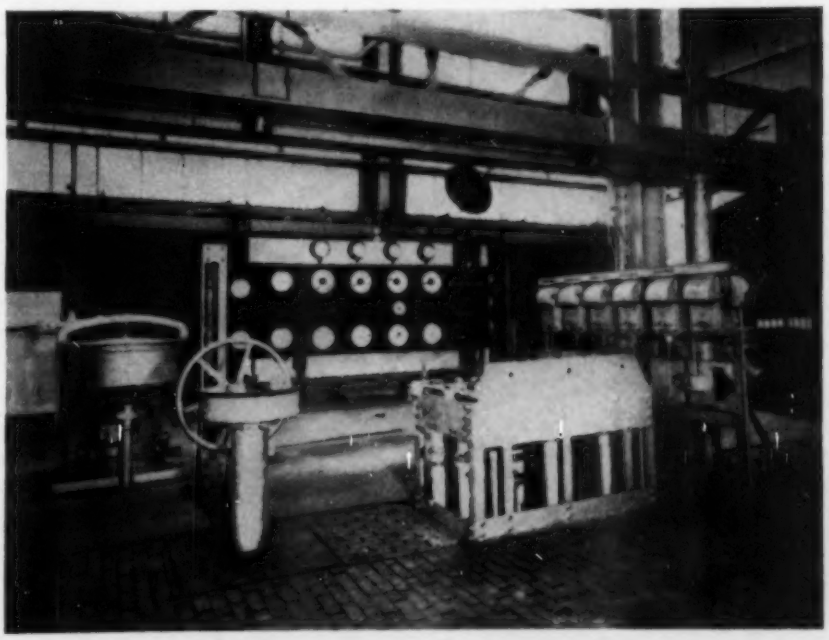
One of these is the carbanalyzer which is based on the fact that magnetic permeability of the steel is directly related to the carbon content. This instrument will give readings in two minutes. One disadvantage of it is that it is not sufficiently accurate in the presence of alloys, unless it has been calibrated for that particular range of alloy content.

When alloys are present, the most reliable method is the combustion method wherein the carbon in the sample is burned to carbon dioxide and then the amount of carbon dioxide formed is measured, either gravimetrically or volumetrically, to give the carbon content.

Apparatus for the volumetric method is shown in Fig. 2. The volumetric analysis gives an accurate determination in two minutes which is considerably faster than the gravimetric process.

Another method of carbon analy-

Fig. 1—Instrument panel for open hearth furnace control.



sis is based on the fact that the as-quenched hardness of a steel is dependent on the carbon content and thus by taking a hardness reading on a quenched sample the carbon content is obtained. This method of determining the carbon content is not satisfactory when the total alloy content is too high. All these methods are described in detail in the literature.

Manganese Content. An analysis for manganese must also be made before tapping so that additions may be made in finish in the required range. Manganese does not vary so rapidly as carbon so that the time element in analysis is not as important as in carbon.

Silicon Content. The common silicon analyses are not rapid enough for use on the melting floor and are not usually needed except on the acid electric furnace where silicon can be reduced back into the bath from the slag in the furnace and thus cause troublesome variations in the final silicon content.

However, a test has been devised which will give a rapid estimate of the silicon content. It consists of pouring a sample of the steel into a small, cup-shaped, dry sand core. The height to which the sample rises is a measure of the silicon content. The more silicon that is present, the less carbon monoxide gas is evolved and the flatter is the crown of the sample. A series of these tests is shown in Fig. 3. After a definite procedure has been developed, this test is accurate within at least ± eight points of silicon.

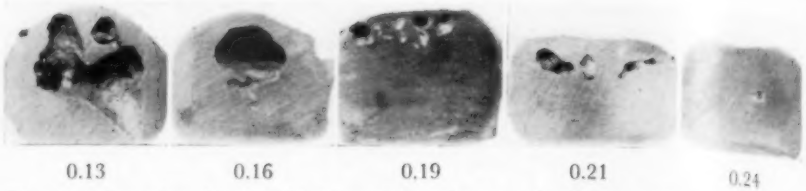


Fig. 3—Samples from which rapid estimate of silicon was made during process of acid electric furnace heat.

Phosphorus and Sulphur. Phosphorus and sulphur are usually checked during the refining period, particularly in the open hearth.

Spectographic Analysis. Another means of analysis which is gradually being more widely adopted is the spectograph. This machine gives a rapid analysis of all elements except carbon, phosphorus and sulphur from one sample. It is particularly valuable when it is necessary to know the analysis of all the residual elements, such as in melting to hardenability specifications.

Slag Control

The most important single factor in steelmaking is slag control. In fact, it has been aptly stated that if the slag is properly made and controlled, the steel will take care of itself. The whole aim of slag control is to determine the most desirable slag conditions for a particular shop and then to manipulate the slag so that these conditions may be consistently obtained.

In the basic open hearth, this generally resolves itself into a satisfactory removal of phosphorus and sulphur and a desired degree of oxidation at end of refining period.

Several methods of rapidly estimating slag compositions have been developed. One of the most popular of these is the slag pancake test in which a sample of molten slag is poured into a flat round mold to produce, actually, a pancake of slag.

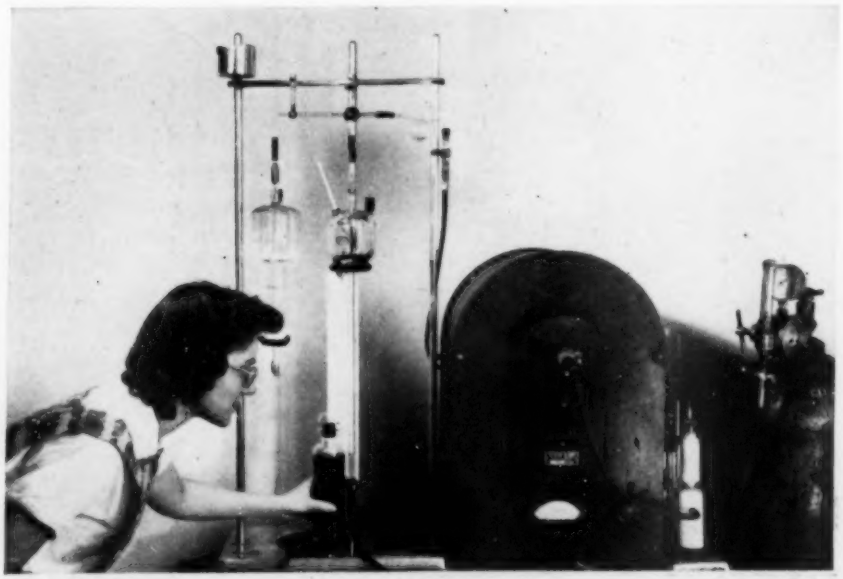
An experienced operator can estimate the basicity of the slag rather closely by an examination based on the presence of creases or furrows on the upper surface, the "shine" of the lower surface, and the color of the slag. The iron oxide content is estimated from the fracture of low-iron slags and from films appearing on the upper surface as iron content and basicity increase.

Any detailed plan of slag control is necessarily valuable only to the plant for which it was developed. In general, a lime-silica ratio of about 2.7 to 3.0 is necessary for efficient phosphorus and sulphur removal.

Another method of determining the lime-silica ratio is the slag viscosity test. This is used principally in acid open-hearth practice and is based on the fact that the fluidity of the slag is proportional to its base-acid ratio, the more acid the slag the greater distance slag will run.

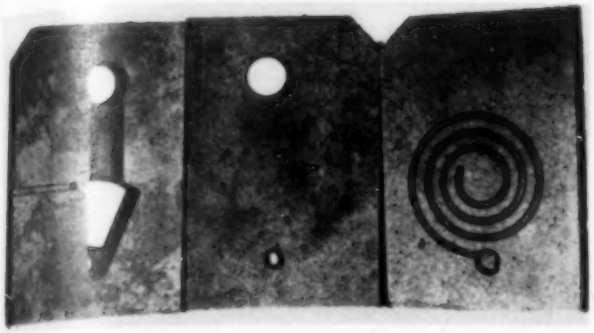
In acid melting, as we know, there is no phosphorus or sulphur removal so that attention is generally directed toward obtaining the desired slag volume and oxidation conditions.

Fig. 2—Volumetric analysis for carbon content.



Temperature Control

An important factor which the melter must have under control is temperature. The importance of temperature is self-evident. Aside from its effect on furnace slag-metal reactions, every ladle skull or misrun in a casting due to low metal temperature raises the cost of foundry operation. On the other hand, when the metal is too hot, there is a tend-



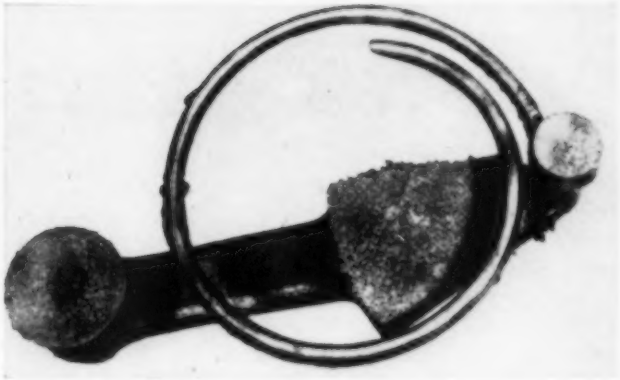


Fig. 4 (left)—Molten metal fluidity test mold. Fig. 5 (right)—Fluidity test casting from a medium plain carbon steel.

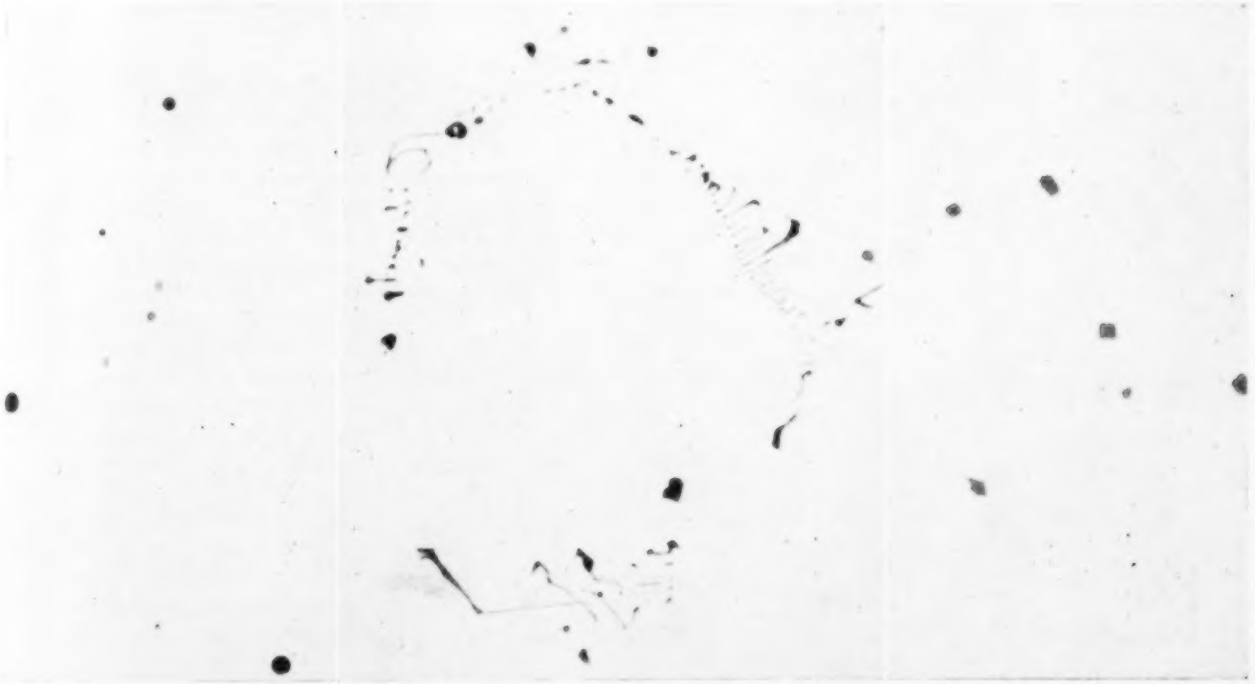


Fig. 7 (left), Fig. 8 (center) and Fig. 9 (right)—Photomicrographs showing three types of inclusions found in cast steel.

Fig. 10—Exograph of unsatisfactory pouring nozzle.

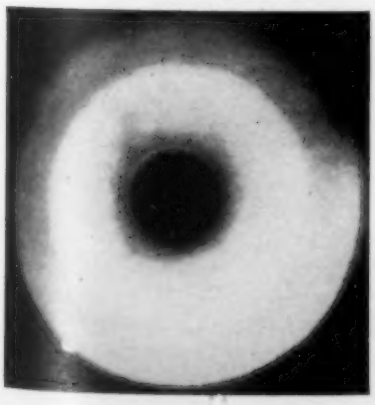
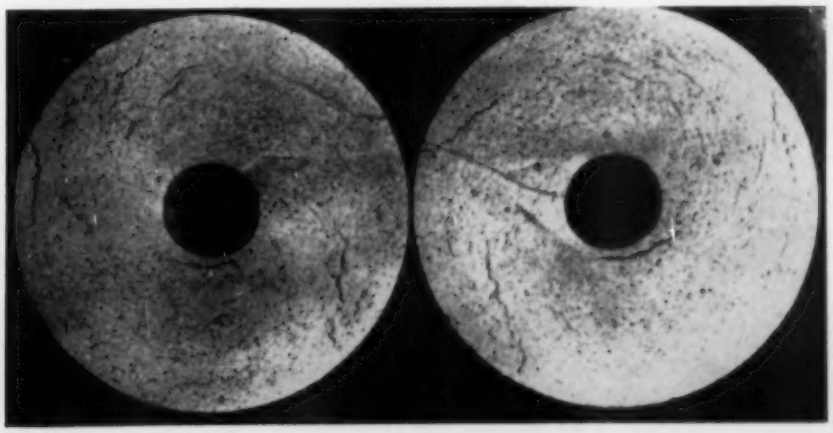


Fig 11—Photo of sectioned pouring nozzle shown in Fig. 10.



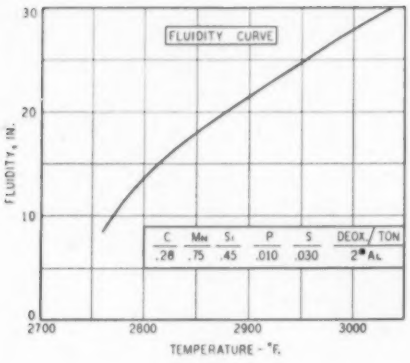


Fig. 6—Fluidity curve for steel of the analysis indicated above.

Table 1

EFFECT OF WET RECLAMATION ON SHAKEOUT SANDS FROM TWO FOUNDRIES HAVING DIFFERENT SAND PRACTICES. RECLAMATION OF SLUDGE PRODUCT FROM WASHING SYSTEM FOR FINE SAND APPLICATION

U. S. Bureau of Standard Sieve No.	Per Cent Retained — Foundry "A" Before Washing	Foundry "A" After Washing	Foundry "B" Before Washing	Foundry "B" After Washing	Fine Sand Reclaimed from Sludge
6	—	—	—	—	—
12	.11	.03	.26	.06	—
20	.34	.24	.26	.21	.05
30	3.16	3.28	2.84	2.95	.34
40	28.28	31.75	26.94	30.35	3.30
50	31.00	36.16	31.40	35.67	4.27
70	16.11	16.80	14.28	16.22	7.57
100	8.11	8.17	7.67	8.09	29.74
140	2.80	2.13	3.95	3.30	33.94
200	1.26	.46	1.41	.92	12.31
270	.84	.20	.76	.31	4.11
Pan	1.94	.26	2.44	.60	2.12
AFA Clay	5.13	.30	7.53	1.08	2.20
AFA Fineness No.	50.2	43.0	52.5	45.4	95.3
Per Cent Passing No. 70 Sieve	14.96	11.24	16.23	13.28	82.22

ency to obtain a burned-in sand condition. Naturally, the best results are obtained from the middle of the temperature range.

Spiral Fluidity Test

Several methods are used to determine metal temperature. One of these tests is the spiral fluidity test which is based on the fact that the fluidity of steel is dependent mainly on its temperature. Thus, if a sample is poured into the spiral mold, the number of inches it travels along the spiral gives the temperature from a perviously calibrated chart.

Actually, in practice the fluidity test may be standardized directly with pouring conditions so that a definite minimum fluidity value in the furnace may be established before the heat should be tapped. Figures 4 and 5 show a fluidity mold and a fluidity casting from a medium plain carbon steel. Figure 6 shows the type of data obtained with the fluidity test.

Immersion Type Thermocouple

Recently, several direct reading methods of obtaining molten metal temperatures have been developed. One of these is an immersion type thermocouple composed of a platinum-platinum rhodium thermocouple encased in a fused quartz tube for protection from the bath and slag. The quartz protection tube usually must be replaced after each reading.

Rayotube. Another unit which is now in use is the Rayotube. The Rayotube is composed of a long open end tube which is inserted into the bath with a total radiation pyrometer at the one end to obtain the temperature reading. A stream of air is forced through the tube continuously to prevent the metal from entering the tube. Although this instrument gives only the apparent bath temperature, the process is consistent from heat to heat so that the readings may be correlated with foundry practice and control thus be maintained.

Deoxidation Control

Another factor which cannot be emphasized too strongly is the control of deoxidation. Since the use of aluminum as a deoxidizer is neces-

Fig. 12—Chemical analysis quality control chart.

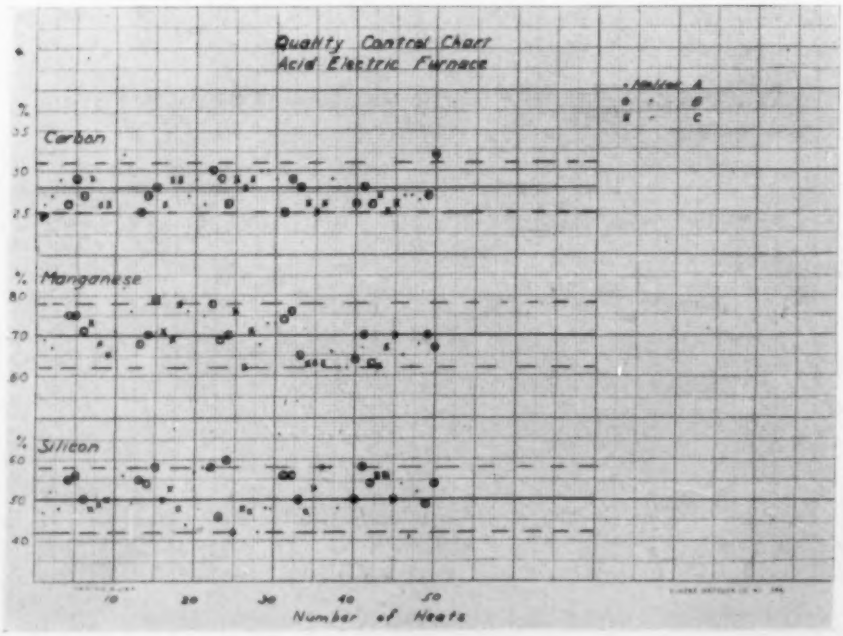


Table 2

SIEVE ANALYSES OF TWO SILICA FLOURS

U. S. Bureau of Standard Sieve No.	Silica Flour A	Silica Flour B
6	—	—
12	—	—
20	0.02	0.02
30	0.02	0.04
40	0.12	0.10
50	1.04	3.25
70	1.42	8.00
100	0.72	1.14
140	6.44	10.54
200	12.06	28.16
270	42.58	25.84
Pan	33.20	19.88
AFA Fineness No.	214.9	172.7

Table 3

PHYSICAL PROPERTIES OF A CORE SAND MADE WITH TWO SILICA FLOURS, A AND B

Composition of Sand Mix
1.5 %—Gelatinized Corn Flour
0.15%—Western Bentonite
12.5 %—Silica Flour
0.9 %—Linseed Oil
85.85%—AFA No. 40 Washed and Dried Sand

	Silica Flour A	*B*
Moisture, per cent	5.8	5.9
Green Permeability	140	135
Green Compression Strength, psi	1.6	1.5
Green Deformation	.044	.039
Dietert Flowability	85	85
Mold Hardness	24	24
Baked Permeability*	160	160
Baked Tensile Strength, psi.*	270	242
Core Hardness*	95	94
Compression Strength at 2500° F., psi.*		
1 Min. Heating Time Interval	50	48
2 Min. Heating Time Interval	87	77
4 Min. Heating Time Interval	54	24
8 Min. Heating Time Interval	138	75

*Specimens Baked at 450° F. for 3 hr.

sary in many foundries to eliminate porosity, the addition of aluminum must be closely controlled to secure the desired type of inclusions. As we know, the type of inclusions has a great influence on the physical properties.

Inclusions

Figures 7, 8 and 9 show three types of inclusions found in cast steel.

Type I (Fig. 7) inclusions are found in steels to which no aluminum additions have been made. These inclusions result in optimum ductility properties in the casting.

Figure 8 shows the undesirable Type II inclusions which are formed when aluminum has been added but not in amounts sufficient to form the desirable Type III inclusions. This type of inclusion causes a severe reduction in the ductility properties of the steel.

Type III inclusions are shown in Fig. 9. These inclusions are formed when sufficient residual aluminum is present in the steel. The ductility properties of steel with this type of inclusions are slightly lower than with Type I inclusions, but considerably higher than those of the Type II inclusions. When the aluminum additions are under control, the Type III inclusions are formed with resulting good ductility properties.

A rather unusual method of control has been employed to help correct trouble with nozzles while pouring. Considerable difficulty was encountered in pouring a medium manganese steel due to nozzles leaking or burning out. In an effort to trace the difficulties, the nozzles were radiographed and a sample exograph may be seen in Fig. 10. After sectioning, the nozzle appeared as shown in Fig. 11.

Naturally, these defects are very harmful with the type of thermal shock which a nozzle encounters. Thereafter, every nozzle was radiographed before use in medium manganese heats and the unsound nozzles rejected. The pouring troubles were eliminated immediately.

Quality Control Chart

A type of control which can be useful is quality control chart for chemical analysis. One type is illustrated in Fig. 12. The results from individual heats are plotted and each melter has a separate designation. The control chart aids in giving a visual picture of operations and efficiency in meeting the chemical standards. It should be emphasized that quality control does not correct bad practice, but it does immediately point out when a plant is out of control so that corrective measures may be taken.

These few factors which have been outlined will serve to demonstrate that effective means of control are available to aid us in our goal of a consistently high quality melted metal.

While steps may have been taken to produce a high grade metal for castings, the quality of the unfinished product may suffer if the sand which has been prepared for its casting is not subject to the same degree of control.

Sand Control

Steel castings present a more difficult sand problem than do metals of lower melting point as the occurrence of surface defects is much more likely at the higher steel pouring temperatures. For this reason, sand mixes for steel castings must be closely controlled if consistent results are to be obtained.

Experience in this country has shown that synthetic sands are the most satisfactory for steel castings

Table 4

ELEVATED TEMPERATURE PROPERTIES OF TWO CORE MIXTURES

	Mix No. 1 Cracked Castings	*Mix No. 2 Good Castings*
Cereal Binder	1.5%	2.5 %
Core Oil	1.0%	.25%
Crude Ottawa Sand	97.5%	97.25%
Moisture	4.2	4.1
Green Permeability	102	125
Green Compressive Strength, psi	2.0	3.1
Dietert Flowability, per cent	90	80
Mold Hardness	32	48
Baked Permeability	235	225
Core Hardness	93	84
Baked Tensile Strength, psi	202	135
Compression Strength at 2500° F., psi.:		
2 Min. Heating Time Interval	19.3	11.5
4 Min. Heating Time Interval	2.3	1.8
8 Min. Heating Time Interval	3.3	2.0
Breakdown Test (85 lb. Load)	26.6 sec.	13.2 sec.
Compression Strength at 2000° F., psi.:		
4 Min. Heating Time Interval	37.5	17.0
8 Min. Heating Time Interval	21.5	16.6
12 Min. Heating Time Interval	19.8	15.6

Note: Elevated Temperature Specimens Sprayed With Silica Wash and Baked for 1 hr. at 400° F.

because of their flexibility and the ease with which they may be closely controlled. The term "synthetic" as used here does not imply artificiality but a sand mixture which has been compounded or synthesized from suitable amounts of ingredients which have been successful.

The method of attack for a sand control program using synthetic sand is first to find a suitable base sand and determine what ingredients should be used to make the castings in question. Control of the sand mixes to be prepared is then effected by controlling the weights of sand, binders, and water added to batches of sand, raw materials having previously been checked to determine whether or not they conform to the required standards.

Sand Testing

Sand testing enters the scene at this point to measure the physical properties of the sand mix made and thus ascertain its suitability for the purpose intended. When sand mixes are first established, their physical properties are measured and used as a basis for setting up standards with which the physical properties of production mixes are compared.

At this point, it should be emphasized that sand testing will not be of maximum benefit to the foundryman if it is used only to reject sand whose physical properties are unsuitable. Maximum benefit can result only when results obtained are applied to the control of operations prior to taking the sand sample.

The particular property or properties of a sand mix which should be checked to determine its suitability for a given use is dependent on the casting being made. Following are instances in which various sand tests serve to detect sand mixes which are likely to cause casting defects.

Moisture Content. The moisture content of a sand mix has a very important effect on the physical properties and performance of the mix when used to make actual castings. High moisture content may result in blows, scabs, pinholes, or rough surface. Since moisture content has an important effect on the green strength and deformation of molding sand, it may be a contributing factor in instances in which defects result from improper green strength properties.

Sand Grain Size. The grain size and distribution of a silica sand have an important bearing on its performance when molten metal is poured against it. The foundryman has several means of checking the particle size of a sand. A commonly used means for determining particle size is the use of standard testing sieves. Table 1 shows the use of the sieve analysis to study the effect of a wet reclamation system on shakeout sands.

It will be noted that the sieve analyses of the washed sands are markedly different from those of the shakeout sands fed to the washing system. The steel foundryman knows that if he uses sand over and over that the amount of fines and clay in the sand will increase in amount and eventually result in an increase of such casting defects as scabs and erosion.

Fig. 13—Scab on surface of casting resulting from continued reuse of sand whose grain distribution and clay character went unattended.



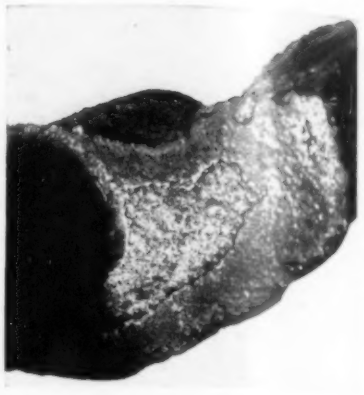


Fig. 14—Eroded in-gate caused by use of an underbaked core oil silica flour mix.

Figure 13 shows a scab such as may result from continued re-use of sand without attention being paid to grain distribution and clay character.

The use of the sieve analysis enables the foundryman to detect the exact amount of fines in his sand so that he may take the necessary corrective action. The sand washing system whose results are demonstrated by these sieve analyses is a comparatively recent development in sand control methods. Under suitable circumstances, its use has resulted in saving of money and the sand reclaimed is more uniform in clay content and grain fineness than new, crude silica sand. Our experience indicates that washed sand is suitable for use in either cores or molds and casting weighing as much as eight tons have been satisfactorily produced with the exclusive use of such sand.

Also shown in Table 1 is the sieve analysis of a fine sand reclaimed from the overflow of the classifier in the wet reclamation system. What was formerly a waste product has thus been converted to a fine sand which can be substituted for silica flour in many applications.

An instance occurred in one of our foundries which demonstrated the need for a more careful check of incoming raw materials. Trouble was experienced in the preparation of silica wash, an excessive amount of silica flour settling out in batches which were otherwise satisfactory. The cause of the settling was quick-

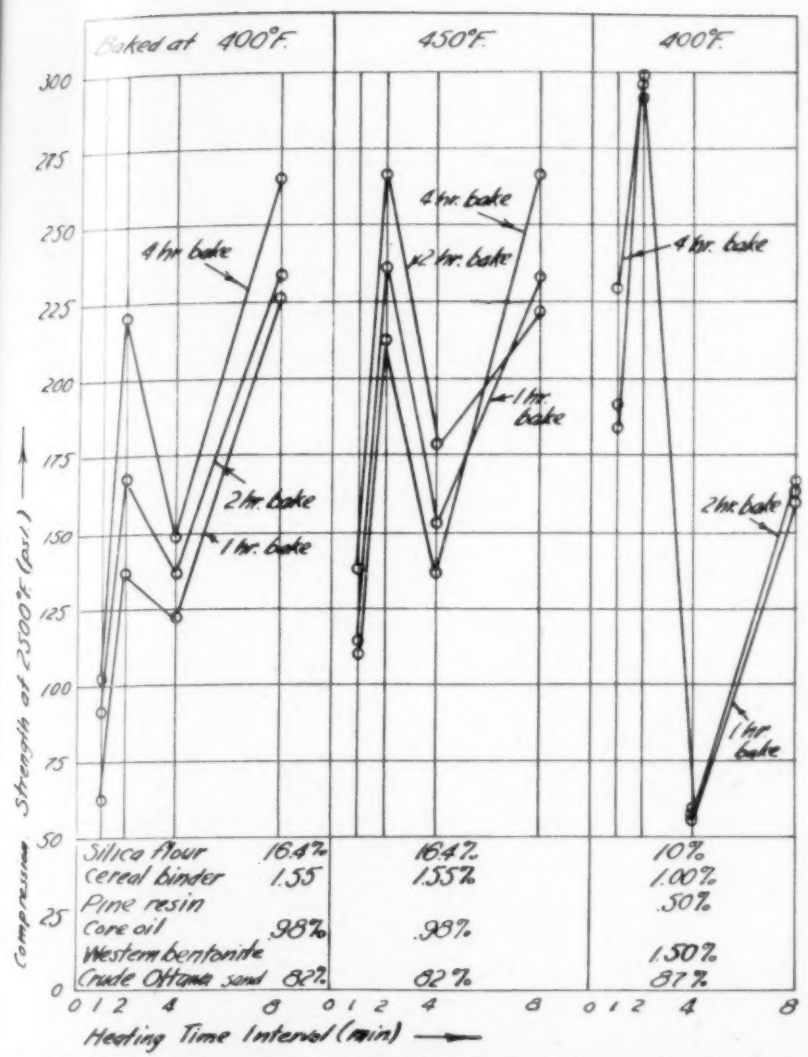


Fig. 15—Elevated temperature properties of a typical core oil silica flour mix, baked at 400 and 450° F. for periods indicated, and properties of new type mix baked at 400° F. for similar time intervals.

ly found by making a sieve analysis of the silica flour in question. Table 2 shows sieve analyses of two silica flours, one being the grade usually specified and the other being the unsatisfactory material.

It will be noted that the unsatisfactory material is markedly coarser than the satisfactory silica flour. Table 3 shows the room and elevated temperature properties of a core sand mixture made with each of the two silica flours.

It will be noted that the finer silica flour markedly increased elevated temperature compression strength at heating time intervals of 4 and 8 min.

Often castings are encountered which are very sensitive to hot tearing. In many cases, the tears may result from the failure of an internal core to collapse soon enough. The solution to such a problem involves the development of a core mixture with a faster rate of collapse. Table 4 shows the changes that occurred in elevated temperature properties when the collapsibility of a core mix was increased.

To be particularly noted is that the compression strength as measured at 8 min. heating time was similar in both instances, but the strengths at the short heating time intervals were quite different. The weaker of the two mixes as judged by elevated temperature properties was successful in producing the casting in question while the other mix cracked the castings. This instance is not cited for the primary purpose of showing what sand additions will increase collapsibility successfully, but to demonstrate the use of the sand laboratory in solving such problems.

If a similar problem presents itself and the elevated temperature properties of the mixture in use are known, laboratory experiments may then be made in which likely changes are tried. When these tests appear promising, foundry trials should follow. Thus, a suitable mixture may be found at a minimum expenditure of time and effort.

Elevated temperature compression tests can be quite usefully applied to foundry problems as just demonstrated. Another useful elevated

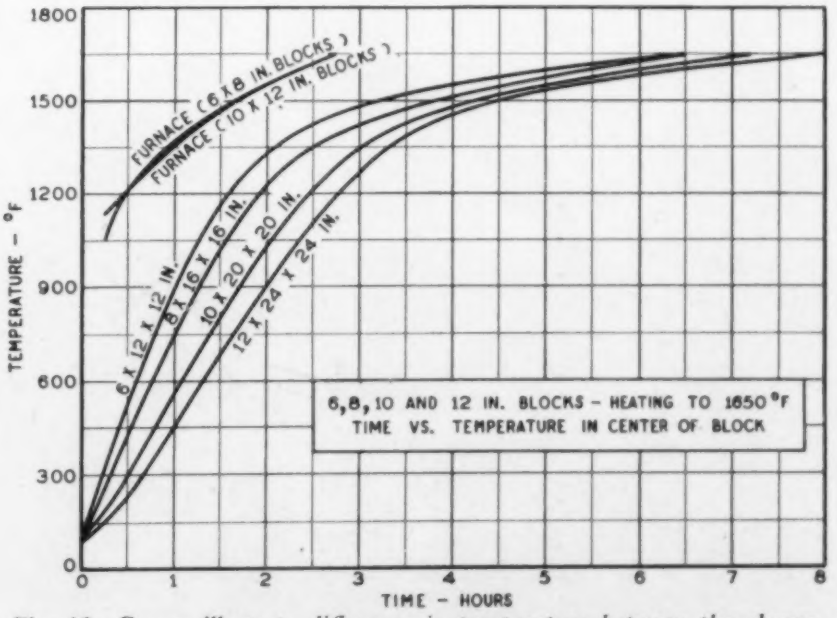


Fig. 16—Curves illustrate difference in temperature between the charge and that indicated by control thermocouples.

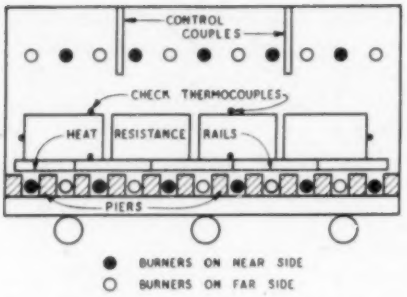


Fig. 17—Construction of batch type furnace car with brick piers.

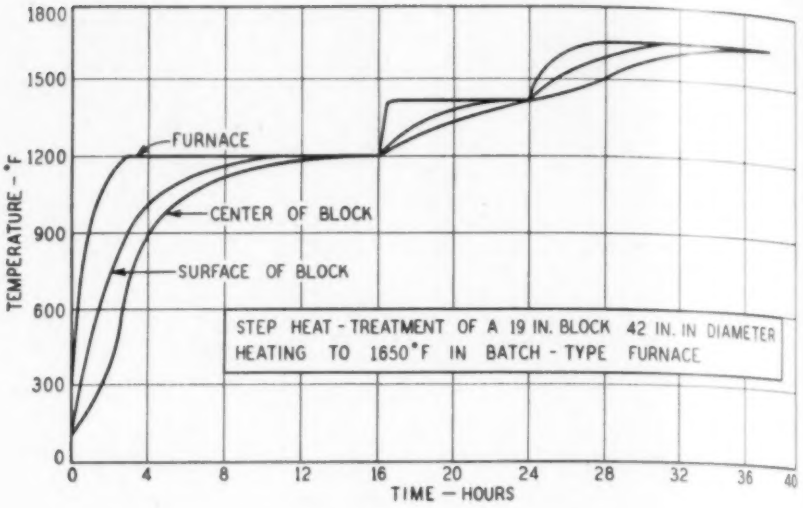


Fig. 18 (right)—Results of "check-run" on 19-in. section to establish "equalizing times" at various temperatures.

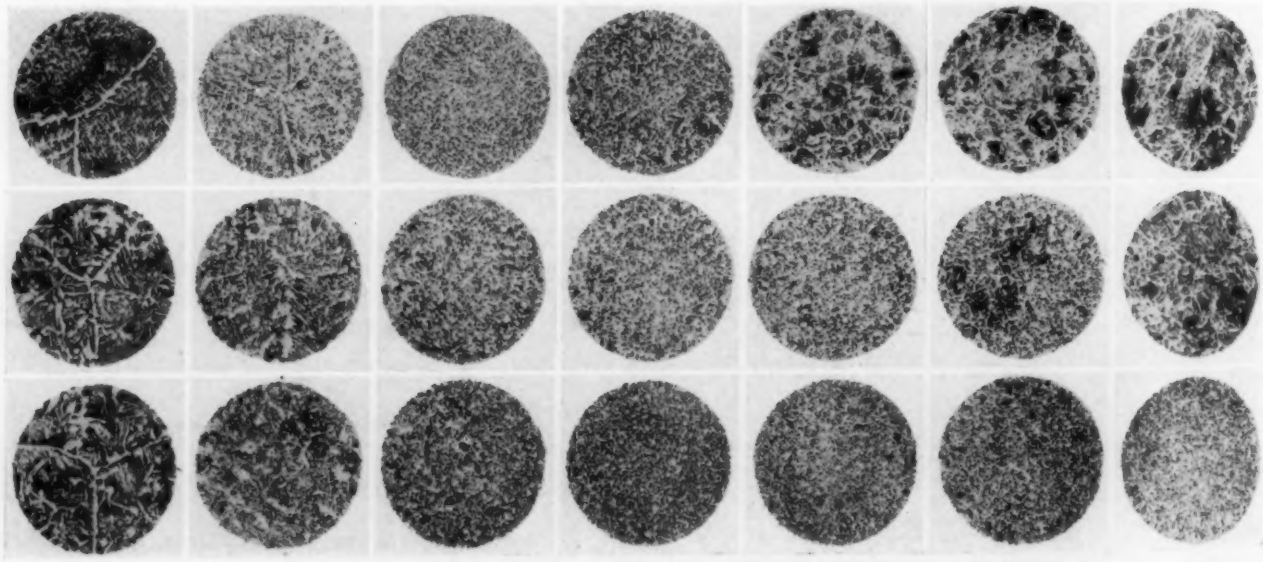
temperature sand test is the shock heating test in which a specimen of the sand in question is placed in a furnace which has previously been heated to some predetermined temperature such as 2500° F. which temperature we have found to be useful. After heating, the specimen is examined for cracking, spalling, or other surface changes which might have a bearing on the sand's performance in an actual mold.

This test is a measure of the thermal stability of a sand mix, one of the chief factors in determining the suitability of a sand. Work on elevated temperature tests and their correlation with casting results has demonstrated that a sand for a particular application must fulfill three fundamental requirements:

1. It must be thermally stable.
2. It must possess a certain minimum elevated temperature strength so as to withstand the erosive action of the molten metal which passes over it.
3. If it is in a position to restrict the contraction of the casting, its elevated temperature strength must not exceed a certain maximum value beyond which hot tears will be produced in the castings being made.

The necessity for thermal stability will be appreciated when it is realized that an inherently strong sand may easily be eroded if a crack develops and puffs out on its surface, thus allowing the molten metal sufficient purchase to start an eroding action.

Figure 14 shows a closeup view of an eroded in-gate, an underbaked core oil silica flour mix being used.



	C	Mn	Si	P	S	Al
Top Row	0.267	0.74	0.42	0.010	0.028	None
Center Row	0.251	0.74	0.38	0.010	0.031	¾ lb./ton
Bottom Row	0.262	0.73	0.37	0.008	0.032	2½ lb./ton

Fig. 19—Microstructures of underheated and overheated Grade "B" steel heated 15 min. at temperatures indicated followed by air cooling. The as-cast specimens were from keel-block shaken out 10 min. after pouring. Heat treated specimens were 1-in. cubes. Etch—4 per cent Nital. Mag.—Appr. 100x.

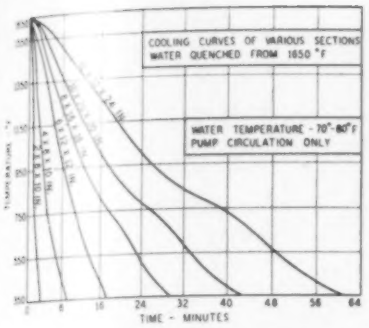
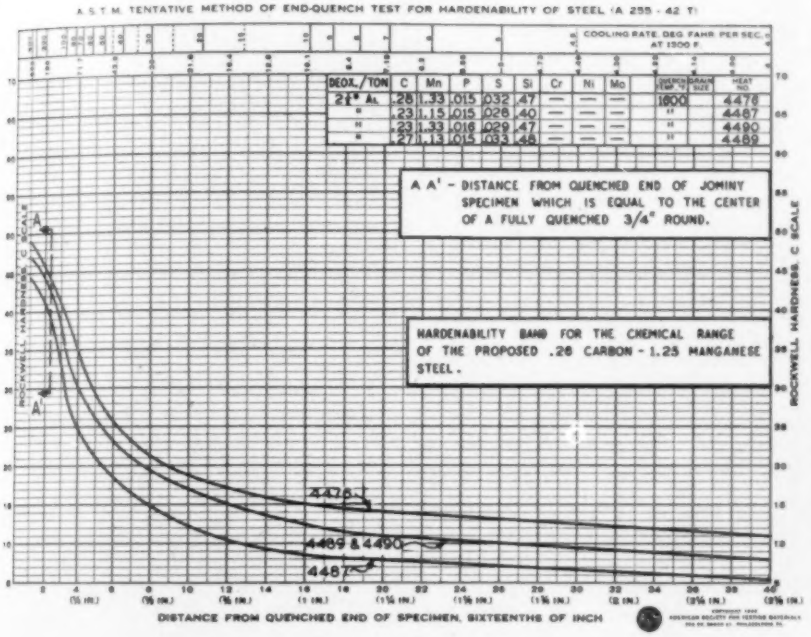


Fig 20—Cooling curves of various steel sections water quenched from 1650° F.



Fig. 23—Roughing floor inspection where gates, risers and other excess metal is removed by flogging or burning.



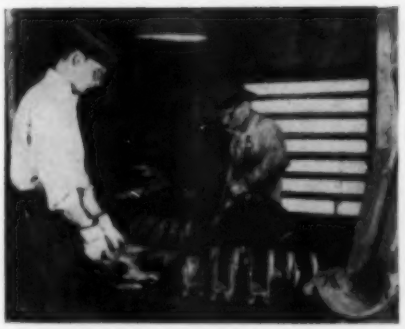


Fig. 24—Rough inspection where minor defects such as cracks, tears, dirt and design inconsistencies are uncovered after burning, shot blasting and chipping operations.

Fig. 21 (above) and Fig. 22 (below)—End-quench steel hardenability curves.

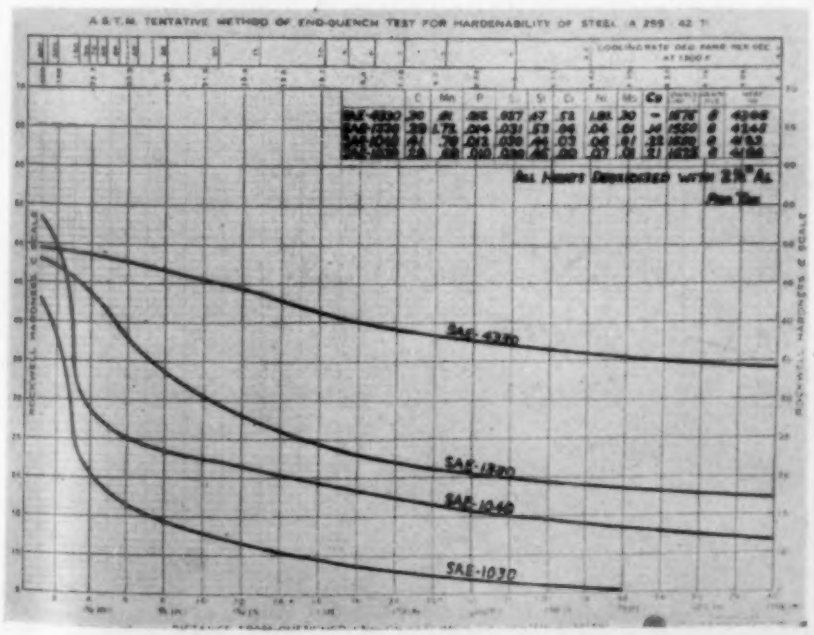


Fig. 25—Final inspection where welds, gauging operations, hardness reading and general appearance are noted.



This type of mix is not thermally stable and which can cause defects such as that just shown.

Figure 15 shows the elevated temperature properties of a typical core oil silica flour mix, baked at 400° F. and 450° F. for various times, and the properties of a new type mix baked at 400° F. for similar times.

To be noted is the large variation in properties of the core oil mixture with baking cycle changes. The new type mix offers a decided contrast in this respect varying almost immeasurably as the baking time is increased. Not shown by the data presented is the marked difference in the thermal stability of the two mixes. The new type mix when shock heated shows only very fine cracks if at all, while the comparison mix is prone to crack readily, especially at the shorter baking treatments.

The steel casting as removed from the mold, while it may be the result of our best efforts in melting and molding, is still a relatively crude product and in most cases must be subjected to heat treatment.

Heat Treatment Control

In the as-cast condition, steel casting consist of large segregated grains in a dendritic network formed during the solidification period. In this condition, they possess low ductility and impact properties. To make steel castings more useful from a standpoint of service, the ductility and impact properties must be increased to an optimum value by means of heat treatment.

Heat treatment breaks up the large crystals formed during solidification into much finer grains and homogenizes the structure by diffusion of the segregated constituents. Thus, the ductility and impact properties are increased. Besides increasing ductility and impact properties and relieving casting stresses formed during solidification and cooling in the mold, various types of heat treatment are employed to produce special properties such as increased hardness, toughness and increased tensile strength.

Some of the common types of heat treatments employed by the steel foundries are full annealing, normalizing, tempering, stress-relieving, and hardening and tempering.

The above heat treatments are combinations of heating and cooling operations. The rate of heating and cooling and the temperature to which the steel is heated have a definite influence on the crystalline structure of the steel and the resulting physical properties. The rates of heating and cooling as well as the temperatures employed must then be *controlled* in order to produce the desired properties and in order to produce a consistently uniform product.

Any one heat treatment can be broken down into three cycles; namely:

1. Heating to Temperature
2. Holding at Temperature
3. Cooling.

With the existing variables in a steel foundry in regard to casting size and thickness, it is difficult to have an ideal heating cycle, i. e., to have all parts of the heat treating load at the same temperature at all times. Due to lack of control and lack of knowledge of the furnaces, great temperature gradients can exist in an individual load.

Temperature surveys of furnaces will often reveal shockingly poor

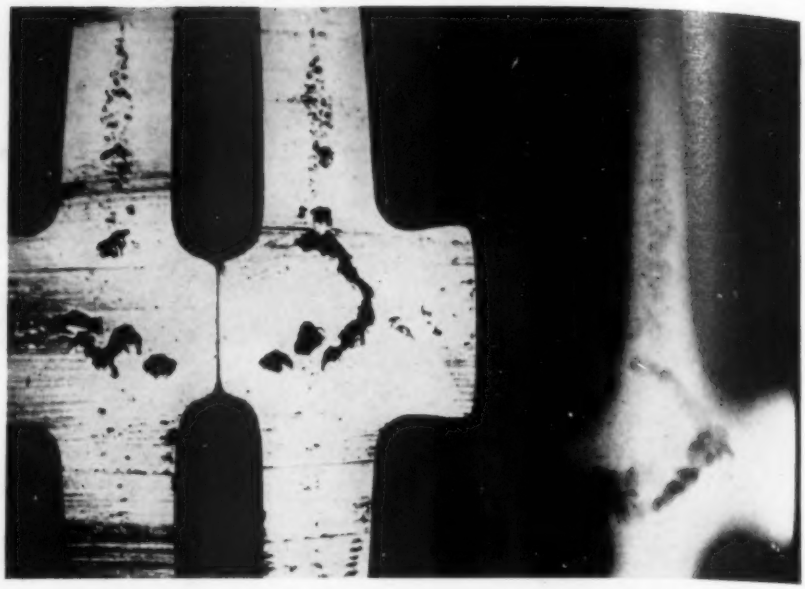
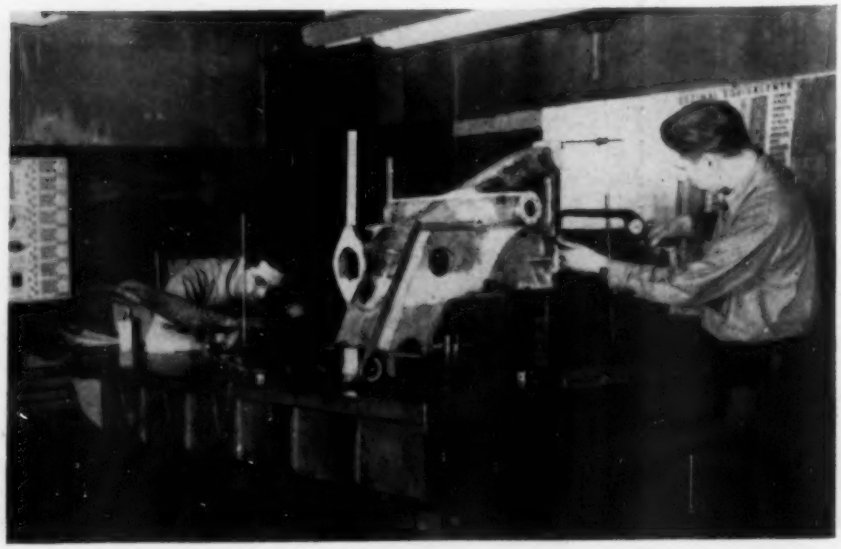


Fig. 27—Exograph (right) revealing internal defect, and sectioned casting (left) which substantiates x-ray findings.

Fig. 26—Layout inspection for metal thickness, dimensions and weight.



temperature distribution, the variations in many cases running as high as 250° F. or higher depending on the method of loading and section sizes involved. By having a good knowledge of the operating peculiarities of each furnace in the foundry and by controlling methods of loading the furnaces, we can strive to obtain optimum heating conditions with the existing variables.

It is one task to insure that all parts of a furnace are at the same temperature and quite a different one to see that all parts of the charge itself are at uniform temperature. Normally, the temperature control is maintained by several thermocouples protruding through the roof of the furnace. Usually the thermocouples rest in a heat resistant pipe above the top of the charge.

In the majority of cases, the temperature indicated by these couples is not necessarily the temperature of the charge itself. The couples merely indicate the temperature in the furnace at that one particular location. The temperature of the charge itself due to the method of loading and less ideal conditions in regard to circulation might be much different. To determine the actual temperature of the charge "check" thermocouples should be placed in the different parts of the charge, top and bottom, front and back of the furnace.

To illustrate the difference in temperature between the charge and that indicated by the control couples, let us refer to Fig. 16.

"Check" thermocouples were placed in the center of 6, 8, 10 and 12-in. blocks and readings were taken periodically to produce the heating curves shown. The control couples reached the desired temperature of 1650° F. in 2¾ hr. However, the blocks did not reach temperature until 4 or 5 hr. later, depending on section size. The length of time necessary for the charge to reach temperature after the control couples indicate the desired temperature has been labeled as "equalizing time."

The equalizing time needs to be determined for different types of loads in each furnace in the foundry. In cases of heavy sections, as described above, the check couples have to be placed in the center of the sections. However, for smaller sections, up to 2 or 3 in., the check couples can be set in small 1 or 2-in. sections and placed in different parts of the charge.

By so checking the temperature of various types of charges in the furnaces, a working knowledge of the operating peculiarities of the furnaces can be obtained. In many cases, the actual temperatures of the charge can be correlated with the temperatures indicated by the control thermocouples so that equilizing time can be determined and a definite practice can be set up for each furnace for reproductive results.

In these "check" runs, the conditions can be varied to study which one will yield best results. The variation in temperature in the different parts of the load, however, can be reduced by the following factors:

1. Better circulation
   a. Furnace car construction
   b. Method of loading
2. Segregation of loads according to thickness of casting sections
3. Location and adjustment of burners.

Increased circulation naturally will reduce temperature gradients. The furnace car should be constructed so that the charge is not placed directly on the bottom of the car. Brick piers built on the bottom of the car make excellent supports for loading of racks in which the castings are placed or for direct loading of larger castings. The brick piers should run across the width of the car so as not to deflect the flame of the burners and should not be built in line with the burners.

Loading of castings is very important. The castings loaded on racks should not form a compact mass, but should be so loaded as to allow circulation between the castings. The charge also should be segregated as much as possible according to section thickness to reduce temperature gradients between light and heavy sections.

The gas pressure of the various burners also plays an important role

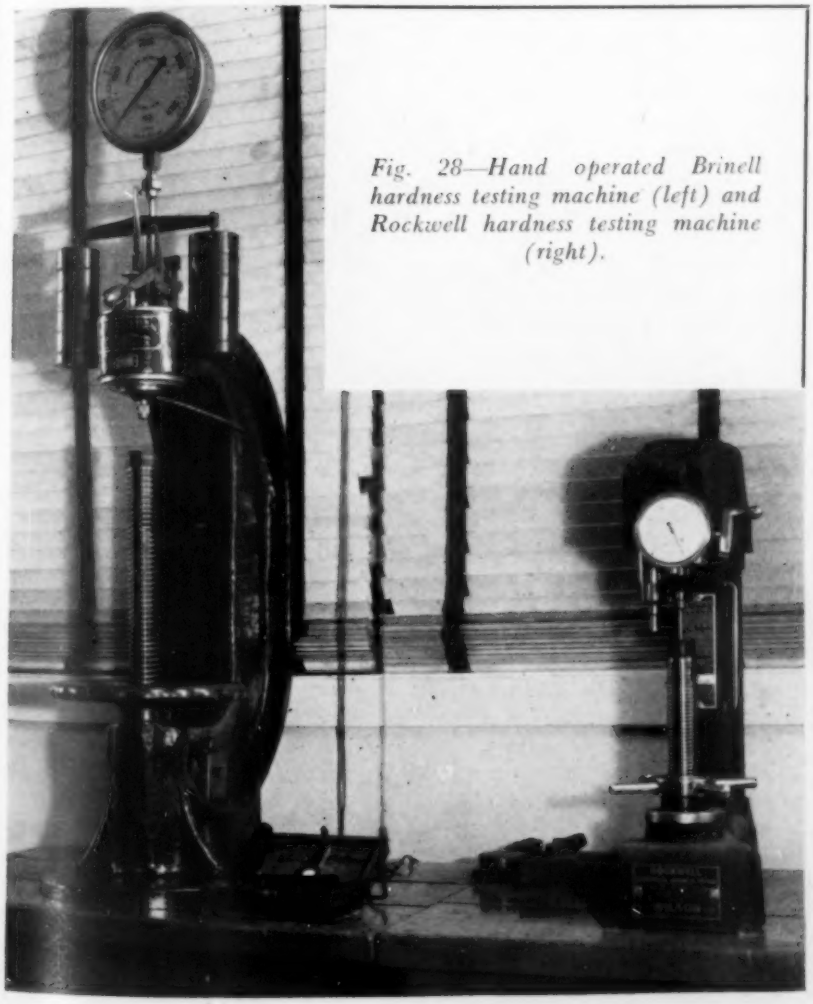


Fig. 28—Hand operated Brinell hardness testing machine (left) and Rockwell hardness testing machine (right).

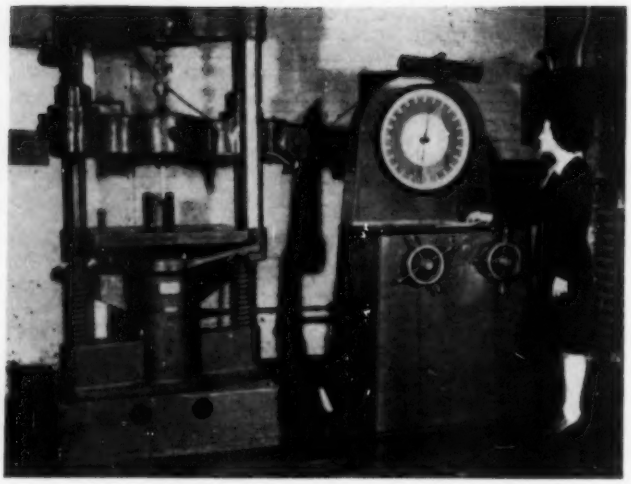


Fig. 29—Determining tensile properties in a tensile machine.

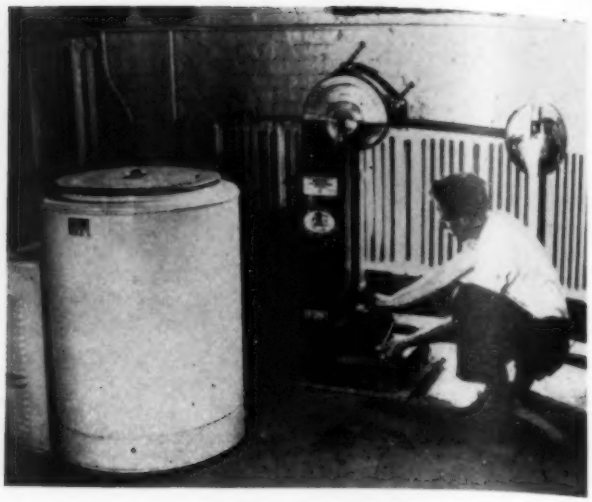


Fig. 30—Determining impact strength of a cast steel specimen.

in controlling temperatures. Each individual furnace will have its own characteristics as to hot and cold spots. By varying the gas pressure of burners in these regions, the temperature gradients can be reduced. By making these changes during an actual furnace check when thermocouples are located in the different regions of the charge, the influence of these changes can be studied and adjustments can be made to operate the furnace to obtain the best temperature control.

Figure 17 shows the construction of a batch type of furnace car with brick piers, method of loading racks of castings, and burner arrangement which would give good circulation.

Also shown are the location of "check" couples used to determine the temperature of the various parts of the charge.

In the heat treatment of large castings, such as 19-in. sections, a step heat treatment should be employed in order to minimize the stresses caused by the difference in temperature between the center and surface of such large castings. Figure 18 shows the results of a "check run" made on a 19-in. section to establish "equalizing times" at the different temperatures.

Step heat treatment consists of heating to a temperature below the critical, holding at that temperature until the entire casting, surface and center, reached the furnace temperature, heating further to slightly above the critical and again repeating the equalizing in temperature, and then heating to the desired temperature.

Knowing when the entire charge has reached the desired temperature, one can easily calculate the holding time at temperature.

With proper furnace control, the temperature gradient when the charge has reached temperature should be within ± 25° F. of the desired temperature and within ± 10° F. for the tempering or stress-relieving treatments.

Metallographic Examination. The microscope is used to determine if the steel casting has been properly heat treated. Figure 19 consists of a series of microstructures of a medium plain carbon steel heated to different temperatures.

Three different heats were used for this study. All heats are of approximately the same composition, but they differ in the deoxidation practice. No aluminum was added to one heat, ¾ lb. aluminum per ton of charge was added to the next heat, and 2½ lb. aluminum per ton of charge was added to the last heat. A study of the microstructures will reveal the role played by aluminum in preventing grain growth. The heat with no aluminum coarsened at 1750° F., the heat with ¾ lb. aluminum per ton coarsened at 1850° F., while the heat with 2½ lb. aluminum per ton showed no sign of coarsening even at 1950° F.

In general, Fig. 19 illustrates the effect of under or over-heating on the microstructure which, in turn, will have a definite bearing on the physical properties especially impact values. We must then, not only have a knowledge of the heat treating furnaces to control heat treatment, but we also must have a knowledge of the steel which is being heat treated. Heat treatment is performed for a definite reason, and, therefore, to receive the full benefit we must control it.

To receive the full benefit of water or oil quenching, the time of quench must be controlled very closely. Cooling curves for the various thicknesses to be quenched must be established by experimental work. Thermocouples inserted in the center of different sections quenched in a similar manner in which castings are to be quenched, will yield results from which cooling curves can be drawn and proper quenching time determined. Figure 20 is an example of such established cooling curves secured from experimental work described above.

In quenching castings, other factors need to be controlled also, such as loading of castings and removing castings from the furnace to the quenching tank. Loading of castings is important in order to insure good water circulation and to prevent warping. The method of removing castings from the furnace is important to insure that the castings remain above the critical temperature up to the time they are placed into the quenching medium. Water temperature and circulation should also be controlled.

Steel Hardenability. Hardenability

is the relative ability of a steel to harden under heat treatment. It is being rapidly recognized to be of great importance and it appears likely that, in the future, hardenability will become a part of alloy steel specifications.

The Jominy end-quench test which has become a standard for measuring hardenability, consists of quenching one end of a 1-in. round steel bar in a stream of water, thus producing a series of cooling rates in the bar ranging from an ideal water quench to air cooling. A quenching jig is used for quenching the end-quench specimens. Rockwell "C" hardness readings on a flat surface produced by grinding off 0.015 in. are taken along the bar and plotted against the distance from the quenched end.

The cooling rates at different positions from the quenched end of the Jominy specimen have been determined. These cooling rates have been correlated with cooling rates of sections which are fully quenched, so that the hardness of various fully quenched sections can be determined from the curves established from results of the end-quench test.

Through the end-quench hardenability test we can very readily choose steels which will yield the desired hardenability in the sections involved. Figure 22 contains hardenability curves of four different steels on a standard form.

Through the use of these curves, the hardenability of various compositions can be studied and compared. The end-quench hardenability test is also an excellent method to study effects of different alloying elements on hardenability.

While we have discussed at some length the methods of control which are in common use to insure the production of a quality product, there must necessarily be other methods of checking and controlling to insure the casting meeting the requirements of the drawing or specification used in its manufacture.

Inspection Control

Inspection has often been referred to as a necessary foundry evil. This criticism is far from the truth. A well trained Inspection Department is the only known method in maintaining foundry control and assuring a uniform quality in castings.

The dictionary defines inspection as a "critical investigation." But this term does not define how this investigation is to be made, what tools are available, or how the information secured during the investigation must be disseminated or used.

A well organized Inspection Department must not only disclose a defect, but also its cause and to which department the defect is chargeable. This, however, does not mean that the Inspection Department is required to know the remedy for the elimination of the defect.

An Inspection Department uses various methods and tools in conducting their investigation. This condition is created by an effort to eliminate or replace the human element with precision instruments that are not influenced by human judgment. In recent years, great advances have been made in inspection tools such as magnetic particle testing, gamma rays, and X-rays and these implements have been used successfully in foundry control. Greater advances can be anticipated as progress is made in the field of electronics.

Inspection in a modern foundry generally follows a definite pattern. Therefore, the writer shall endeavor to illustrate how a well trained Inspection Department operates and how it controls production and the quality of a finished casting.

Roughing Floor Inspection

The first inspection area is the roughing floor. In this area gates, risers, and other excess metal are removed by flogging or burning. Here also is stationed the first inspector whose duty is to separate usable castings from defective ones. See Fig. 23.

At this inspection station, all major defects are classified as follows:

1. General appearance
2. Burned-on sand
3. Riser location
4. Condition of riser
5. Fins
6. Location of chills, chaplets, nails
7. Blows
8. Misruns
9. Core shifts
10. Hot cracks

It is, therefore, quite evident that this inspector must have considerable ability and must be familiar with foundry practice pertaining to the different patterns being produced.

Here is the first instance that inspection leads to control. The inspector fills out a "Report of Discrepancies in Castings."

In this report are included Pattern Number, Drawing Number, and Description of Trouble. This report is then sent to the Process Inspector who relays it to the department to which the trouble has been charged. Also on the same report is the "Action Taken on Report on Discrepancies in Castings" which must be filled in and returned to the Chairman of the Planning Committee.

Following the above procedure, you will note that inspection has not only made a critical investigation, but also evaluated the cause and forwarded this information to the department responsible for the cause.

Rough Inspection

The second inspection area is the "rough inspection" illustrated in Fig. 25.

Here minor defects, such as cracks, tears, dirt, and design inconsistencies which are uncovered after burning, shot blasting, and chipping operations, are noted. Here, again, all discrepancies are listed on the "Report on Discrepancies in Castings." This inspection area is also given the job of seeing that all minor defects are properly removed before being welded.

Final Inspection. The third inspection area is generally known as the "final inspection," and is shown in Fig. 25.

All welds, gauging operations, hardness readings as well as the general appearance of the finished casting are noted here. Any discrepancies caused by improper cleaning and finishing are brought to the attention of the department in error so that better control is maintained in these departments.

Outside Inspection

Probably the most important function of a good Inspection Department and the one most overlooked is that known as "outside inspection." In this function, it is the duty of an inspector to contact all customers receiving castings to check if any discrepancies were overlooked

during the plant inspection or if there are any discrepancies which were not uncovered by routine inspection methods.

These discrepancies are hard spots found during machining, untempered welds in machined areas, internal defects, holes, excess shrinkage, improper removal of sand in cored areas, etc. It is also this inspector's duty to ascertain that his plant and the customer are using the same inspection procedure.

The above covers what is generally considered production inspection and its aid in maintaining foundry control. This does not include all the work of an Inspection Department. We stated at the beginning that the Inspection Department utilized various methods and tools in conducting their investigation, and it is these tools and how they are used by the Inspection Department that will now be discussed.

A Layout Inspection Department is shown in Fig. 26.

In this department sample castings are checked for:

1. Metal thickness
2. Drawing dimensions
3. Weight

All discrepancies are noted on form "Order and Report on Sample Castings."

The Layout Department is also utilized to check production castings with regard to changes in metal thickness due to:

1. Normal pattern wear.
2. Excessive wear on pattern equipment due to design.
3. Weight.

Magnetic Particle Testing. Magnetic particle testing is one of the more specialized tools used by the Inspection Department in locating defects which are generally classified as discontinuities on the surface or slightly subsurface. The defects are usually attributed to design, grinding, burning or heat treatment.

Radiographic Inspection. Probably the best and most modern tool used in recent years by the Inspection Department is radiography.

It is a means to determine existence of any discontinuity within a casting. The discontinuities uncovered by radiography are classified as shrinkage, entrapped gases due to improper deoxidation or improper molding, internal tears or cracks, the presence of internal chills, chaplets and also non-metallic materials such as sand, slag, and other refractory materials (Fig. 27).

All defects located with these special tools are recorded and this information is relayed to the department whose duty it is to correct their processes so that the defects are eliminated or reduced to an acceptable amount.

The entire discussion so far has been the methods used by the Inspection Department to ascertain the quality of a product and to insure foundry control by experienced visual examination and highly specialized equipment.

Physical Testing. The methods of inspection to be discussed now are generally referred to as physical tests and include hardness, tension, and impact tests. All these tests are made with precision instruments and the results obtained are used in ascertaining quality of a product with regard to what can be expected of it in service.

Hardness Test. One of the most used physical tests is that of determining hardness and it is used in the control of the production of heat treated castings. It is often used to determine the machinability of a casting. There are approximately 36 different instruments manufactured to determine hardness, but the ones most generally used in a foundry are the Brinell Hardness Testers, either fixed or portable, and various Rockwell Hardness Testers (Fig. 28).

Tensile Test. Another test which is made continually is the tensile test (Fig. 29).

Tensile machines are produced by various manufacturers, but all of them are similar in that a tension load is applied to a specimen at a uniform rate. Usually, as far as a foundry is concerned, this test is used as a control test for production or as an acceptance test for specifications.

Impact Strength Test. Impact testing has come to the front in recent years and is being received more favorably. The method of testing is to apply a sudden blow to a specimen sufficient to break it and to measure the energy absorbed by the specimen as it breaks. Various machines are used, but all are similar (Fig. 30).

The only variation in the different methods is how the specimen is held and the method of preparing the notch.

In the Izod test the test specimen is held rigid in the vertical position with the notch horizontal. In the Charpy test, the specimen is laid across two supports in a horizontal position, the notch vertical.

Various notches have been advocated, but at the present time, the V-notched Izod or Charpy and the keyhole Charpy specimens are used.

This test at the present time is being used in the control of heat treating processes. It has also been used to determine the properties of impact resistance in steels at room temperatures and at elevated and subnormal temperatures.

It has been necessary to discuss methods of control only briefly. It is hoped that this quick tour through the lesser known side of the steel foundry has conveyed some idea of the means that are available to measure or weigh what is going into the product. Only by keeping one hand firmly on the "controls" can we be assured that our heat to heat or day to day production is up to or even better than standard.

DISCUSSION

Chairman: C. E. SIMS, Battelle Memorial Institute, Columbus, Ohio.

MEMBER: I would like to have Mr. Lillieqvist explain more fully the rapid silicon determination.

MR. LILLIEQVIST: The setup consists of a small cup made of a dry core mold. A sampling spoonful of molten steel is poured into the core mold and the reaction is observed. If the silicon content is low, the molten steel will rise in the cup. If the silicon content is slightly higher, the surface of the steel in the cup will be straight or level. If the silicon content is approximately 0.30 per cent, the steel in the cup will shrink. This condition is shown in Fig. 3. In many cases, we had heats where the silicon content went up to 0.35 or 0.40 per cent, especially when the metal temperature ran as high as 3200° F.

MEMBER: Specifically, I wanted to know the action of it and the amount.

MR. LILLIEQVIST: You can only guess at plus or minus five points, but it is sufficiently good for that purpose.

MEMBER: Is it in dry sand?

MR. LILLIEQVIST: Yes, it is in dry sand.

MEMBER: Could you make the test without dry sand?

MR. LILLIEQVIST: You will get a different condition if you pour the steel into chilled molds. We tried chilled molds first and did not get as good results as when we used dry core molds.

MEMBER: Do you have to be sure

there are no blow holes in the castings?

MR. LILLIEQVIST: Yes. It is through control of the silicon content that we get away from pinhole porosity. You also have to use aluminum as a deoxidizer especially in the acid electric furnace and the basic open hearth. You do not have to use aluminum in the acid open hearth.

CHAIRMAN SIMS: This silicon test is quite interesting, in view of some of the early experiences with castings. About 20 or 30 years ago, it was customary to pour steel castings with a silicon maximum of about 0.30 per cent. Just how that range was ever attained, it is difficult to say, but that was common practice in the early days of casting steel. They did get into much porosity, but porosity was not considered as serious then as it is now. I do not think the inspection was as close.

Some tests were made a number of years ago to determine the minimum amount of silicon necessary to suppress porosity in a steel casting if one could eliminate such factors as wet sand. We arrived at the conclusion at that time that 0.25 per cent silicon was right on the border between porosity and no porosity and to have a margin of safety one must have had at least 0.30 per cent silicon. That seems to hold pretty well with these tests.

Now, pouring steel in a cast iron mold, as they do in ingot practice, they consider that they get a killed steel with as little as 0.15 per cent silicon. There the quicker chilling seems to have some influence.

J. A. DUMA[1]: On your cooling curves on heavy sections, with water quenching, I think you said about 2 hr. cooling time was required for a 15-in. section.

MR. LILLIEQVIST: For an 18 in. thick, 48 in. diam. section it takes 2 hr. 20 min. to cool the center of the section in water to a temperature of 350° F.

MR. DUMA: Is it your practice to cool the entire casting to room temperature or the temperature of the water?

MR. LILLIEQVIST: No, the practice is to cool the casting to 350°-400° F.

MR. DUMA: At what temperature did you temper it?

MR. LILLIEQVIST: That depends upon the hardness desired. If 350 Brinell is desired, temper between 800° and 1,200° F. depending upon the composition of the steel.

MR. DUMA: It would not harm at all to withdraw your casting from the quench tank when the center of the thickest section has reached the drawing temperature.

MR. LILLIEQVIST: I believe it would be dangerous to withdraw it earlier.

MR. DUMA: We have used that practice extensively in hardened steel roll practice where alloy steels are quenched to about 100 scleroscope. The drawing temperature employed is from 275°-475° F.

MR. LILLIEQVIST: What composition do you want?

MR. DUMA: 44-85 and 62-85.

CHAIRMAN SIMS: I believe this question might be compromised in this manner. Mr. Lillieqvist, I think, is perfectly right in his assumption that when your steel is under-quenched you will often get into poor properties. It was shown definitely in investigations of some of the armor casting during the war, that with under-quenching poor properties were obtained. However, it may be you will not have to take the center of the casting down to a temperature below the martensite transformation before you withdraw it from the water for the reason that the outside is probably down to room temperature. There is considerable lag between the outside and inside. The heat is flowing to the surface and the lag, of course, is entirely due to the slow heat conductivity through the steel.

Now if the outside is down to say 40° F., which may be the temperature of the water, and the inside at 500° F. or 600° F., you could take the block out of the water at that time and the inside would continue to cool down because of the heat flow to the surface. The heat would flow to the surface almost as fast as if you left it in the water. By the time it equalized the whole block may be below 350° F. and at that temperature you would have both the condition that Mr. Lillieqvist desires and that you desire.

MR. LILLIEQVIST: We took an 18 in. thick, 48 in. diam. casting and quenched it to water temperature and obtained a large quenching crack. In order to avoid quenching cracks, the next casting was heated to 1650° F., then cooled in the furnace to 1400° F. and then quenched in water to 350° F. No quenching crack was obtained.

MR. DUMA: I have just one more question and that is on molding control. You have given us four controls, namely, heat treatment, sand processing, inspection and melting. What do you do in the way of molding control, that is arrangement of gates, risers and the planning and designing of the casting to hold down defects to a minimum?

MR. LILLIEQVIST: That goes mostly to our Planning Committee, the foundry shop's chief metallurgist, chief chemist, chief inspector and assistant works manager, and they decide what the heading and gating shall be.

MEMBER: At what point in the heat is the fluidity test taken?

MR. LILLIEQVIST: From the electric furnace, we have not taken as many fluidity tests. In our acid electric practice, we pour the steel into a teapot ladle and then into small hand shanks. It makes a big difference if the hand shank is at a temperature of 600° F. or if it is at a temperature of 1200° F. and having only 100 or 150 lb. of metal. In the basic open hearth, we know the temperature of the ladle and we obtained satisfactory results, but in the acid electric, we do not have sufficient information.

Unfortunately, the human element is involved in this test, hence it has to be well standardized. Keep your sampling spoon about 5 sec. in the furnace, slag the spoon then go down 3 or 4 in. below the slag, take the steel out of the furnace and put aluminum in and then pour. It has to be accurately timed in order to get good results. In order to get away from the fluidity test, we are going to install equipment for measuring bath temperatures at one of our plants.

C. H. LORIG[2]: Have you noticed any effects of melting practice on the fluidity of steel? Assuming the temperature of steel is the same and the composition is the same, have you noticed any effect from processing?

MR. LILLIEQVIST: We know, for instance, in the basic open hearth if you block your steel with 15 per cent ferrosilicon and let the steel stay there about 10 min., your fluidity is rather low, but if you let the steel come back, in order to get the boil, your fluidity will come back again.

CHAIRMAN SIMS: I would like to ask, in the case of the fluidity test falling below your standard, what you would do about it?

MR. LILLIEQVIST: Leave it in the furnace longer. Add some spiegeleisen to the bath and get high temperature.

MEMBER: Is there any relation between the amount of oxygen present in the steel and your fluidity?

MR. LILLIEQVIST: I do not know.

CHAIRMAN SIMS: The oxygen is pretty well fixed by the carbon content in the first place and by whatever deoxidation practice may have been used.

MEMBER: I have a case of basic electric steel and we have had a temperature of well over 3,000° F. and we had sluggish steel. During the next heat we added some oxide to the slag, iron oxide and in pouring the same casting we had much more fluidity and there seemed to be a definite relation between the amount of oxygen present and the fluidity of the spiegel.

CHAIRMAN SIMS: There are conditions of that kind that are not too well explained. Oxygen analysis would probably have shown no difference in oxygen content.

MEMBER: We went too far on the carbide side.

CHAIRMAN SIMS: It may well be, as pointed out earlier, that either hydrogen or nitrogen absorption may make the difference.

MEMBER: You would naturally expect the oxygen.

MEMBER: The iron-oxide would give the necessary boil, would it not?

CHAIRMAN SIMS: Sufficient iron oxide would certainly cause a boil under the conditions you mention. After killing a steel, and in the presence of a carbide slag and the possibility for gas absorption is favorable. Steel may absorb hydrogen very rapidly in that condition.

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UNSOUNDNESS IN CAST LIGHT ALLOYS

A.F.A.
Aluminum and Magnesium
Division
Subcommittee on
Shrinkage and Porosity

MANY INVESTIGATIONS AND opinions have been published regarding the occurrence of unsoundness in aluminum and magnesium alloy castings. As with other alloys, many types of unsoundness in light-alloy castings are possible, but this paper will deal mainly with unsoundness which arises (1) from shrinkage, (2) from the gas evolved from the melt when it solidifies, or (3) from the combined effects of both.

Gas holes resulting from the mechanical entrapment of mold gases in the casting or defects from dross, which is actually a mixture of melt, oxide, gas, or air, and the many other types of defects leading to voids in the casting will only be mentioned.

Furthermore, the unsoundness in the light-alloy castings, as discussed in this paper, will be confined to castings produced by either the sand-cast method or permanent-mold method, but not by the die-cast method.

An extensive literature review does not appear to be necessary. It is sufficient to point out that there is considerable confusion among foundrymen, metallurgists, inspectors, etc., who have occasion to deal with the problem of defects arising from dissolved gases and from shrinkage in light-alloy castings.

However, if several scores of such individuals were discussing the problem together, and if all used the same terms of the same meaning, the chances are reasonably good that for the most part they would find themselves substantially in agreement on the general subject of unsoundness arising from gas evolution and shrinkage in light-alloy castings.

One of the factors leading to a confusion of terms, and consequently to differences of opinion, is the fact that in both magnesium and aluminum alloys unsoundness arising from gas evolution and from shrinkage are closely related. Furthermore, there are all gradations between localized shrinkage on the one hand, and microporosity on the other.

In aluminum alloys there are also all gradations between gas porosity (pinhole porosity) and microporos-

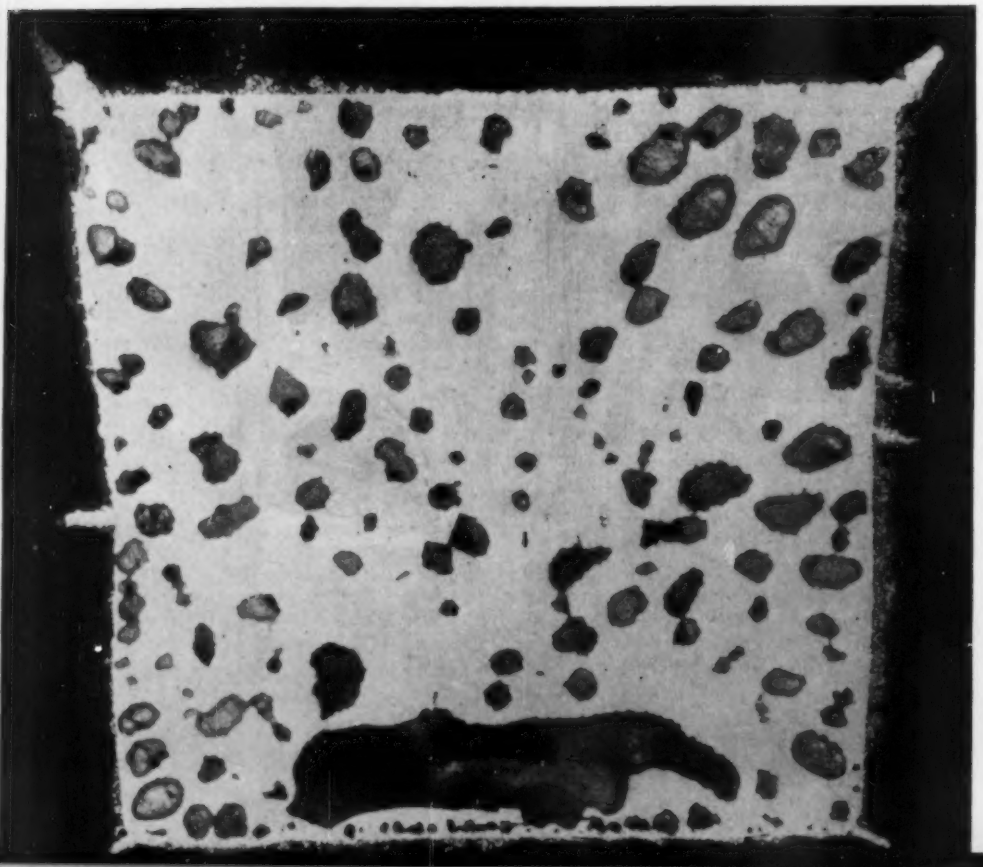


Fig. 1—Section of a 4-in. cube, no riser, poured from an ASTM-AZ-63 alloy melt which had been severely gassed with CaH_2. The large cavities formed by the hydrogen precipitated when the melt solidified are ample evidence of the considerable hydrogen solubility in magnesium melts. These "gas holes," which may be much finer where solidification is more rapid, correspond to pinholes having a similar origin in aluminum-base castings. ×1.

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ity. The primary purpose of this paper then will be to supply additional, and possibly clarifying, information on the following points:

1. Relationship of shrinkage to microporosity, and the gradation of one into the other.
2. Relationship of gas porosity to microporosity, and the gradation of one into the other.
3. Causes of the various types of unsoundness, and the effects of several important factors on the form of the defects.
4. Terms employed to describe the defects, some of which have already been mentioned in this paper.
5. A correlation of the appearance of defects as viewed on the as-cast or machined surface and by radiography.

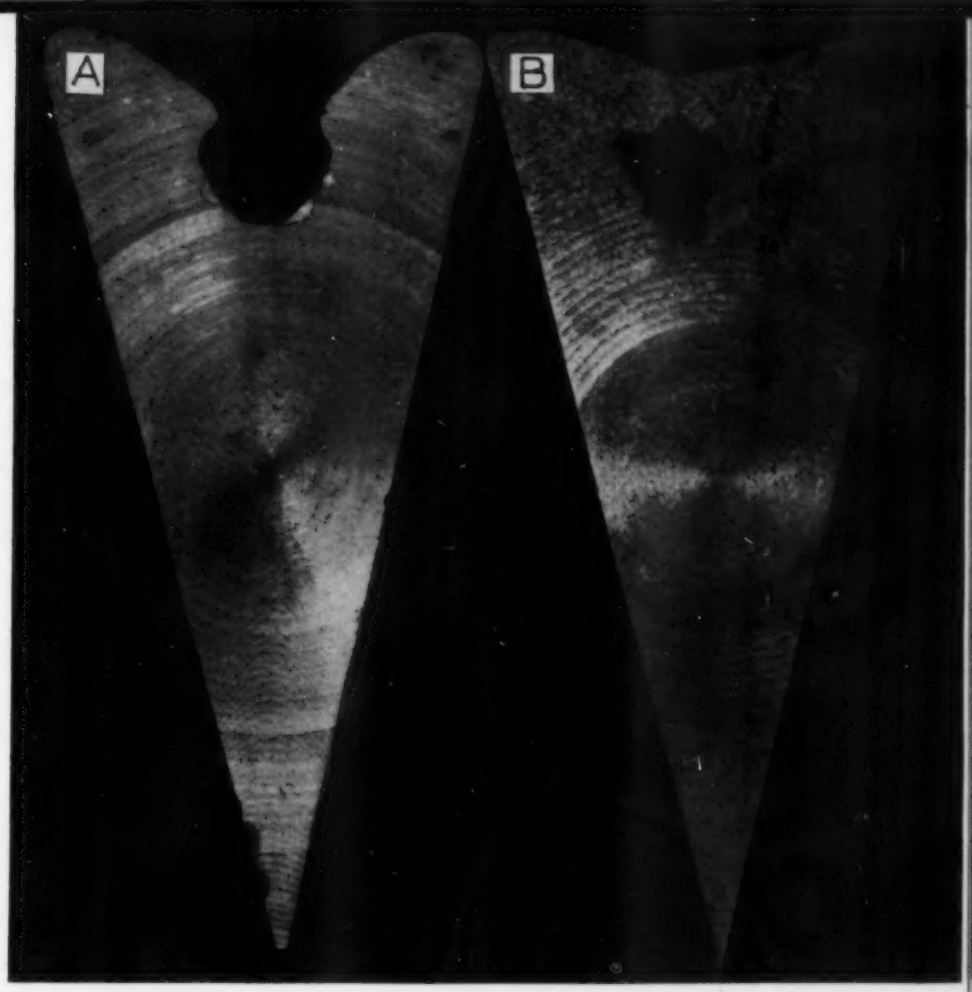
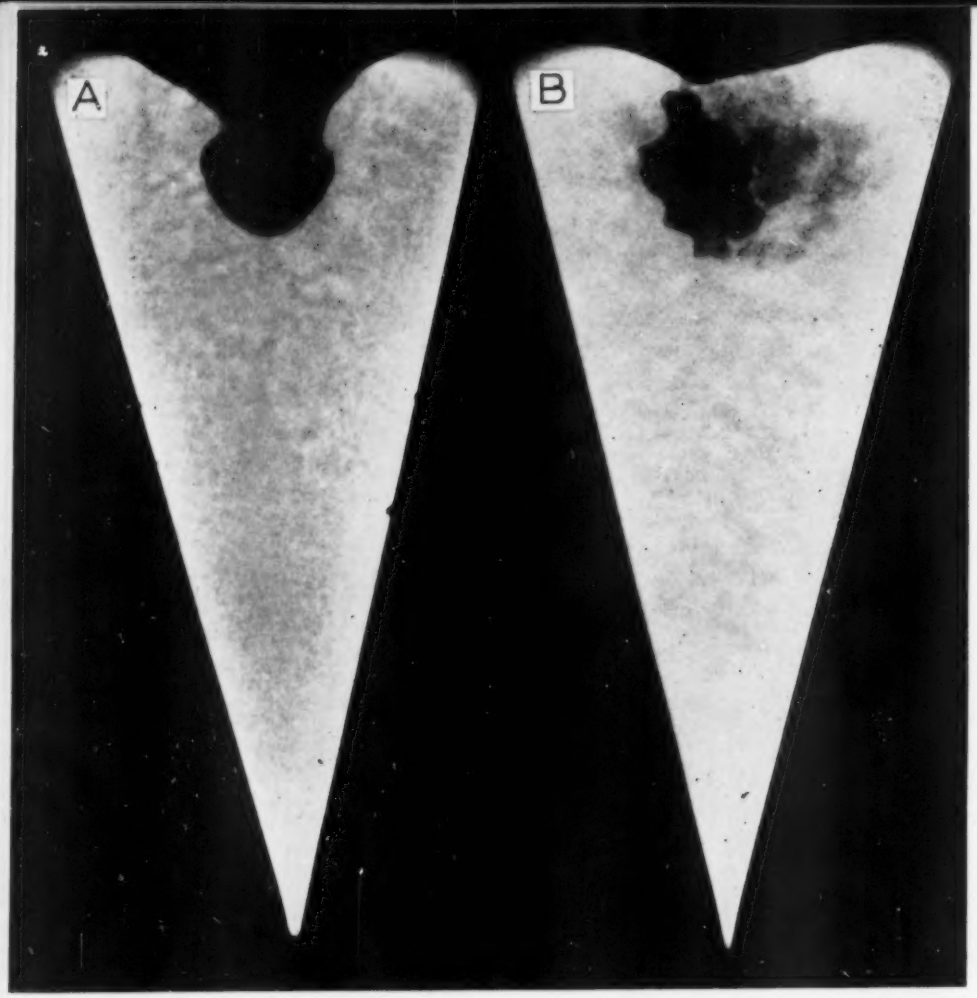
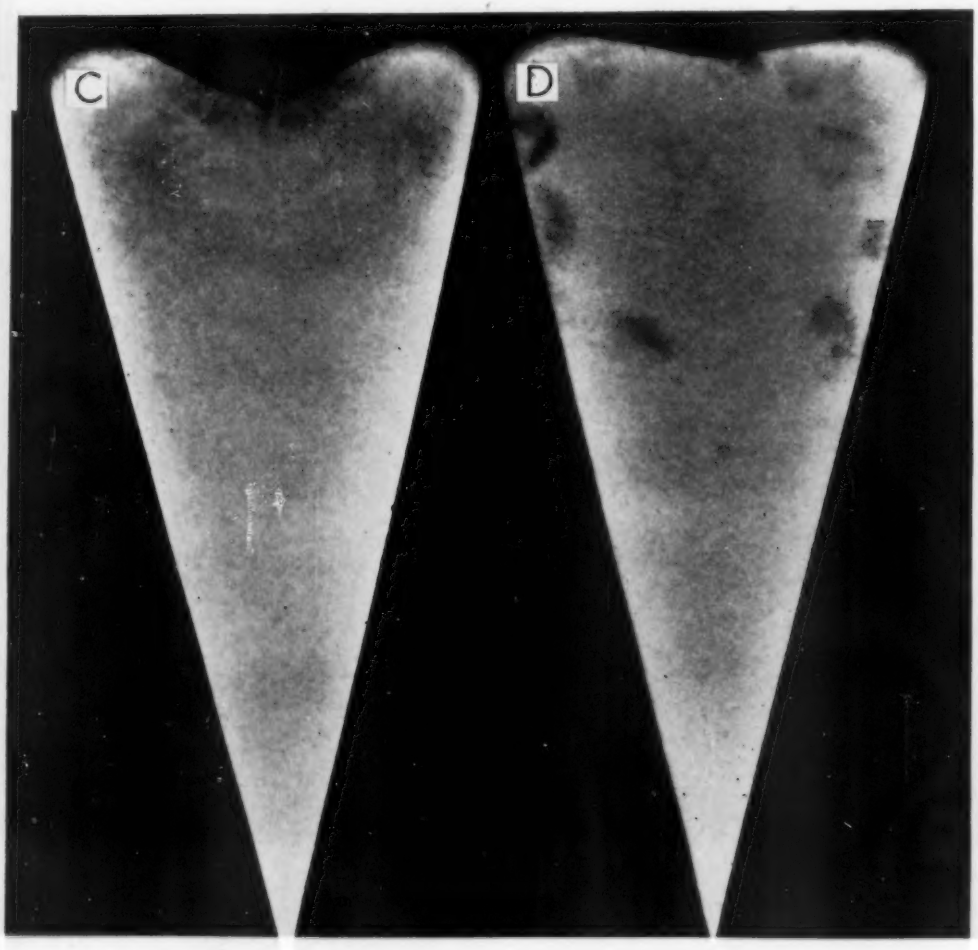


Fig. 2—(right above) Machined vertical sections of degassed metal cast in dry sand. A—99 per cent Al. B—No. A-132 alloy containing 0.8 per cent Cu, 0.8 per cent Fe, 12 per cent Si, 1 per cent Mg, and 2.5 per cent Ni. ×5/6.



Fig. 3—(right) Machined vertical sections of degassed metal cast in dry sand. C—No. 214 alloy containing 3.8 per cent Mg. D—No. 122 alloy containing 10 per cent Cu, 1.2 per cent Fe, and 0.2 per cent Mg. ×5/6.





Consideration will be given to the general relationship of the various types of unsoundness, and specific examples of unsoundness will be shown to illustrate the general principles.

Having specifically indicated the purpose of the paper, it will be well to point out here that no attempt is made to evaluate the importance of these defects on the properties or the serviceability of the castings containing them.

No formal attempt will be made to prepare a complete classification of terms. Instead, to the various forms of unsoundness, terms will be applied which are recommended (1) because they describe the cause of the defect and its appearance, or (2) if this is not possible, a term will be used which describes the defect but does not suggest an erroneous explanation of the cause of the defect.

In order to reduce the number of illustrations, reference will be made to photographs in other A.F.A. papers.

General Features of Unsoundness Caused by Shrinkage and Gas Evolution

Shrinkage. It is well known that when a liquid melt changes to a solid a contraction in volume occurs, which normally amounts to 4 to 6 per cent of the total volume of light-alloy melts, and this contraction will result in certain types of unsoundness known as shrinkage unless the casting is fed properly. The proper feeding involves the entire problem of proper casting design, pattern design, and the gating, risering, and chilling methods employed. The amount, shape, and distribution of this defect, known as shrinkage, will depend upon a number of factors. The main factors are:

1. Composition of the melt; namely, its content of various elements and metallic impurities.
2. Gas content of the melt.

Fig. 4—(left above) Same as Fig. 2, except that radiographs of vertical sections 1 in. thick are illustrated. ×5/6.

Fig. 5—(left) Same as Fig. 3, except that radiographs of vertical sections 1 in. thick are illustrated. ×5/6.

3. Manner and rate of solidification, and the degree of the directional solidification, as dictated by pouring temperature, the casting design, gating method, section thickness, etc.

The effect of some of these factors will be illustrated later in the paper.

Unsoundness Caused by Gas Evolution. It is also well known that in the light alloys of aluminum and magnesium the main if not the only gas which causes trouble is hydrogen, and it produces unsoundness as a result of the decreased solubility of the hydrogen in the melt when it solidifies. As a result of this decreased solubility, the gas is evolved.

Hydrogen Solubility

While the maximum solubility of hydrogen in aluminum depends upon several factors, in general, only a maximum of 2 to 4 cc of hydrogen per 100 grams of metal are dissolved in the melt. At ordinary pressure, however, the equilibrium solubility of hydrogen in the solid aluminum, even near the solidus line, is almost zero.

As a result, practically all of the gas in solution in the aluminum melt is normally evolved during solidification. When this gas is evolved, it forms a defect in the casting known as pinhole porosity. This porosity may be round or angular, coarse or fine. There are all gradations of this type of gas porosity, from coarse, rounded holes on the one hand, through angular pinholes to very fine porosity known as microporosity on the other.

It cannot be overemphasized that pinhole porosity is caused only by the evolution of hydrogen from the aluminum-alloy melt when it solidifies. There are a number of factors which affect the amount, size, shape, and distribution of the pinholes.

However, this subject has been

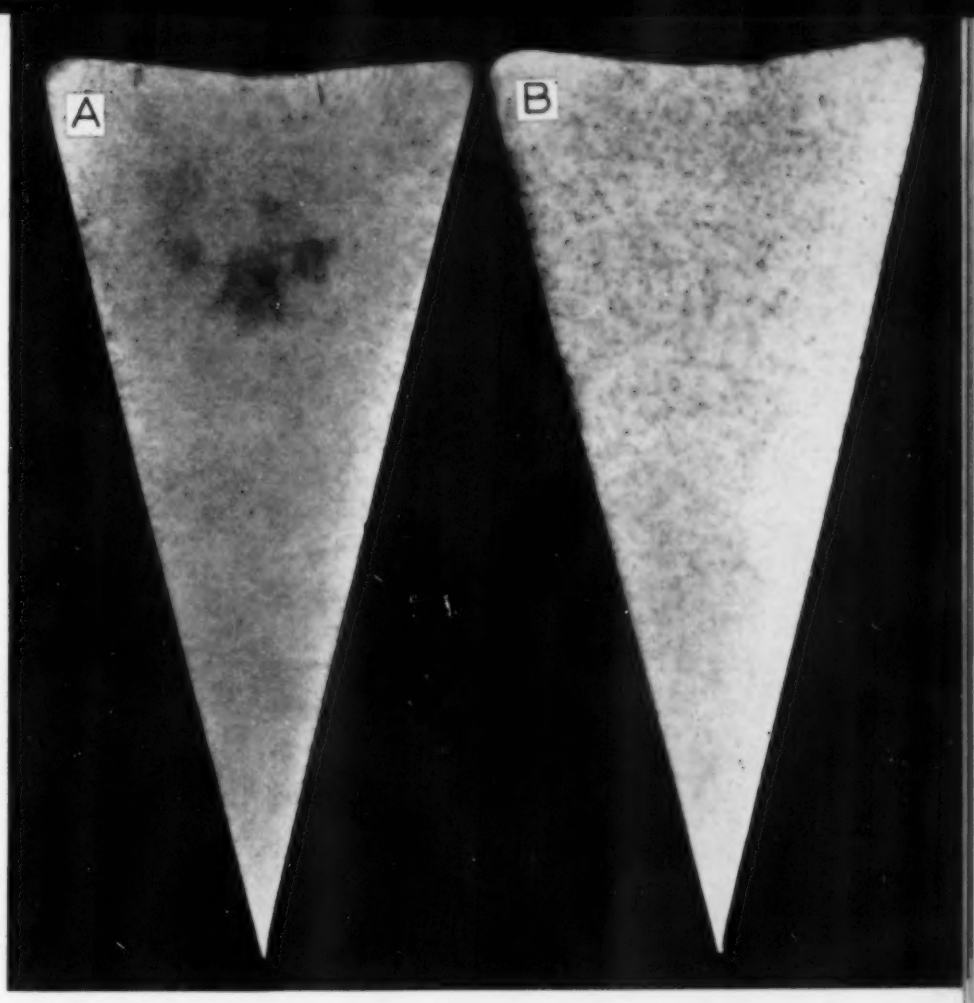


Fig. 6—(right above) Machined vertical sections of an aluminum-copper-silicon alloy. A—Poured after degassing by fluxing with chlorine. B—Poured before fluxing with chlorine. ×5/6.

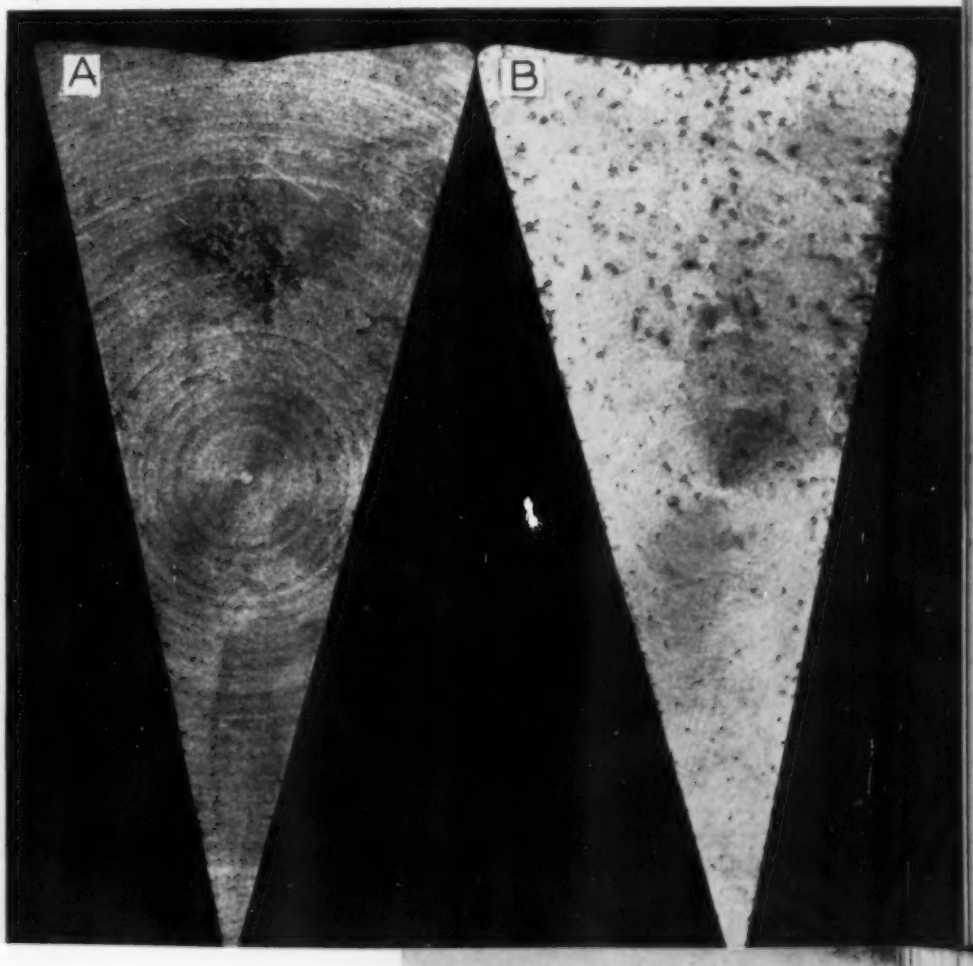


Fig. 7—(right) Same as Fig. 6, except that radiographs of vertical sections 1 in. thick are illustrated. ×5/6.

considered in detail[1] and a further account of this problem need not be made here. It is sufficient to point out that the following factors are important in determining the amount, size, shape, and distribution of the unsoundness resulting from gas evolution in aluminum melts:

1. Alloy composition.
2. Gas content of the melt.
3. Grain size.
4. Solidification rate, degree of directional solidification, and manner of solidification as determined by casting design, gating practice, etc.

In contrast to aluminum casting alloys, magnesium alloys contain a maximum solid solubility near the solidus line of about 18 cc of hydrogen per 100 grams of metal, and this may increase to a maximum of 26 to 30 cc of hydrogen per 100 grams of melt in the temperature range 1200 to 1700 F.

When these high gas contents are encountered in the magnesium foundry, it is found that the risers swell, forming surfaces convex upwards, and that the casting contains large, rounded, or angular voids. They are quite similar to the pinholes in aluminum, but they are larger (perhaps ¼-in. diameter or more) and, therefore, cannot be accurately called a pinhole. Instead, they may be called "gas holes."

An outstanding example[3] is illustrated by Fig. 1. Normally, however, such high gas contents are not encountered in magnesium foundry operations because (1) solid solubility of hydrogen in magnesium near the solidus line is high, and (2) the gas which is in the melt accentuates microporosity instead. Therefore, these large gas voids are not usually observed. However, they fit into the general over-all picture of the problem of unsoundness in light-alloy castings.

Unsoundness

As with the other types of unsoundness, the unsoundness resulting from gas in magnesium-alloy melts may manifest itself in quite a variety of defects. As later illustrated, all gradations of rounded or oblong gas holes through angular, slit-like gas holes to small interconnected angular pores referred to as microporosity may be observed. Again, the main factors which determine the amount, size, shape. and distribution of the voids resulting from the gas in magnesium melts are:

1. Alloy composition.
2. Gas content of the melt.
3. Solidification rate, the degree of directional solidification, and other factors affecting the manner of the solidification of the casting.

As illustrated later, in either aluminum or magnesium alloys the unsoundness may grade from highly localized shrinkage through various more generally distributed types of unsoundness referred to as "shrinkage of a spongy type" to microporosity.

Likewise, the unsoundness resulting from gas evolution will grade from pinholes in aluminum or gas holes in magnesium through various angular pores to fine angular porosity known as microporosity. Hence, both shrinkage and the gas content of the melts combine to determine the amount, size, shape, and distribution of microporosity in the castings. In addition, the gas evolved will influence the amount of localized shrinkage.

Factors Affecting Shrinkage Porosity

The various main factors affecting the amount, size, shape, and distribution of shrinkage defects have already been listed. These factors were alloy composition, gas content of the melt, and the manner of solidification of the casting.

Effect of Alloy Composition. The type of shrinkage with which most foundrymen are familiar is localized shrinkage. The most localized shrinkage consists of a surface set or cavity. The effect of composition on the type of localized shrinkage in aluminum castings is well illustrated by Figs. 2, 3, 4, and 5.

The test castings are wedges 6 in. high, 3 in. wide at the top, and 2 in. thick, and a vertical cross section at the mid-point of the casting is illustrated. These melts, degassed by chlorine fluxing, were poured into the top of the mold cavity in a dry sand mold without the use of a cope or pouring sprue and gate. The compositions of the alloys shown in Figs. 2, 3, 4, and 5 are:

A. Ninety-nine plus per cent Al.

B. No. 132, aluminum containing approximately 0.8 per cent Cu, 0.8 per cent Fe, 12.0 per cent Si, 1.0 per cent Mg, and 2.5 per cent Ni.

C. No. 214, aluminum containing approximately 3.8 per cent Mg.

D. No. 122, aluminum containing approximately 10 per cent Cu, 1.2 per cent Fe, and 0.2 per cent Mg.

These four alloys illustrate quite well the effect of composition. Composition *A* is a substantially pure metal solidifying at constant temperature. Composition *B* is nearly 100 per cent eutectic which solidifies over a narrow range of temperature but not entirely at constant temperature. Composition *C* is a solid solution type solidifying over a considerable range in temperature, but no eutectic is formed. Composition *D* solidifies over a considerable range in temperature, but a large amount of eutectic with more intermetallic compound is formed.

Machined Surfaces

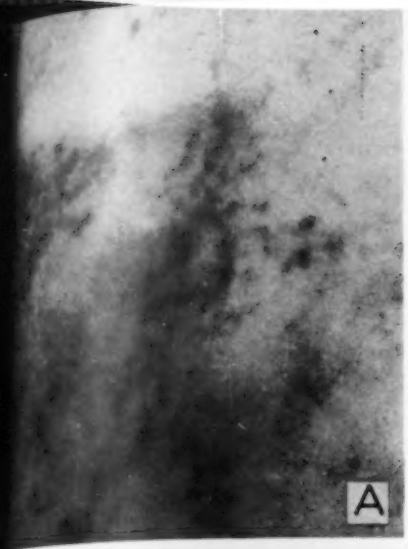
On the basis of the examination of Figs. 2, 3, 4, and 5, the first two of which are photographs of machined surfaces and the last two radiographs of the same specimens 1 in. thick, the following conclusions can be made:

1. With a pure metal solidifying at constant temperature, the shrinkage is highly localized; namely, it forms a surface set with a clean cavity with little or no surrounding spongy metal, as shown by Fig. 2 (composition *A*), and the radiograph, Fig. 4 (composition *A*).

2. When the alloy solidifies nearly but not entirely at constant temperature, such as composition *B*, the shrinkage again is highly localized in the form of a surface set and an internal cavity, but surrounding this cavity is a zone of spongy metal. Had composition *B* been 100 per cent eutectic, the unsoundness probably would approach in appearance that illustrated by composition *A*. The spongy area around the shrinkage cavity in composition *B* is well illustrated by the radiograph (Fig. 4) of this same specimen.

3. Composition *D*, which solidifies over a range in temperature with a considerable amount of eutectic solidifying at the end, shows a minor surface set; the shrinkage is not localized, but rather widely distributed, as indicated by the radiograph (Fig. 5, composition *D*). This localized shrinkage is not much in evidence on the machined surface (Fig 3).

4. Composition *C* solidifies over a range in temperature without the



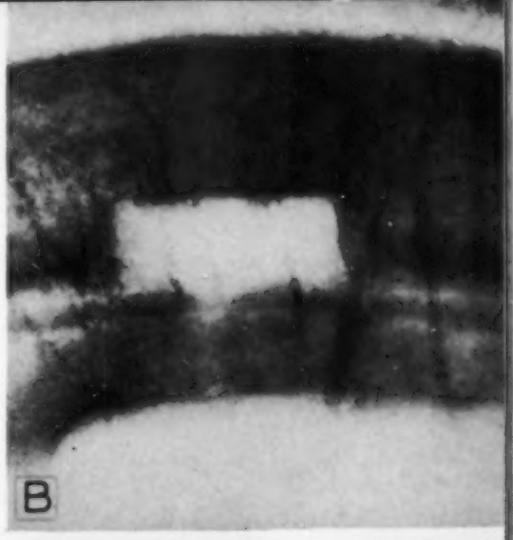


Fig. 8—(A) Radiograph showing shrinkage cavities in an inadequately fed 4-in. thick section of No. 220 alloy containing 10 per cent Mg. (B) Radiograph showing shrinkage cavities of a peculiar type in an inadequately fed ½-in. thick section of No. 355 alloy containing 1.2 per cent Cu, 5 per cent Si, and 0.5 per cent Mg. ×1.



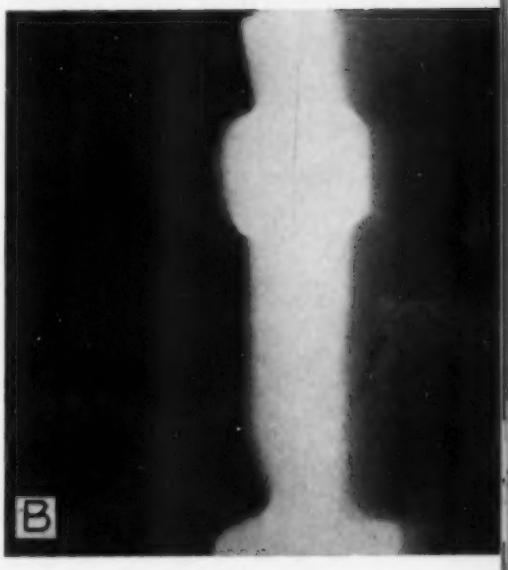


Fig. 9—(A) Radiograph showing localized, spongy-type shrinkage in the upper portion of a 3 in. thick section of an inadequately fed, degassed No. 195 alloy. (B) Radiograph showing surface shrinkage of a filet in an inadequately fed No. 195 alloy casting (see Fig. 10). ×1.

formation of an eutectic, the surface set is relatively small, and there is no highly localized shrinkage. On the other hand, the entire top of the wedge is characterized by a considerable amount of spongy unsoundness. This porosity results largely from poor feeding of this portion of the casting. It is obvious that a difference of opinion can arise at this point. Some observers might call this defect "shrinkage of a general spongy type," the preferred description; others might call it microshrinkage, microporosity, etc, especially if the same defect was in a less obvious place in a commercial casting. This defect at the top of the wedge of composition *C* is well illustrated on the machined surfaces (Fig. 3), and is also well indicated by the dark portion of the radiograph (Fig. 5, composition *C*). Because the sections radiographed were 1 in. thick, the sensitivity is rather low, with the result that the unsoundness in the top of the wedge of composition *C* does not appear in

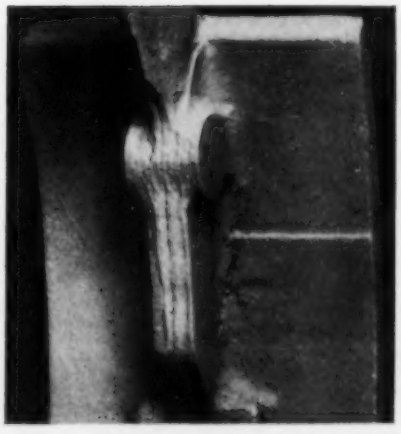


Fig. 10—Photograph of the same casting illustrated by radiograph (B) of Fig. 9. ×1.

a high degree of contrast on the radiograph.

The important point to observe so far is that by pouring a similar casting from four different alloys a marked difference in the location, size, amount, and distribution of the shrinkage has been produced. While these illustrations are of aluminum-base materials, similar effects could be shown for magnesium-base compositions. However, most commercial magnesium-base materials are of the same type as the aluminum-base compositions *A* and *C*.

Microporosity

Figure 16, which will be discussed more fully in its proper place, shows photomicrographs at a magnification of ×8 and ×50, at a location adjacent to and immediately below the lowest point of the surface set illustrated on the wedge of composition *C* of Fig. 3. Reference to these photomicrographs will show that the spongy shrinkage areas grade off into fine, angular porosity, a defect which will be termed microporosity.

The defects illustrated by Figs. 2, 3, 4, and 5 are all shrinkage caused by inadequate feeding of the top of the wedge, but these defects may manifest themselves as a smooth, rounded cavity, as shown by composition *A* of Fig. 2, or a widely distributed, spongy shrinkage zone ac-

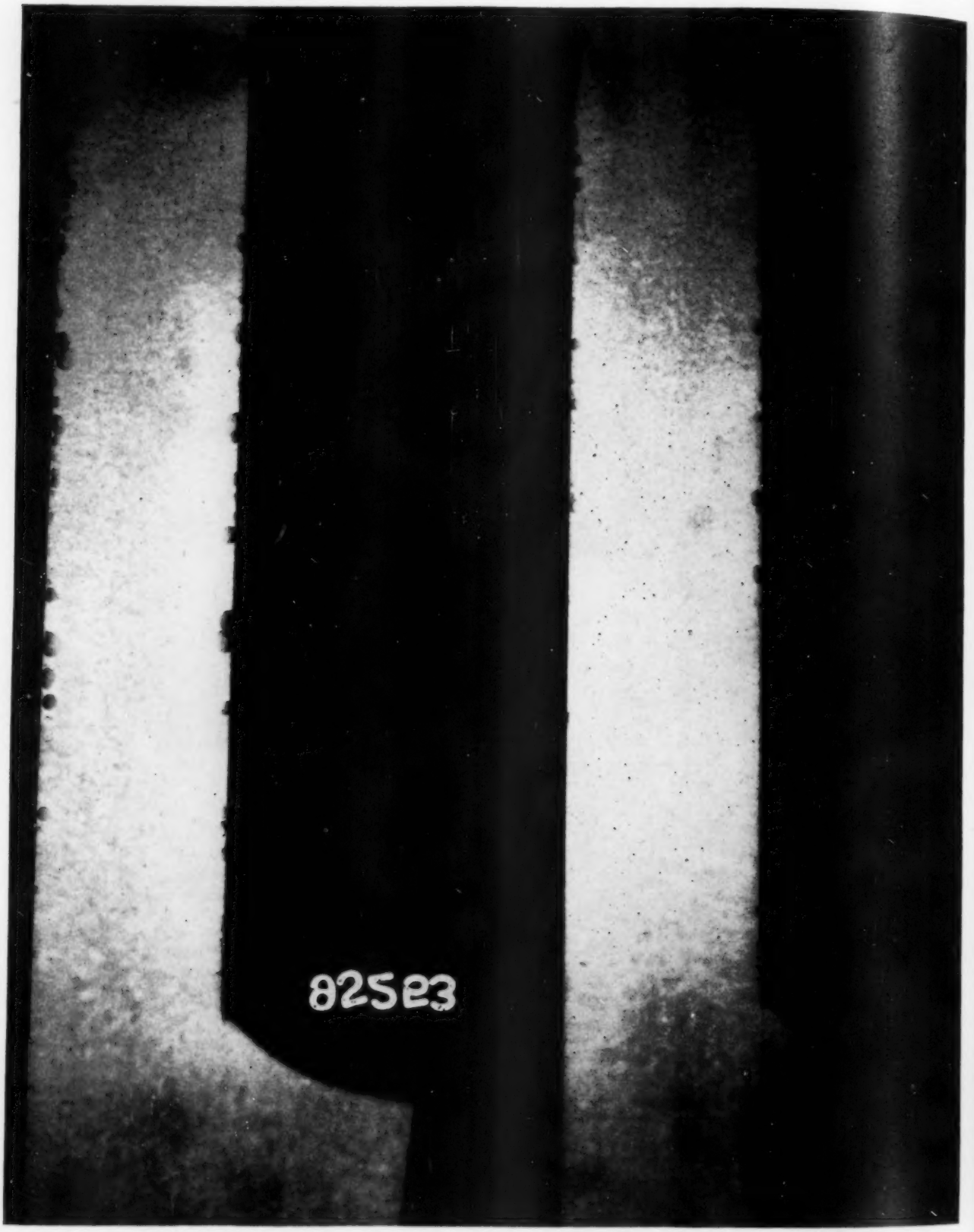


Fig. 11—Radiograph of a cast panel of commercially pure magnesium. The coarse grain is revealed by the mottling on the radiograph. While a number of very small blows formed by a reaction of the melt with portions of the mold are shown at the surface, there is no microporosity because this composition solidifies at substantially constant temperature. This radiograph should be compared with those shown by Figs. 12, 13, 14, and 15, which illustrate the effect of composition on the susceptibility of magnesium alloys to the formation of microporosity.

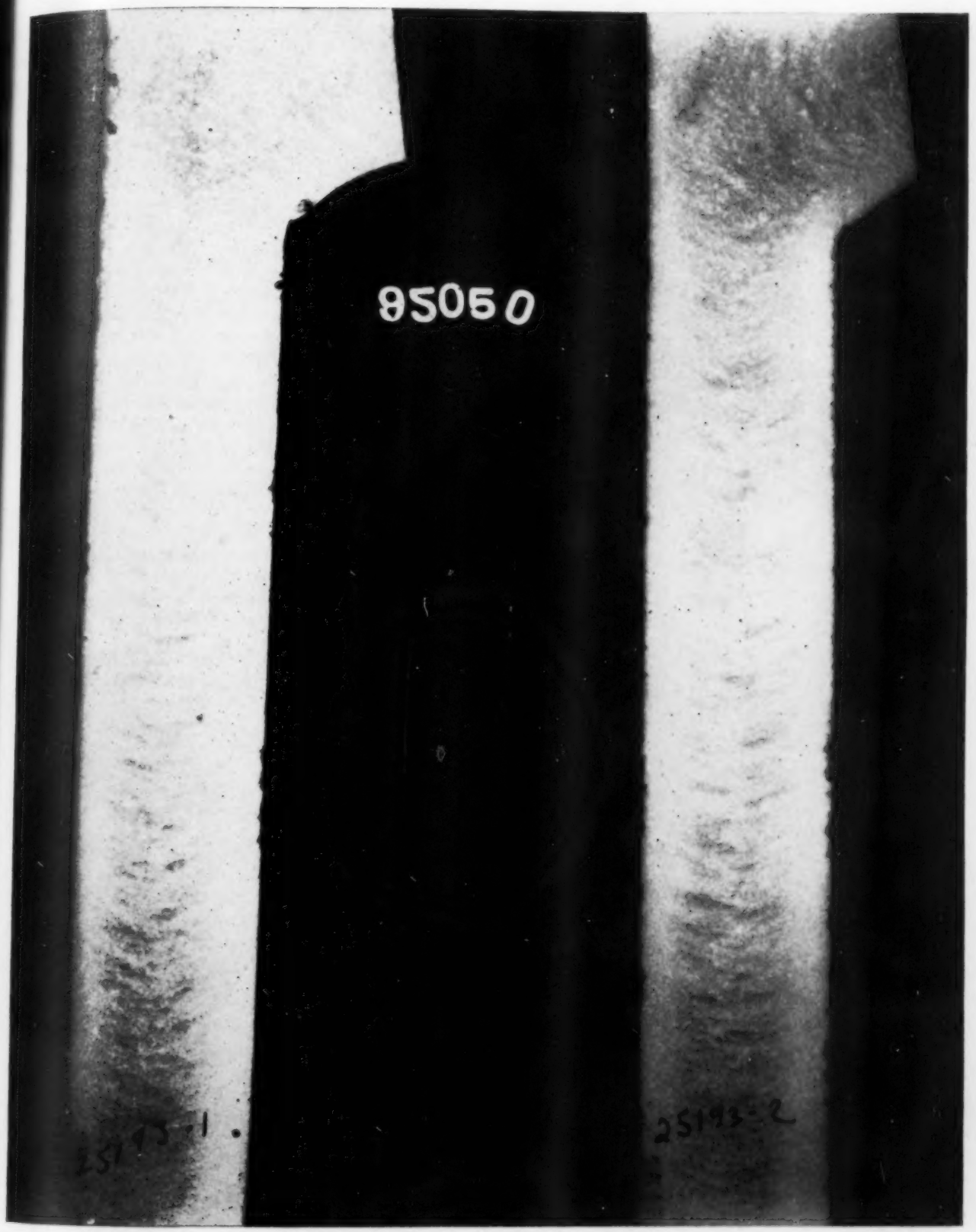


Fig. 12—This casting is entirely similar to the one shown by Fig. 11, except that 3 per cent zinc has been added to the melt. This alloy solidifies over a range of about 290 F without the formation of any eutectic. As a consequence, it is quite susceptible to the formation of microporosity, and this defect is clearly illustrated on the radiograph. This figure should be compared with Figs. 11, 13, 14, and 15.

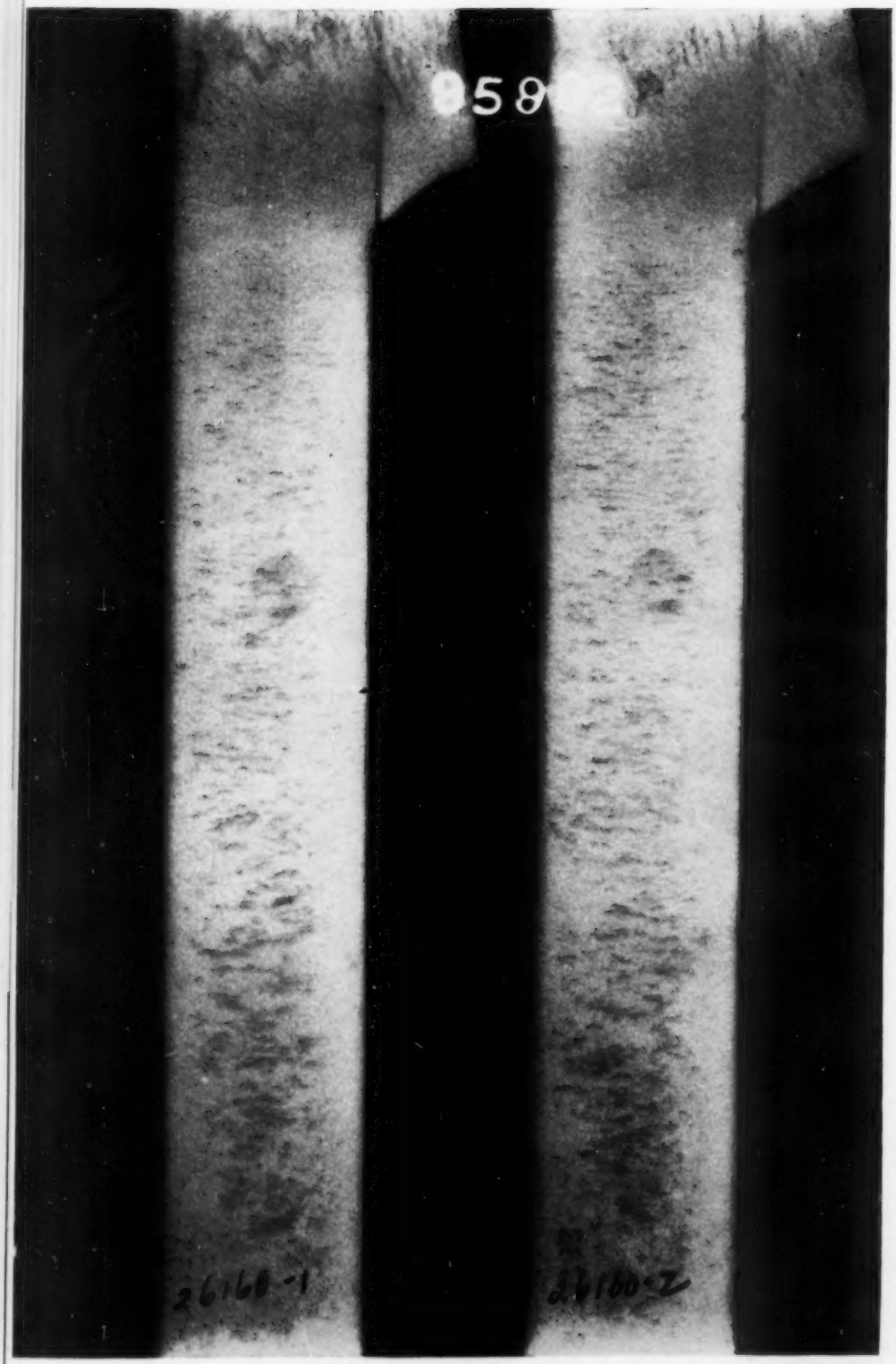


Fig. 13—Radiograph of the same casting shown by Figs. 11 and 12, but the composition is magnesium with 6 per cent aluminum and 3 per cent zinc. This alloy solidifies over a range of approximately 510 F with the formation of a very slight amount of eutectic. As compared with the 3 per cent zinc alloy, shown by Fig. 12, this alloy solidifies over a wider range but slightly more eutectic is formed. The net effect is that this alloy, corresponding to ASTM-AZ-63 is probably slightly more susceptible to the formation of microporosity than the alloy shown by Fig. 12.

companied or associated with microporosity, as illustrated by composition *C*, Figs. 3 and 5.

All of these melts were degassed but, since perfect degassing is not possible, the role of gas in the formation of the microporosity, illustrated by Fig. 16, is not precisely known. To make certain that an erroneous impression as to the cause of this defect is not implied, it is better to call this defect microporosity rather than microshrinkage. This gradation of localized shrinkage into microporosity will be considered in greater detail in its proper place.

Effect of Gas Content. It has been noted that when a melt solidifies a contraction in volume occurs. Unless properly fed this contraction will result in shrinkage. However, if a gas is evolved simultaneously with the contraction less localized shrinkage will occur. Instead, the unsoundness will occur more generally distributed in the form of pinhole porosity in eutectic-containing alloys, and in the form of microporosity in solid solution type alloys.

Gas Reduction

The first of these is illustrated by Fig. 6, which shows machined wedges of a copper-silicon alloy cast in green sand. Casting *B* was poured before fluxing with chlorine, and casting *A* was poured after fluxing. The chlorine flux has reduced the gas content of the melt, with the result that the gas evolution did not entirely compensate for the shrinkage. The area of localized shrinkage near the top center of the wedge is therefore apparent. However, in casting *B* the large quantity of pinhole porosity has prevented the formation of localized shrinkage. The appearance of this localized shrinkage and of the pinhole porosity is further illustrated by the radiograph (Fig. 7) of the same specimen.

The important point for the foundrymen is that aluminum castings poured from degassed melts require better feeding to avoid shrinkage than castings poured from gassy melts. More generally stated, castings which are to be free of pinhole porosity also require better feeding than castings which normally contain large amounts of pinhole porosity, if shrinkage defects are also to be avoided.

Illustrations of Typical Shrinkage Defects in Commercial Castings. On the basis of the various types of shrinkage which have already been observed in the different types of alloys, it is to be expected that the same variations will occur in commercial castings of alloys. Figure 8 illustrates radiographs of shrinkage defects in two different alloys. Figure 8*A* shows shrinkage in No. 220 alloy, which is aluminum containing approximately 10 per cent magnesium.

The shrinkage resulted from inadequate feeding of a 4-in. section. It is obviously a generally distributed or spongy type of shrinkage, and it would be expected that the shrinkage would be similar to that which occurs in the aluminum-4 per cent magnesium because both alloys are solid solution type. Only the No. 220 composition produces a small amount of eutectic. In fact, the unsoundness revealed by Fig. 8*A* is quite similar to that in composition *C* of Fig. 5.

Shrinkage

A different form of more concentrated localized shrinkage is illustrated by Fig. 8*B* showing localized shrinkage in a ½-in. section of No. 355 alloy. This alloy contains a nominal 1.3 per cent copper, 5.0 per cent silicon, and 0.5 per cent magnesium.

This composition solidifies over a considerable range of temperature, but a large amount of eutectic solidifies at the end. It is known definitely that this is localized shrinkage and that it results from poor feeding in the casting. The exact explanation for the rather smooth-walled annular shape of the cavities is not known precisely.

Figure 9 illustrates two other types of shrinkage defects in aluminum castings. Both of these are radiographs, and Fig. 9*A* is of a 3-in. section of an inadequately fed No. 195 alloy, which is aluminum containing approximately 4.5 per cent copper and 0.6 to 0.8 per cent each of iron and silicon.

Alloy No. 195 solidifies over a considerable range in temperature, with only a small amount of eutectic at the end. It is to be expected, therefore, that the shrinkage defect would be less localized than in a pure metal or 100 per cent eutectic,



Fig. 14—This radiograph is of the same casting as shown by the preceding three photographs, except that the alloy contains 9 per cent aluminum and 2 per cent zinc, corresponding to ASTM-AZ92 alloy. This composition solidifies over a range of approximately 380 F with the formation of several per cent of eutectic. As compared with the alloy shown by Fig. 13, the range in solidification of the alloy shown here is much less and more eutectic is formed. As a result, this composition is much less susceptible to the formation of microporosity. As compared with the 3 per cent zinc alloy, shown by Fig. 12, the range in solidification is slightly more, but a great deal more eutectic is formed. The latter factor greatly reduces the susceptibility of this composition to the formation of microporosity.

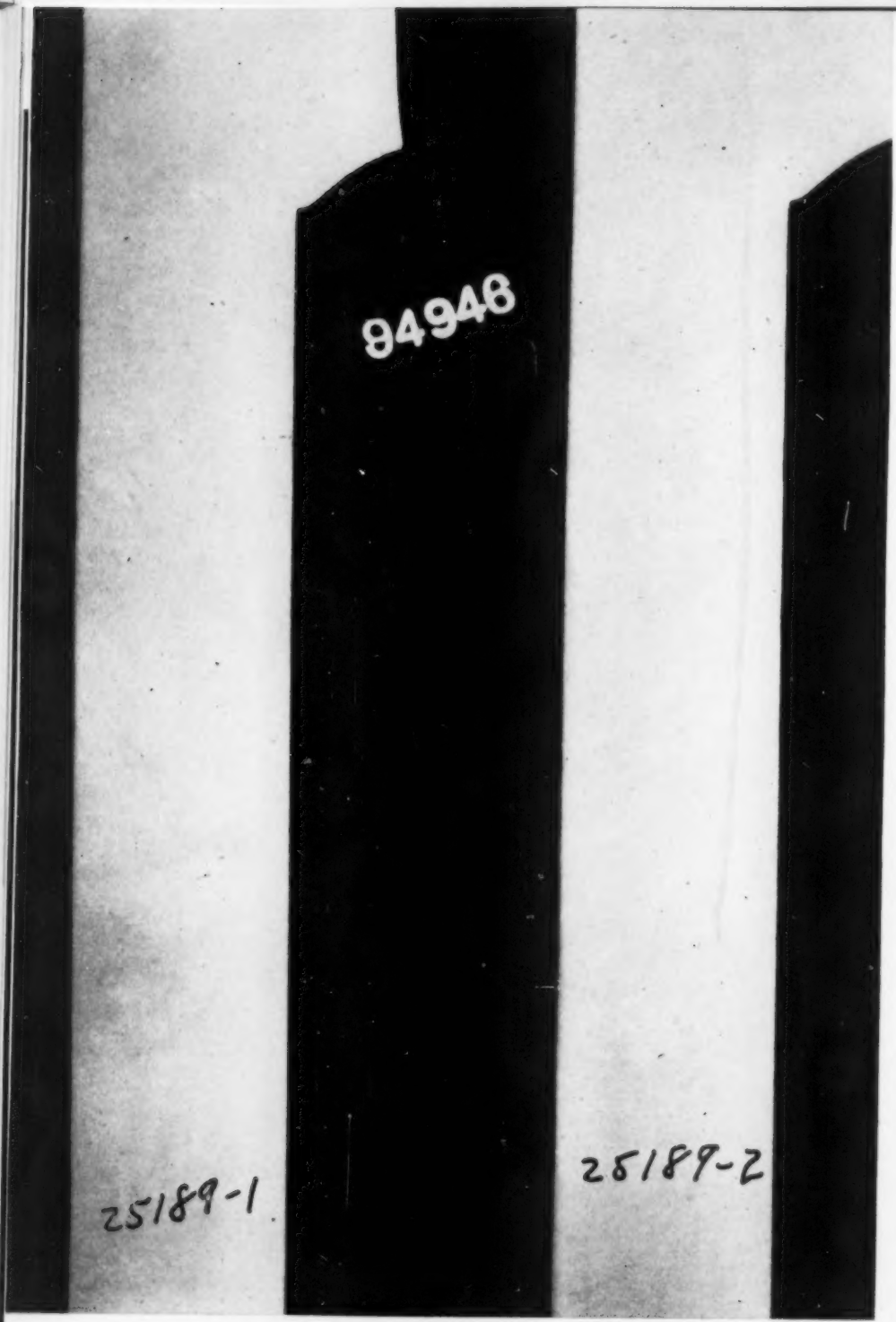


Fig. 15—Radiograph is similar to the preceding four, except that the alloy contains 8 per cent aluminum. This composition solidifies over a range of 230 F with the formation of slightly more eutectic than the 6 per cent aluminum, 3 per cent zinc alloy shown by Fig. 13, but slightly less eutectic than the 9 per cent aluminum, 2 per cent zinc alloy shown by Fig. 14, all other factors the same. As a result, this composition is far less susceptible to the formation of microporosity than the 6 per cent aluminum, 3 per cent zinc alloy shown by Fig. 13, but its susceptibility is comparable to that of the 9 per cent aluminum, 2 per cent zinc alloy shown by Fig. 14. These particular radiographs indicate that the 9 per cent aluminum alloy forms less microporosity than the 9 per cent aluminum, 2 per cent zinc alloy shown by Fig. 14. Other information indicates, however, that the two alloys are comparable and, if anything, the 9 per cent aluminum, 2 per cent zinc alloy is less susceptible to microporosity than the 8 per cent aluminum alloy because the greater amount of eutectic more than offsets its longer solidification range.

but not so widely distributed as in a solid solution type alloy represented by composition *C* in Fig. 3.

Figure 9*B* shows a typical, fairly localized surface shrinkage in a fillet of a No. 195 alloy casting of approximately the same composition as that of Fig. 9*A*. The surface appearance of this localized shrinkage in the fillet is illustrated by Fig. 10.

It is probable that the somewhat directional solidification in the casting has tended to make this defect quite precisely localized, as delineated by the radiograph. It should be emphasized here that, all other factors being the same, rapid directional solidification will tend to produce a sharper more localized shrinkage cavity where feeding is inadequate.

Gradation of Shrinkage into Microporosity. Microporosity is a defect characterized by small, more or less interconnected voids that are now well known to most light-alloy foundrymen. The effect of alloy composition, gas content of the melt, and degree of feeding provided, on the amount, size, shape, and distribution of microporosity in magnesium alloys has been described in separate papers [2, 3, 4, 5].

Metal Affected

It is sufficient to point out here then that either aluminum- or magnesium-base alloys of the solid solution type, such as that represented by composition *C* of Fig. 3, are the most susceptible to this type of defect. On the other hand, a melt which solidifies at constant temperature, such as pure metal, almost never has even a slight trace of this defect. Likewise, alloys which solidify over a range in temperature, but which contain a large amount of eutectic, seldom contain this defect, although it may occur to a slight extent.

These principles are also illustrated by Figs. 11, 12, 13, 14, and 15, with accompanying captions. These show the effect of composition on the range in solidification and alloy type which in turn determine the susceptibility of magnesium and its alloys to the formation of microporosity.

With either aluminum or magnesium alloys, the point to visualize is that as the amount of liquid solidifying at constant temperature de-



Fig. 16—(A) Photograph of a section taken from the No. 214 alloy wedge illustrated by Fig. 3, Composition C, in the zone adjacent to and just below the surface set. This photograph shows this spongy shrinkage at the top grading into microporosity below. ×8. (B) Lower portion of the specimen illustrated by (A) but at a higher magnification to show the microporosity. ×50.



creases from 100 per cent in pure metals or in 100 per cent eutectics to 0 per cent in solid solution alloys, with an accompanying general increase in solidification range, the tendency to form microporosity increases. With this change there is also an increasing tendency to form the spongy, widely distributed shrinkage illustrated by composition *C*, Figs. 3 and 5.

The gradation of localized shrinkage of the spongy type into microporosity has been noted and is illustrated by Fig. 16. This is further illustrated in Figs. 17 and 20. Figure 20, composition *C*, is the same as composition *C* of Figs. 3 and 5, but this wedge was poured from a gassed melt into a green sand mold.

Figure 17*A* at a magnification of ×8 shows the spongy shrinkage in the top of the wedge of composition *C*, Fig. 20, grading off into microporosity below the shrinkage. The photomicrograph shown in Fig. 17*B* is of the fine porosity in the lower portion of Fig. 17*A* at a higher magnification, ×50, to reveal the microporosity more clearly.

Gassed Metal

Comparing Fig. 16 with Fig. 17, it is obvious that the gassed melt has produced a greater volume of microporosity, and the accentuation of this defect by the higher gas content is entirely evident. This principle, applicable to aluminum-base alloys, has also been fully established by data on magnesium-base alloys as described in other A.F.A. papers [2, 3, 4, 6, 7].

Gradation of shrinkage into microporosity in magnesium alloys is illustrated by Fig. 18, which shows a vertical longitudinal section of an inadequately fed 6x10x1½-in. plate of a magnesium alloy containing 9.4 per cent aluminum, 0.4 per cent zinc, and 0.25 per cent manganese, and degassed by chlorine fluxing.

Figure 18*A* shows a radiograph of a vertical longitudinal section of the plate with macroscopic shrinkage at the top and a heavy, coarse microporosity below. This gradation is better illustrated by the photomicrograph (Fig. 18*B*) at ×12, showing the localized spongy shrinkage at the top, grading into microporosity below.

While one might arbitrarily and, with some basis of logic, sharply de-

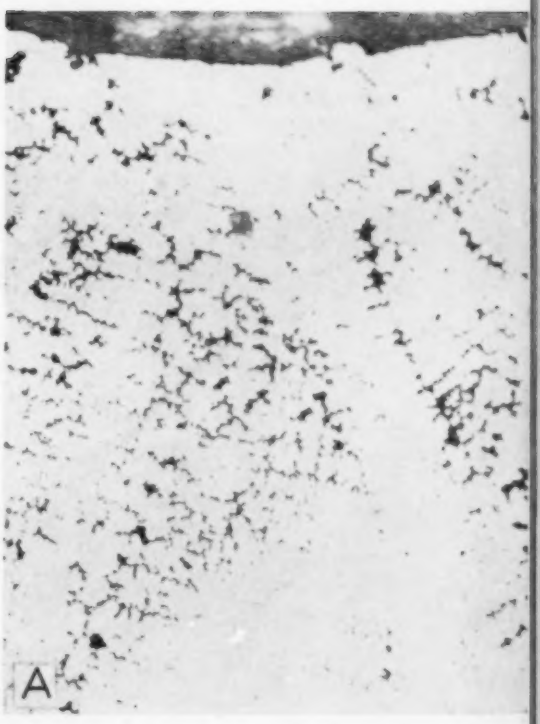
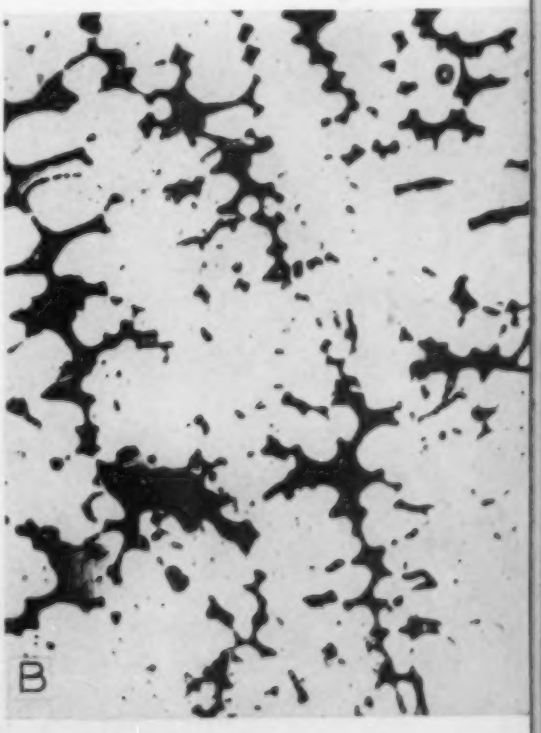


Fig. 17—These photographs are comparable to the preceding ones, except that these specimens were cut from the wedge illustrated by Fig. 20, Composition C. This also is No. 214 alloy, but in this instance the casting was poured from gassed metal, whereas the preceding one, illustrated by Fig. 16, was poured from degassed metal. (A) ×8. (B) ×50.



fine where localized shrinkage ends and microporosity begins in these particular illustrations, it is quite evident that it would be possible in commercial castings to have unsoundness which could be labeled either "shrinkage of a spongy type" or microporosity. Any satisfactory system of nomenclature must recognize the borderline defects.

Gas porosity, it will be recalled, results from the evolution of gas which was in solution in the melt. It is not uncommon for foundrymen and metallurgists to imply that gases from the green sand or cores were trapped in the metal, resulting in pinhole porosity. This is technically incorrect.

While the gas which is precipitated may have been dissolved from the vapors evolved from the core or green sand, the melt first dissolves the hydrogen from this vapor, especially from the moisture, and then reprecipitates it from the melt during solidification, forming the pinholes. Trapped mold gases are not common, but when they do occur, they form a quite different type of defect.

Furthermore, it is quite common to attribute generally distributed gas (pinhole) porosity to shrinkage merely because the pores are angular. This also is incorrect for reasons described elsewhere[1].

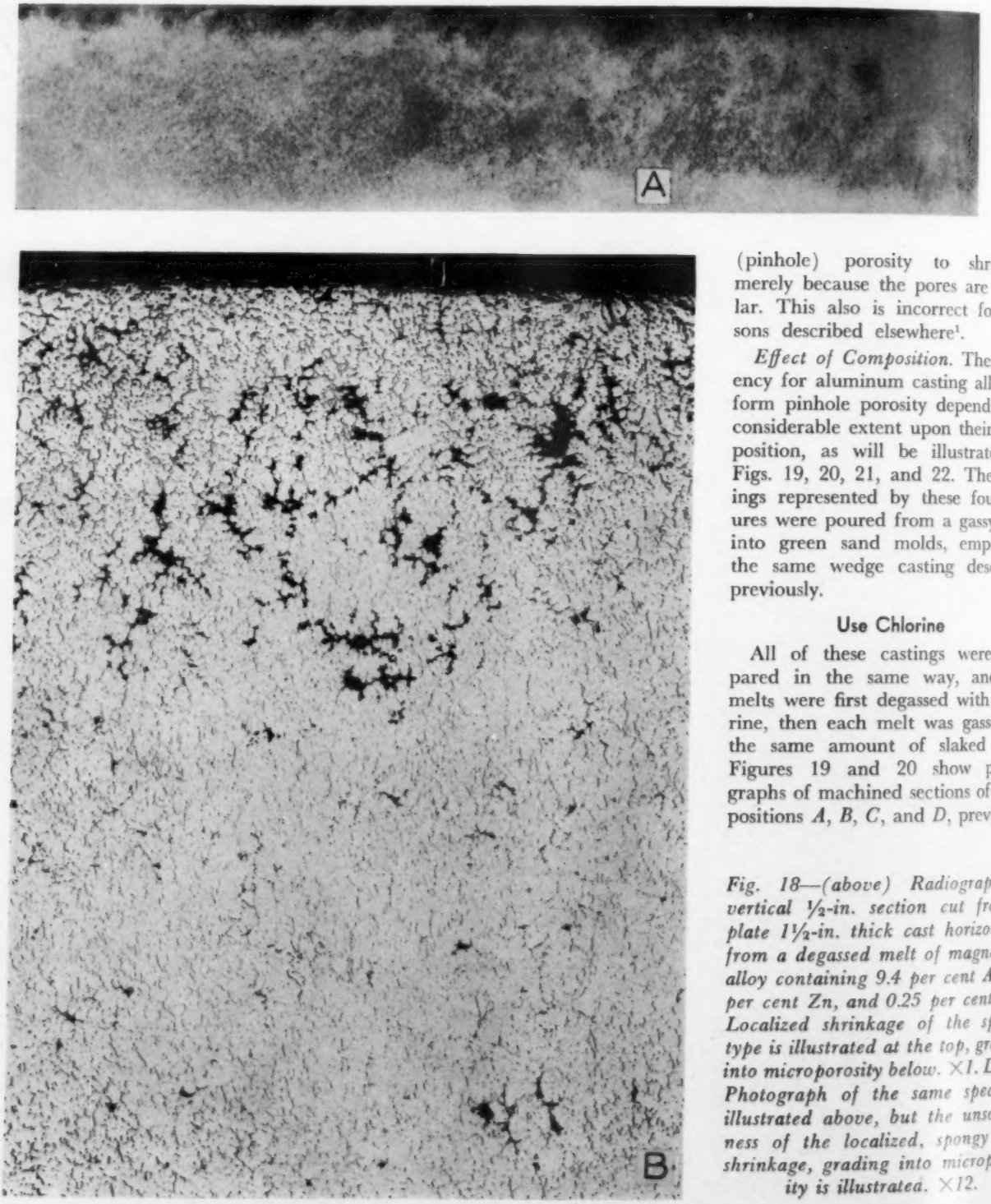


Fig. 18—(above) Radiograph of vertical ½-in. section cut from a plate 1½-in. thick cast horizontally from a degassed melt of magnesium alloy containing 9.4 per cent Al, 0.4 per cent Zn, and 0.25 per cent Mn. Localized shrinkage of the spongy type is illustrated at the top, grading into microporosity below. ×1. Left—Photograph of the same specimen illustrated above, but the unsoundness of the localized, spongy type shrinkage, grading into microporosity is illustrated. ×12.

Effect of Composition. The tendency for aluminum casting alloys to form pinhole porosity depends to a considerable extent upon their composition, as will be illustrated by Figs. 19, 20, 21, and 22. The castings represented by these four figures were poured from a gassy melt into green sand molds, employing the same wedge casting described previously.

Use Chlorine

All of these castings were prepared in the same way, and the melts were first degassed with chlorine, then each melt was gassed by the same amount of slaked lime. Figures 19 and 20 show photographs of machined sections of compositions *A*, *B*, *C*, and *D*, previously

described, while Figs. 21 and 22 illustrate radiographs of 1-in. sections of the same specimens.

The various amounts of pinhole porosity, illustrated by these figures, are a typical result of difference in alloy composition. The conclusions which may be drawn from these four figures, and from comparison of these four with the similar set represented by Figs. 2, 3, 4, and 5, are:

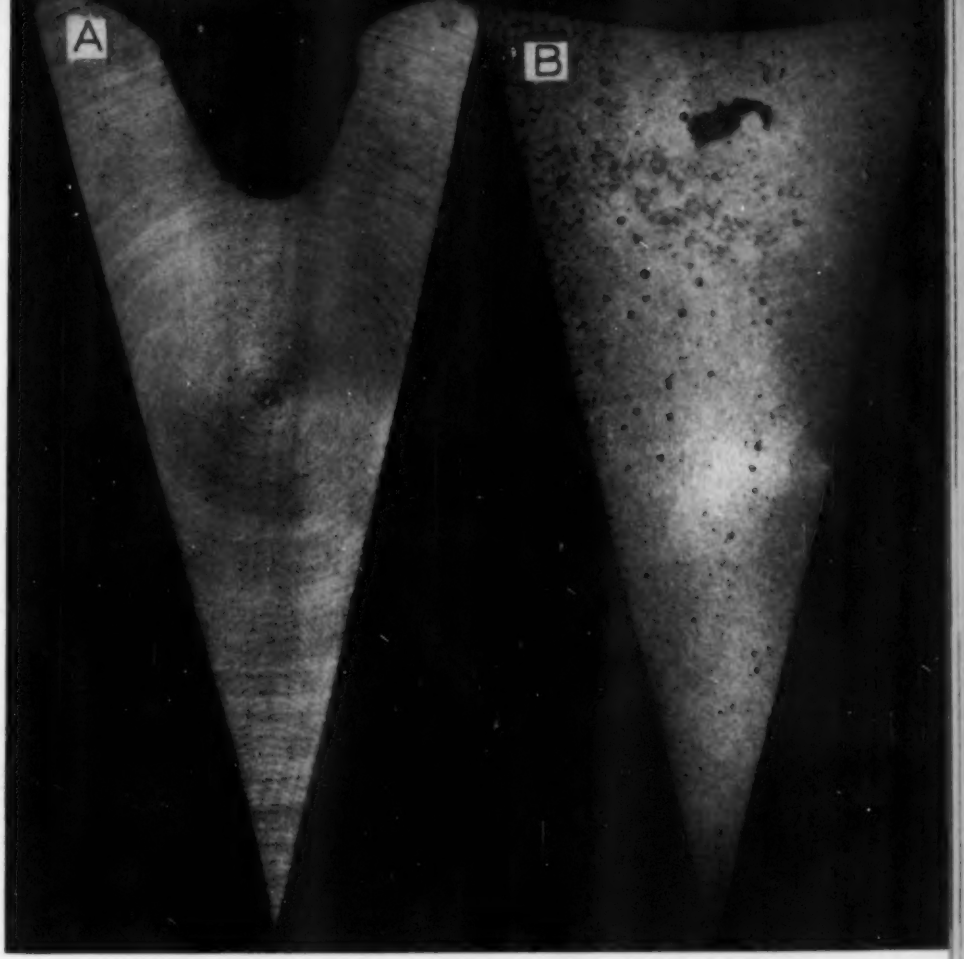
1. It is difficult to entrap hydrogen in the form of pinholes in commercially pure aluminum. Most of the hydrogen will escape from the top surface of the solidifying casting, particularly where the solidification is quite directional from the bottom upwards as it necessarily was in this wedge.

2. The mottling shown on the radiograph of composition *A*, Fig. 21, should not be confused with unsoundness. This mottling is the direct result of the grain structure of the casting. The columnar grains illustrated on the radiograph are quite typical for this composition.

3. Some voids caused by hydrogen in the form of large, widely separated pinholes are illustrated by composition *B*, Fig. 19. It will be recalled that composition *B* is nearly, but not quite, a 100 per cent eutectic alloy. In any event, this alloy solidifies over a fairly narrow range in temperature, resulting in some entrapment of hydrogen in the form of pinholes.

4. Composition *D*, Fig. 20, represents a casting which solidifies over a considerable range but with the formation of a large amount of eutectic. This considerable range in solidification temperature facilitates the entrapment of bubbles of hydrogen, forming a larger number of these pinholes, visible on the machined surface, than was entrapped by composition *B*, Fig. 19.

5. The number of pinholes represented by entrapped hydrogen bubbles, shown by composition *B* and *D*, can best be estimated by the ap-



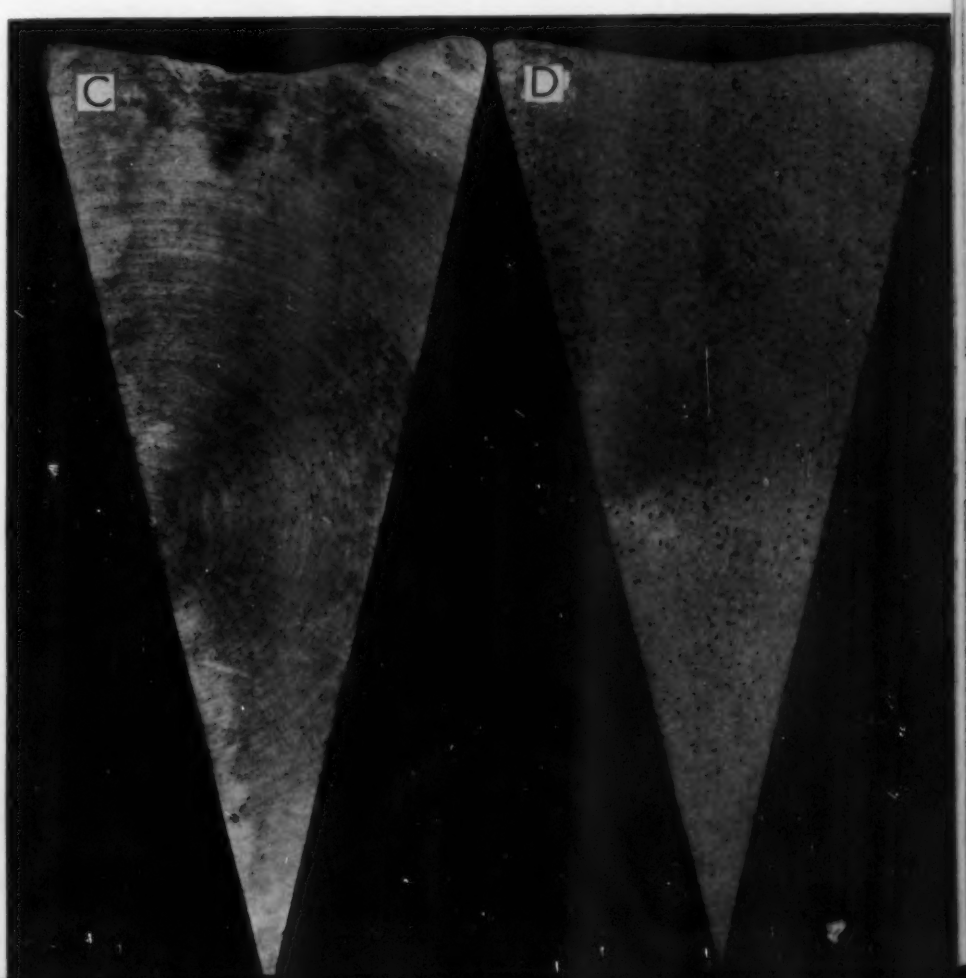
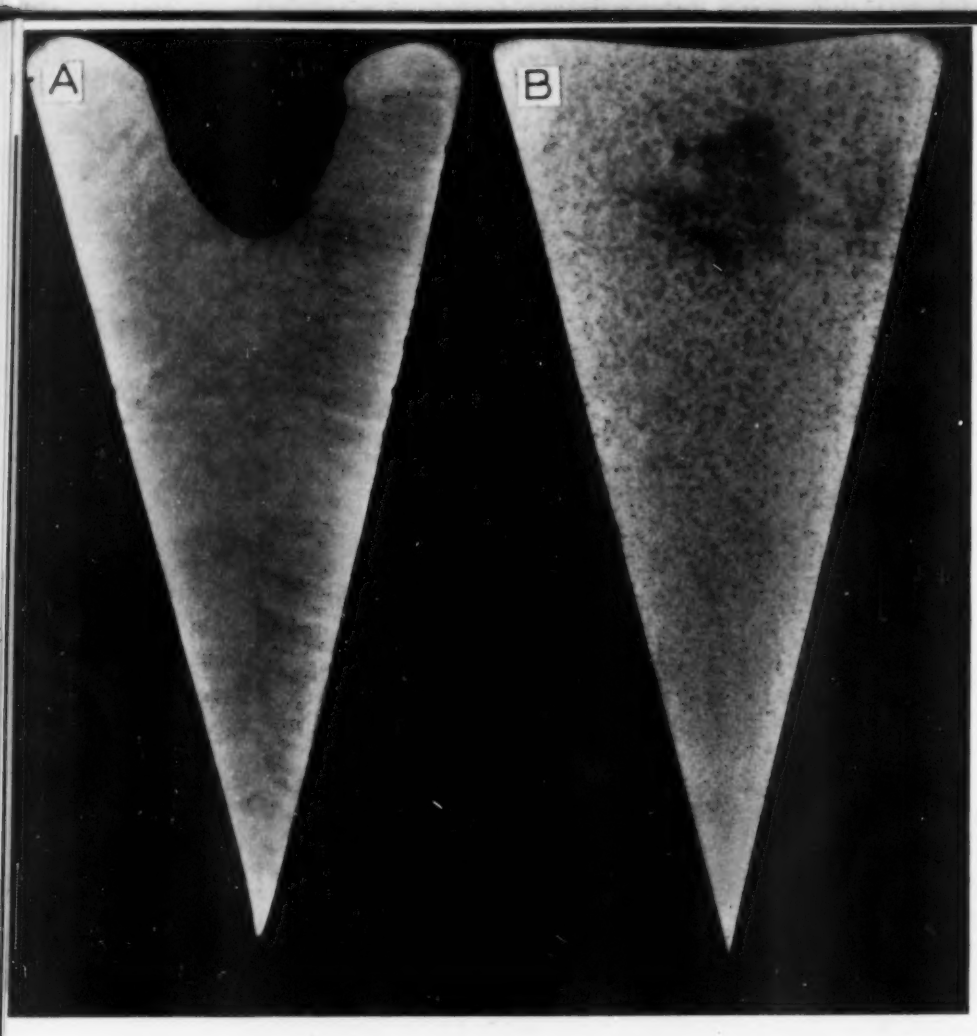
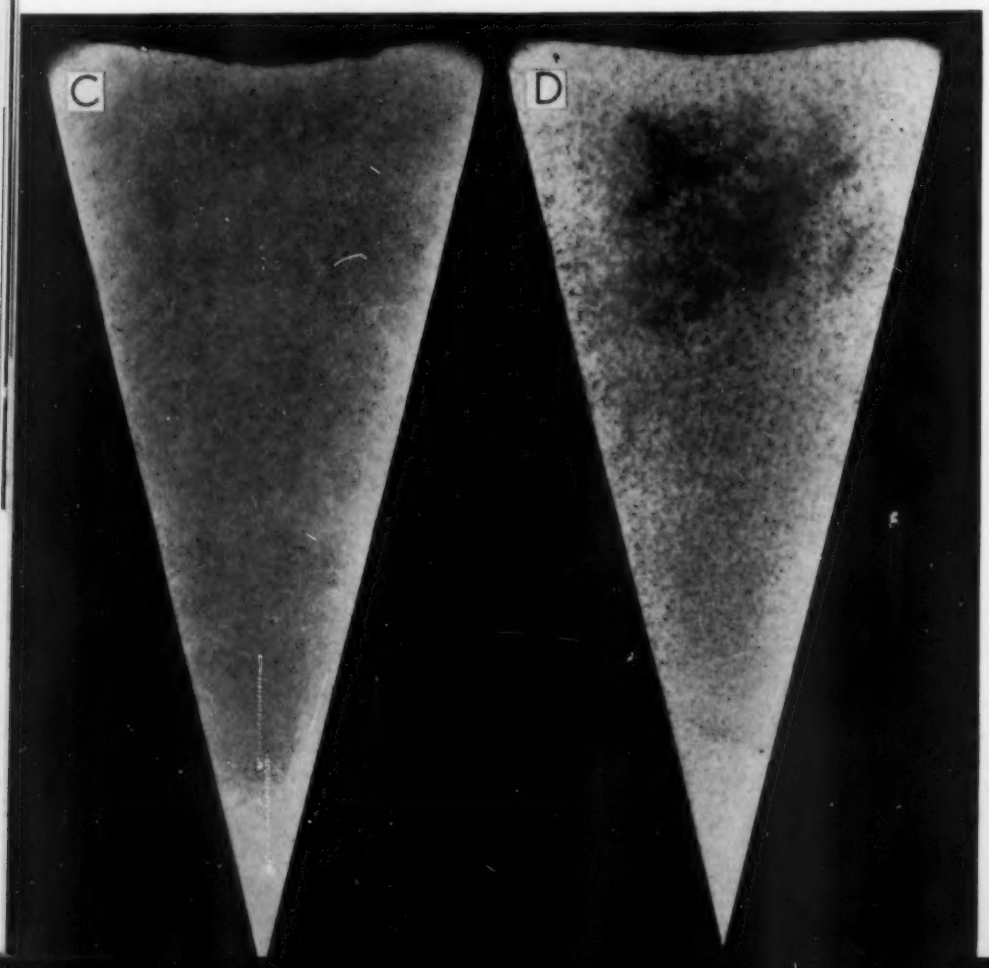


Fig. 19—(right above) Machined sections similar to Fig. 2, except that gassed metal was poured in green sand. ×5/6.

Fig. 20—(right) Machined sections similar to Fig. 3, except that gassed metal was poured in green sand. ×5/6.





pearance of the machined surfaces shown by Figs. 19 and 20. The radiographs of these alloys, shown by Figs. 21 and 22, are of 1-in. sections in which a large number of pinholes are shown on the film, even in the case of composition *B* which contains the least number.

6. Composition *C*, which is a solid solution type, does not readily entrap hydrogen bubbles to form pinholes. However, it should be noted that it is entirely possible for an alloy of this type to contain a considerable quantity of pinholes, but it is much less likely to do so than alloys such as those represented by composition *D*. Instead, the solid solution type alloy represented by composition *C* contains microporosity which is greatly accentuated by the high gas content of the melt. This has been illustrated previously by Figs. 16 and 17, and described more fully elsewhere [2, 3, 4, 5].

7. Comparing Figs. 19, 20, 21, and 22 of castings poured from gassy metal with those of Figs. 2, 3, 4, and 5 of castings poured from degassed melts, it is quite evident that there is a marked tendency for the gassy melts to produce less localized shrinkage.

8. The gas porosity tends to be much lighter in the tip of the cone where solidification is rapid than at the large end where solidification is much slower. This is a typical effect of solidification rate. The radiograph of composition *D*, Fig. 22, shows a typical, high concentration of voids near the top center. The machined surface of this specimen contains no indication of localized shrinkage near the top. It is not possible to determine whether the high concentration of voids shown by the radiograph, Fig. 22, composition *D*, is caused by a high concentration of large pinholes or whether there is some localized shrinkage. It is probably a combination of both because

Fig. 21—(left above) Radiographs of specimens shown by Fig. 19 which are similar to Fig. 4, except that gassed metal was poured in green sand. ×5/6.

Fig. 22 — (left) Radiographs of specimens shown by Fig. 20 which are similar to Fig. 5, except that gassed metal was poured in green sand.

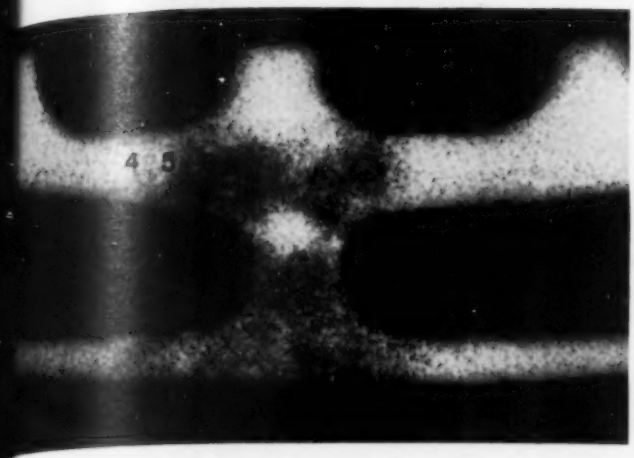


Fig. 23—Radiograph showing localized shrinkage of a somewhat spongy type with pinhole porosity in an aluminum-copper-silicon alloy containing 3 per cent Cu and 7 per cent Si. ×1.

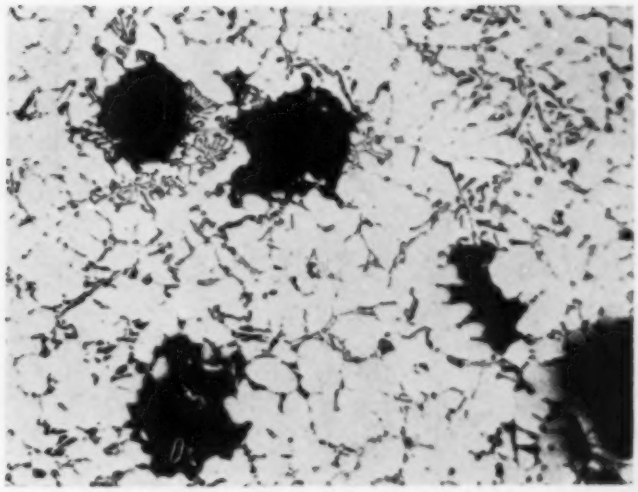


Fig. 24—This photomicrograph shows the unsoundness present at location 4 illustrated on Fig. 23. ×40.

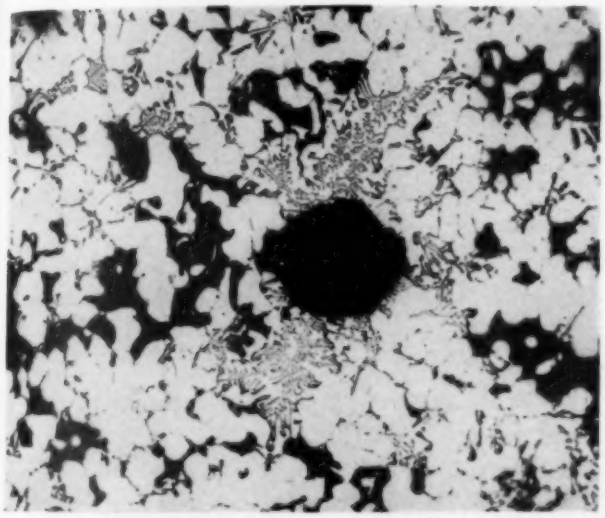


Fig. 25—Photomicrograph showing unsoundness at location 5, illustrated on Fig. 23. ×40.

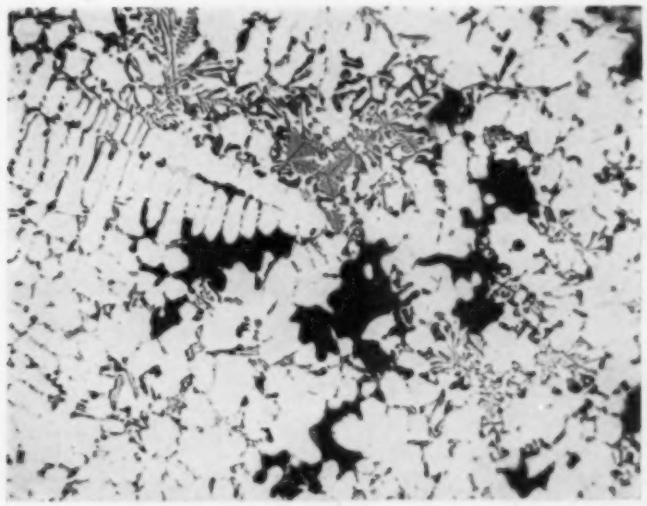
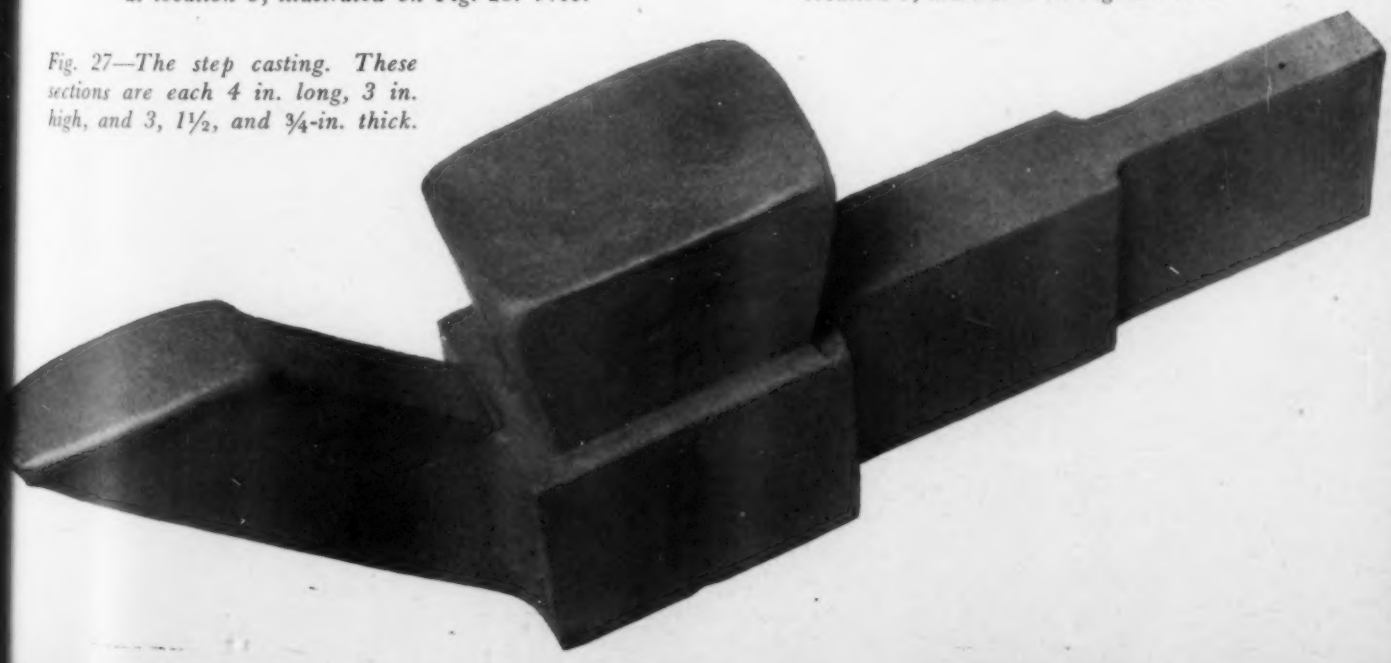


Fig. 26—Photomicrograph showing unsoundess at location 6, illustrated on Fig. 23. ×40.

Fig. 27—The step casting. These sections are each 4 in. long, 3 in. high, and 3, 1½, and ¾-in. thick.



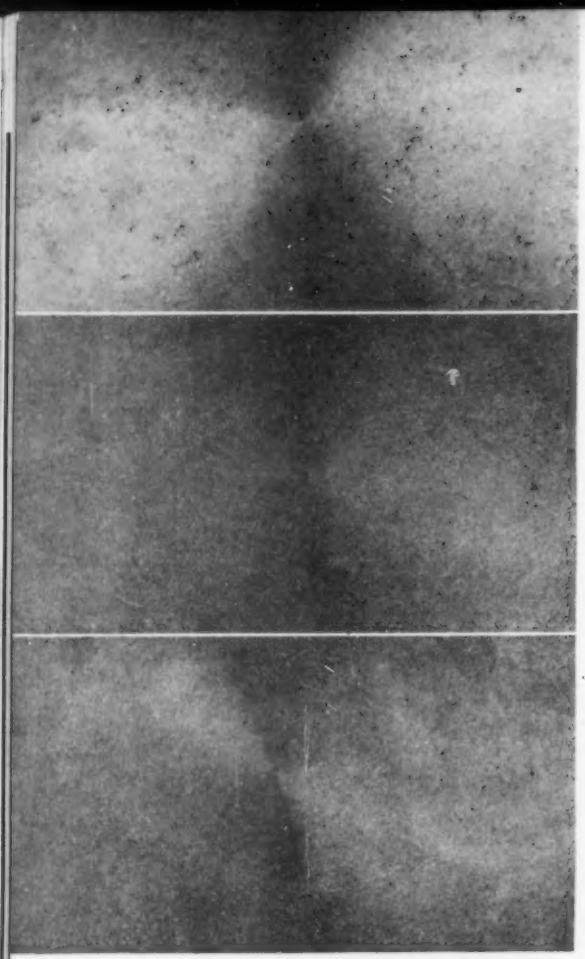


Figure 28

Photograph of machined mid-sections cut from the step casting (Fig. 27) poured in dry sand mold, using degassed No. 122 alloy containing 10 per cent Cu, 1.2 per cent Fe, and 0.2 per cent Mg; 3-in. section top, 1½-in. section center, ¾-in. section bottom. ×1.

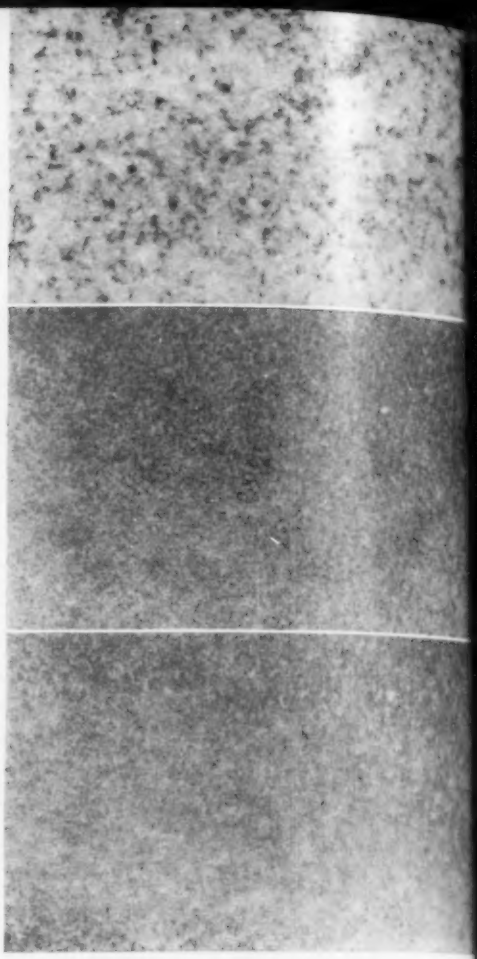


Figure 29

Radiograph of approximately 3/16-in. sections cut from the same casting shown in Fig. 28. ×1.

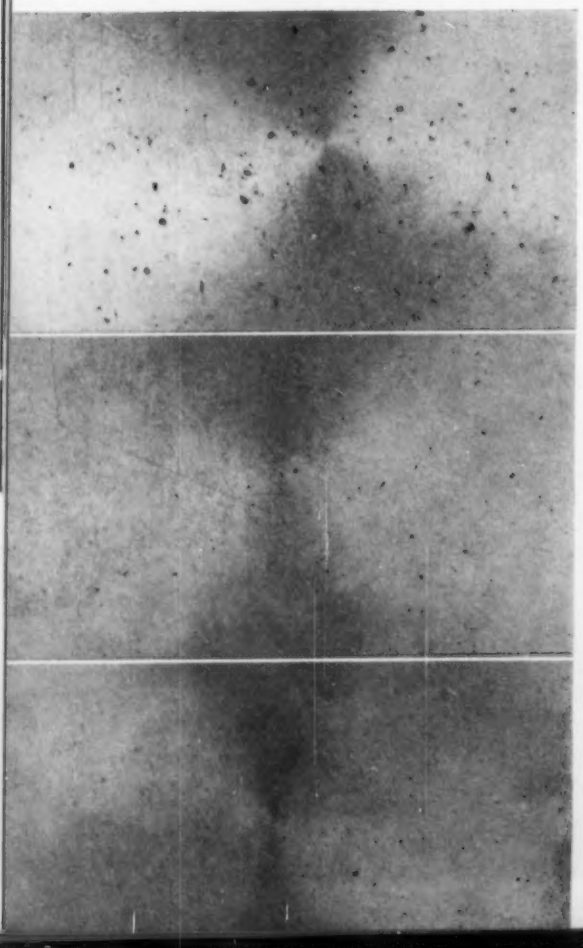


Figure 30

Similar to Fig. 28, except that the metal contains a medium amount of gas. ×1.



Figure 31

Radiograph of sections shown by Fig. 30. ×1.

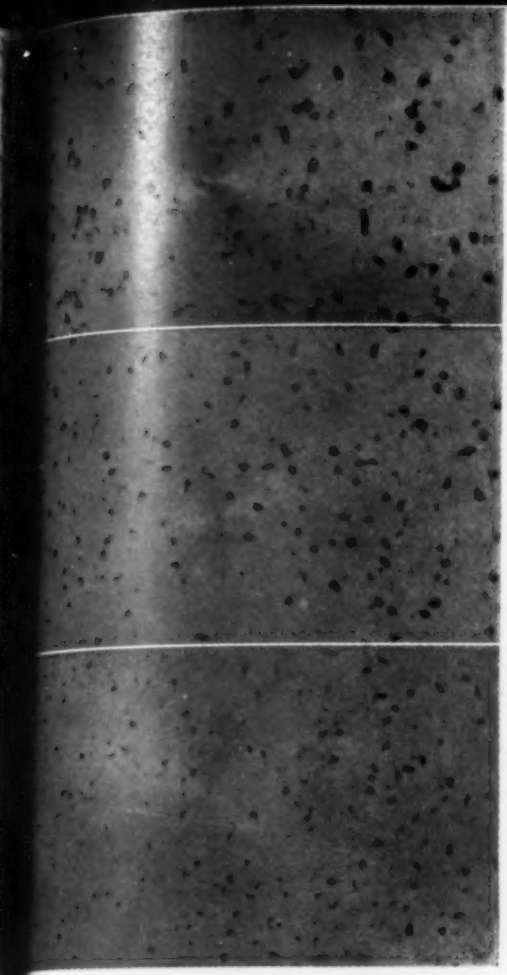


Figure 32

Same as Fig. 28, except that the castings were poured from a melt containing a large amount of gas. ×1.

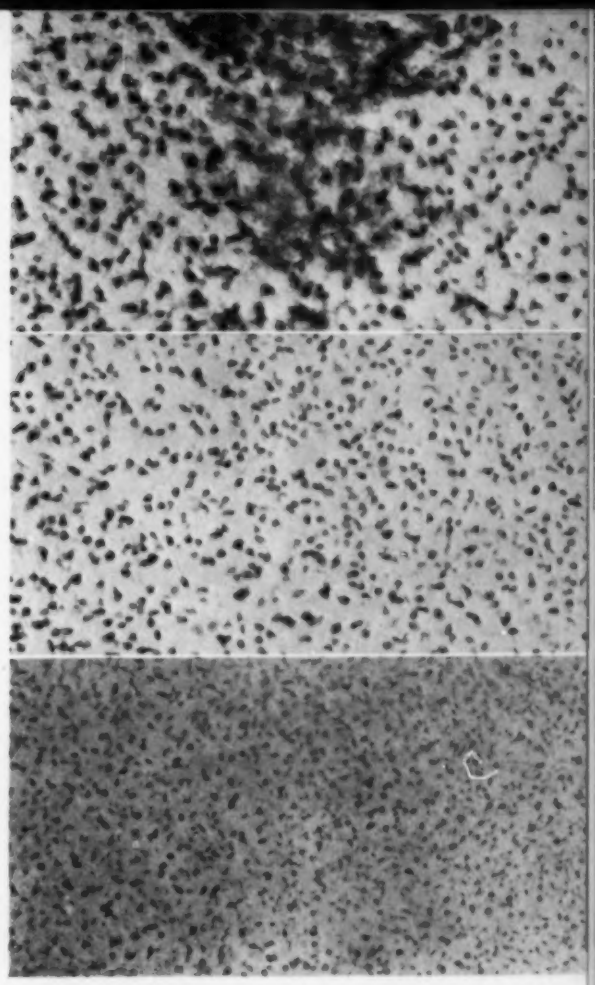


Figure 33

Radiograph of sections shown by Fig. 32. ×1.

the concentration of localized shrinkage and the concentration of pinholes would be expected to be a maximum in this location.

A typical occurrence of a severe localized shrinkage of a somewhat general spongy nature combined with pinhole porosity is shown by Figs. 23 to 26, inclusive. Figure 23 is a radiograph of a section of an ironing buck of an aluminum 3 per cent copper-7 per cent silicon alloy.

Figures 24, 25, and 26 are photomicrographs at ×40 of unetched surfaces at locations 4, 5, and 6 shown on the radiograph (Fig. 23). This alloy solidifies over a range in temperature, with a large amount of eutectic at the end. The spongy shrinkage zone is quite localized. At location 4, only rounded to angular pinholes are evident. At location 5, pinholes and shrinkage are present. At location 6, only shrinkage is formed.

There is no question that gas was precipitated in the shrinkage zone, but the shrinkage voids were large enough to contain the evolved gas without the formation of the pinholes. This defective casting resulted from a considerable gas evolution and inadequate feeding. Had no gas been evolved, the voids in the general spongy shrinkage zone probably would have constituted a greater total volume, but they would have been still more localized.

The principle illustrated here, showing the effect of composition upon the susceptibility to the formation of pinholes in aluminum, probably also applies to the susceptibility of magnesium alloys to the formation of gas holes. However, since gas holes do not normally occur, and commercial magnesium compositions are largely confined to types *A* and *C*, composition effects have no great practical significance to the magnesium founder.

Effect of Solidification Rate and Gas Content. The effects of solidification rate or section thickness and gas content of the melt on the amount, size, shape, and distribution of pinhole porosity in aluminum alloys have been described in detail elsewhere[1]. A few illustrations will be shown which will also indicate the correlation of the appearance of the machined surface and the appearance of the unsoundness on an x-ray film.

Figure 27 illustrates the test casting employed. Figures 28 to 33, inclusive, illustrate sections cut from the test casting having steps 4 in. long, 3 in. wide, and 3, 1½, and ¾-in. thick. The alloy is No. 122 containing 10 per cent copper, 1.2 per cent iron, and 0.2 per cent magnesium. All of the molds were made of dry sand to prevent gas absorption in the mold cavity.

Longitudinal vertical sections were cut from the mid-portion of the casting, and Figs. 28, 30, and 32 show machined surfaces of these sections cut from castings poured respectively from a melt of low gas content, medium gas content, and high gas content.

The 3-in. section is shown at the top, the 1½-in. section in the center, and the ½-in. section machined from the ¾-in. portion is shown at the bottom in each of the photographs. Figures 29, 31, and 33 are radiographs of 3/16-in. thick sec-

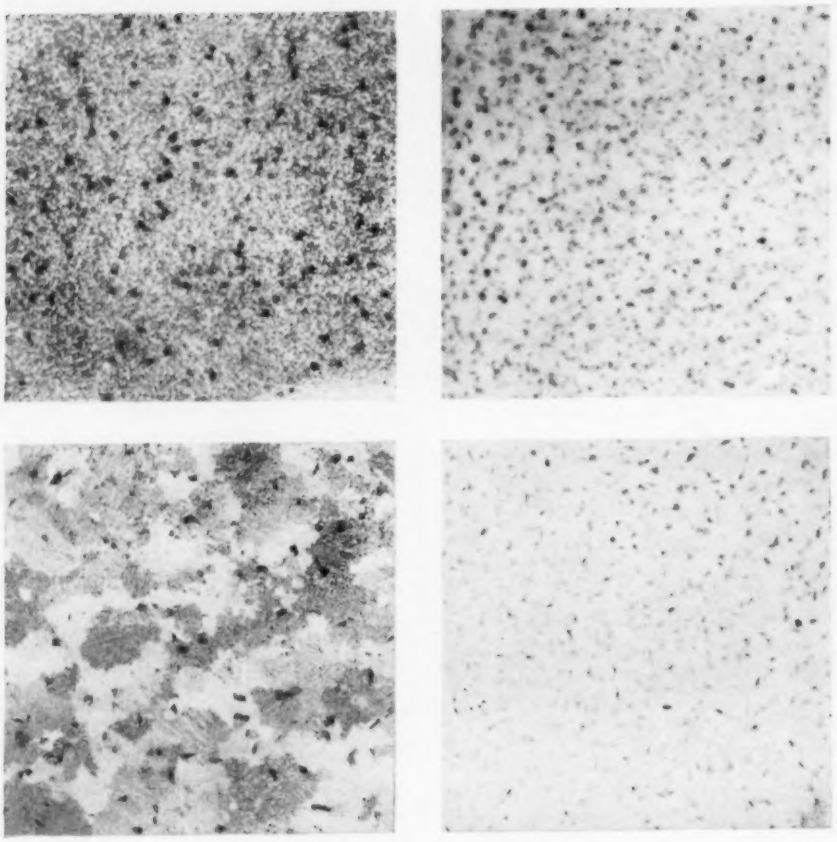


Fig. 34—Photograph at the left and radiograph at the right of polished and etched mid-sections of No. 195 alloy cut from 2½-in. cubes poured in green sand. Metal wax gassed with ammonium chloride. The sections shown by the lower half of the figure were poured at 1450 F, after which grain refinement was obtained by adding 0.10 per cent Ti. The castings illustrated by the upper portion of the figure were then poured at 1450 F. This alloy contains a nominal 4.5 per cent Cu with 0.6 to 0.8 per cent each Fe and Si. ×1.

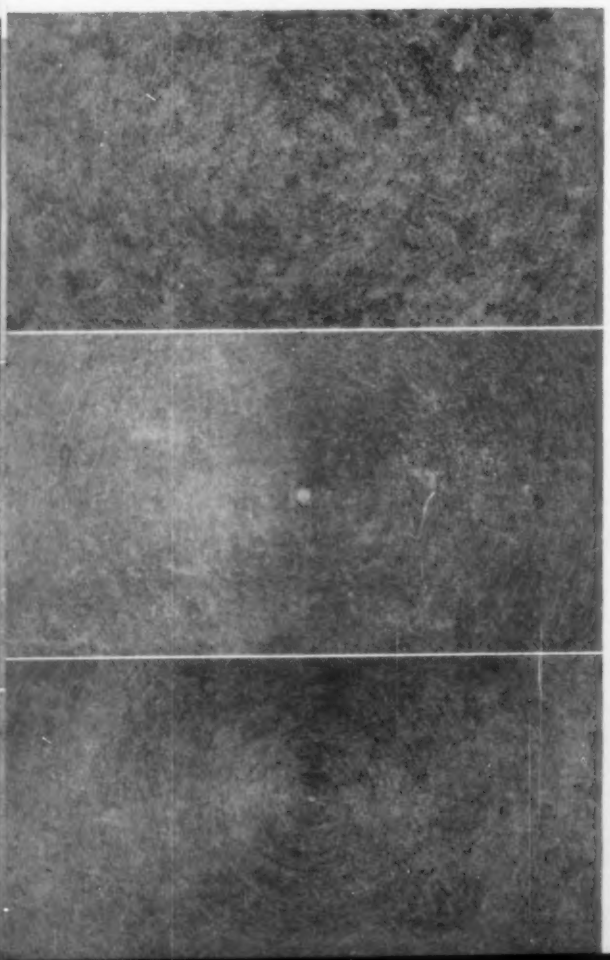


Figure 35

Machined mid-section cut from the step casting (Fig. 27). The melt was poured in green sand molds having an A.F.A. permeability of 35 and 5 per cent water. The metal was substantially free from gas as the result of the chlorine fluxing operation applied. Top—3-in. section. Center—1½-in. section. Bottom—¾-in. section. The alloy is No. 355 containing 1.3 per cent Cu, 5 per cent Si, and 0.5 per cent Mg. Compare with Figs. 36, 37 and 38. ×1.

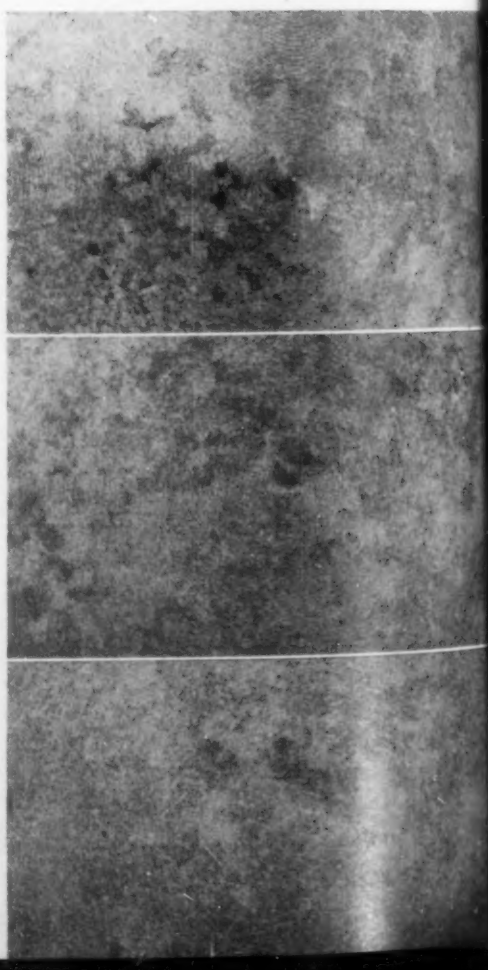


Figure 36

Same as Fig. 35, except that the sections machined were near the cast surfaces of the steps. Top—3-in. section. Center—1½-in. section. Bottom—¾-in. section. ×1.

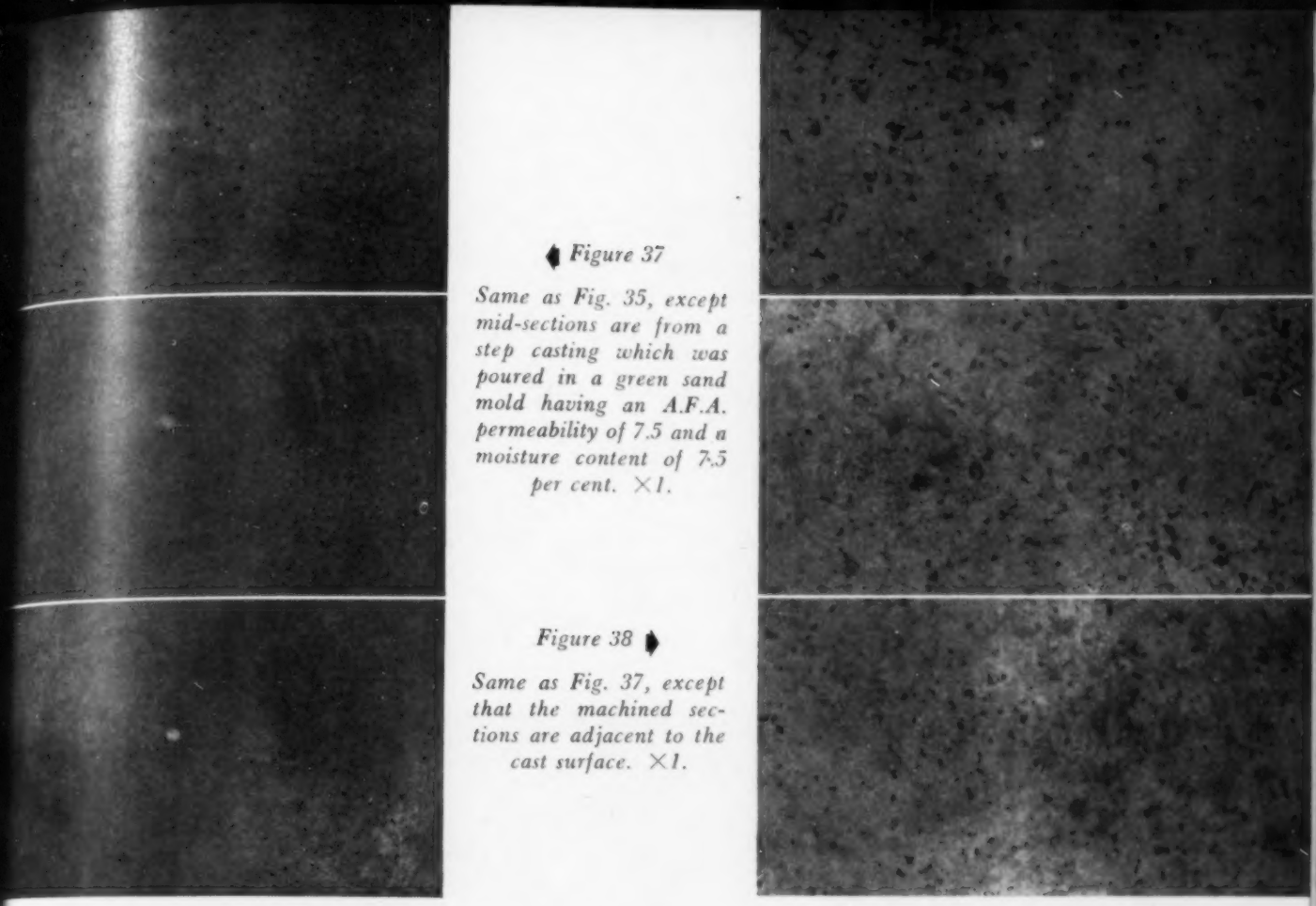


Figure 37

Same as Fig. 35, except mid-sections are from a step casting which was poured in a green sand mold having an A.F.A. permeability of 7.5 and a moisture content of 7.5 per cent. ×1.

Figure 38

Same as Fig. 37, except that the machined sections are adjacent to the cast surface. ×1.

tions of the same castings illustrated by photographs, Figs. 28, 30, and 32, of the machined surfaces.

On the basis of these six figures which show typical effects of gas content and section thickness on the occurrence of pinhole porosity, the following conclusions may be drawn:

1. As the section thickness increases, all other factors being the same, the total volume, size, and angularity of the pinholes increase.

2. Total volume of the pinholes, their size and number increase with increasing gas contents of the melt.

3. The machined surface shows only the voids on a plane; whereas the radiograph presumably shows all or most of the voids in a section of substantial thickness, in this case about 3/16-in. Accordingly, the severity of the porosity revealed on the radiograph is far greater than that revealed on the machined surface.

4. There is some tendency for localized shrinkage of a general, spongy nature to occur in the 3-in. section, which in this case was not fed particularly well. This spongy localized shrinkage is closely associated with, and not entirely distinguishable from, the pinhole porosity.

These principles probably also apply to gas holes in magnesium, but for reasons indicated previously they would have little practical significance for the magnesium founder.

Effect of Grain Size. In general, all other factors being the same, the pinhole porosity in aluminum-base alloys is rounder and smaller, the finer the grain. This is illustrated by Fig. 34 which shows a photograph at the left, and a radiograph at the right of polished sections of No. 195 alloy cut from 2½-in. cubes poured in green sand. These sections have been given a grain-size etch. This alloy contains a nominal 4.5 per cent copper with approximately 0.6 to 0.8 per cent each iron and silicon.

The melt was gassed with ammonium chloride and the castings were poured at 1450 F. The casting at the bottom shows the grain size and porosity obtained before grain refinement, and the two photographs

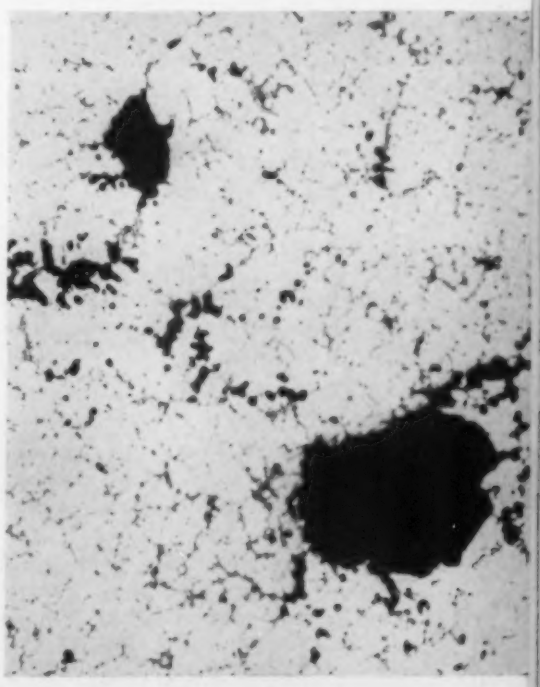


Fig. 39—Photomicrograph showing pinhole porosity grading into microporosity. This aluminum alloy contains 6 per cent Mg, 1 per cent Mn, and 1.5 per cent Cu. ×50.

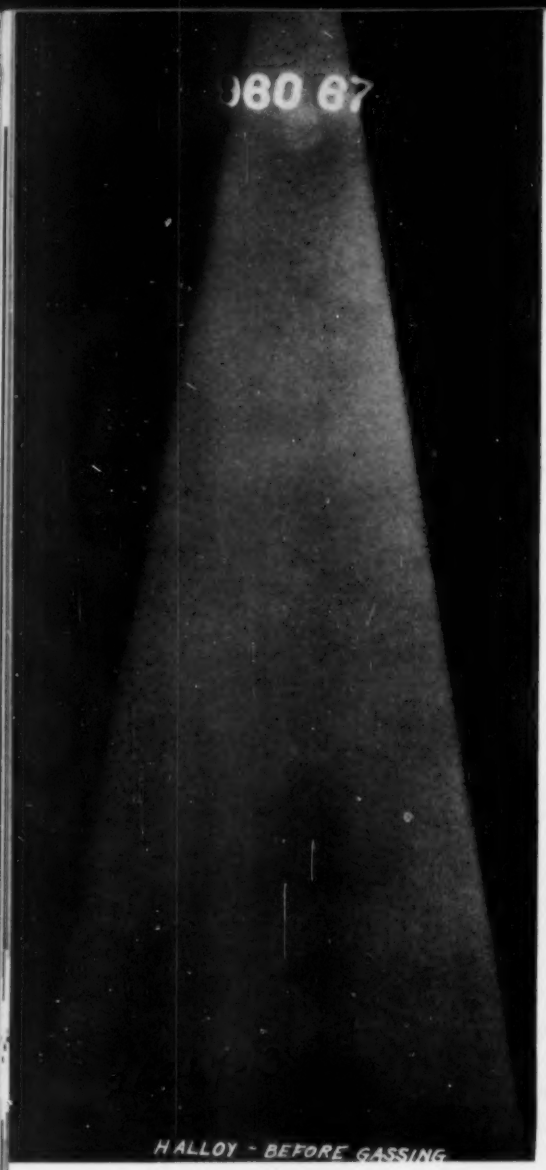
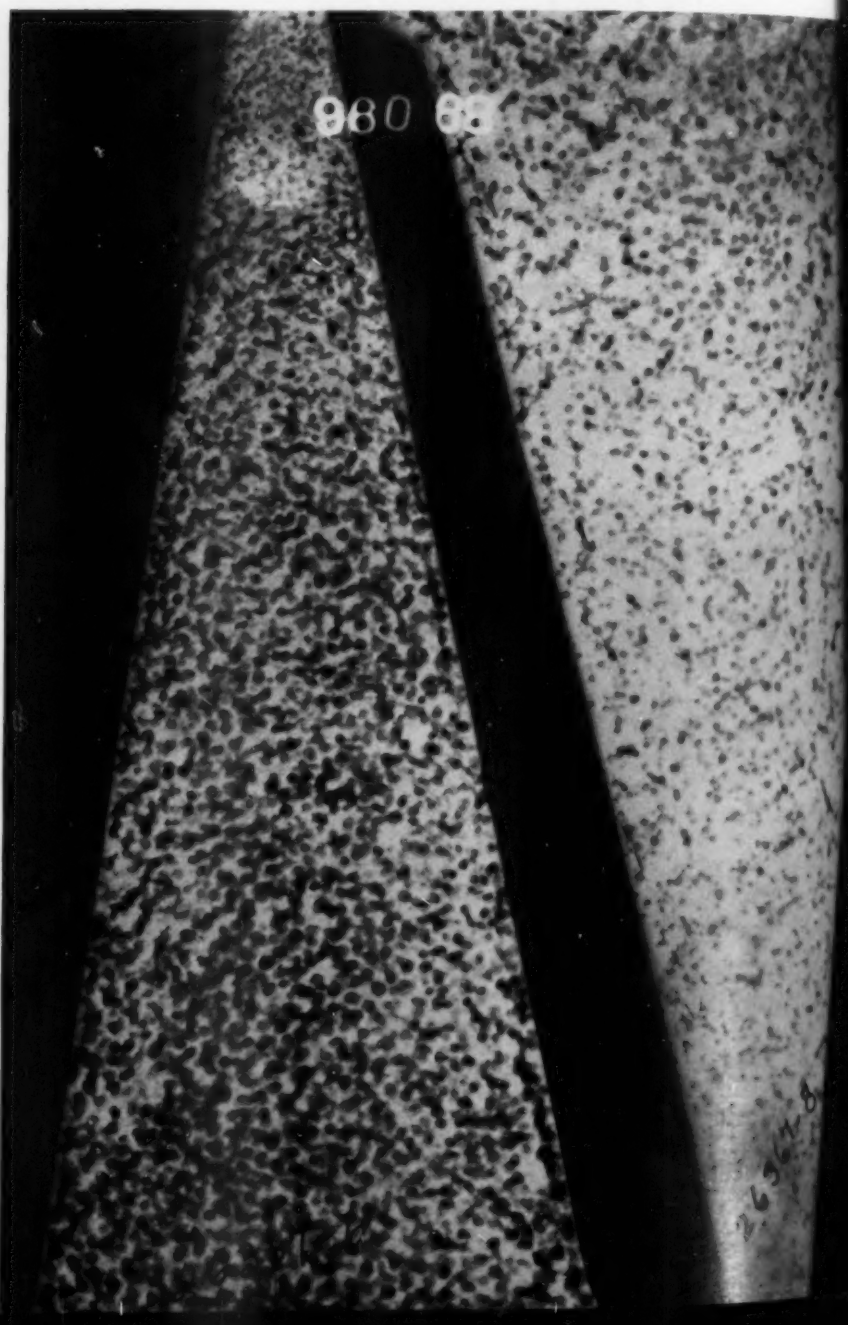


Fig. 40 (left)—Photograph of a triangular panel of ASTM-AZ63 alloy degassed with 0.02 lb of chlorine per lb of metal and poured from 1400 F. This radiograph shows the complete absence of microporosity when the panel is poured from a degassed melt.

Fig. 41 (below)—These two radiographs show the same panel illustrated by Fig. 40, but poured immediately after gassing the melt with hydrogen for 15 min at 1550 F. The other panel shows the same casting after the melt was held at 1400 F for 15 min after the hydrogen gas treatment. These radiographs clearly show the rounded gas holes which, probably because the section is lighter and solidification more rapid, are substantially smaller than those shown by Fig. 1. Even so, these gas holes are much larger than the corresponding pinholes which occur in some aluminum cast alloys. The radiograph at the right shows that, with decreasing gas content, slit-like cavities form near the edges of the panel.



at the top show the castings after grain refinement was obtained by adding 0.10 per cent titanium. Because the melt was gassed, the pinhole porosity is severe. It will be noted, however, that there are many more pinholes of finer, more equiaxed shape in the fine-grained casting than in the other.

Effect of A.F.A. Permeability and Moisture Content of the Molding Sand. Figures 35 to 38, inclusive, show an interesting effect of mold material upon the occurrence of pinhole porosity in No. 355 alloy castings. No. 355 alloy contains a nominal 1.3 per cent copper, 5 per cent silicon, and 0.5 per cent magnesium. This alloy was prepared, degassed by fluxing with chlorine, and then poured into the step castings described previously and illustrated by Fig. 27.

Figures 35 and 36 show castings which were poured in a natural green sand having an A.F.A. permeability

Fig. 43 (right)—Radiograph of a panel similar to those illustrated by the preceding three radiographs. This radiograph shows that, after the melt has been held for 2 hr at 1400 F, the gas content has decreased to such an extent that it is probably near or below the maximum gas solubility in the solid alloy near the solidus temperature. As a result, only a very slight amount of microporosity is formed, but this panel is less sound than the one illustrated by Fig. 40 which was poured from a chlorinated degassed melt. This and the three preceding radiographs clearly illustrate how, with a decreasing gas content of the melt, the rounded gas holes gradually decrease in number, from slit-like cavities, and then with a still further decrease in gas content merely cause the formation of microporosity. In this respect, these magnesium-base solid solution type alloys are entirely analogous to the aluminum-base solid solution type alloys such as that represented by composition C (Fig. 3).

Fig. 42 (below) — This radiograph is of the same triangular panel illustrated by Fig. 41, but the panels were poured 30 min and 1 hr, respectively, after the hydrogen-gas treatment, the melt being held at 1400 F. At this temperature, some of the hydrogen gradually escapes from the melt, with the result that there is insufficient gas in the liquid melt to cause the gas holes, illustrated by Fig. 41. Instead, the gas in the melt accentuates the penetration of the microporosity (shaded portion).





of 35 and 5.0 per cent moisture. Of these two figures, Fig. 35 represents the appearance of the machined surface of longitudinal vertical sections cut through the center of the castings. The 3-in. section is at the top, the 1½-in. section at the center, and the ¾-in. section at the bottom. On the other hand, Fig. 36 shows the same machined section, but the machined surfaces are adjacent to the cast surface.

Figures 35 and 36 should be compared with Figs. 37 and 38 which are entirely similar, excepting that the natural green sand mold had an A.F.A. permeability of 7.5 and 7.5 per cent moisture. Of these two photographs. Fig. 37 shows machined longitudinal vertical sections cut down the center of the casting. On the other hand, Fig. 38 shows similar machined sections adjacent to the surface of the casting. The explanation of the great concentration of pinholes near the surface is unques-

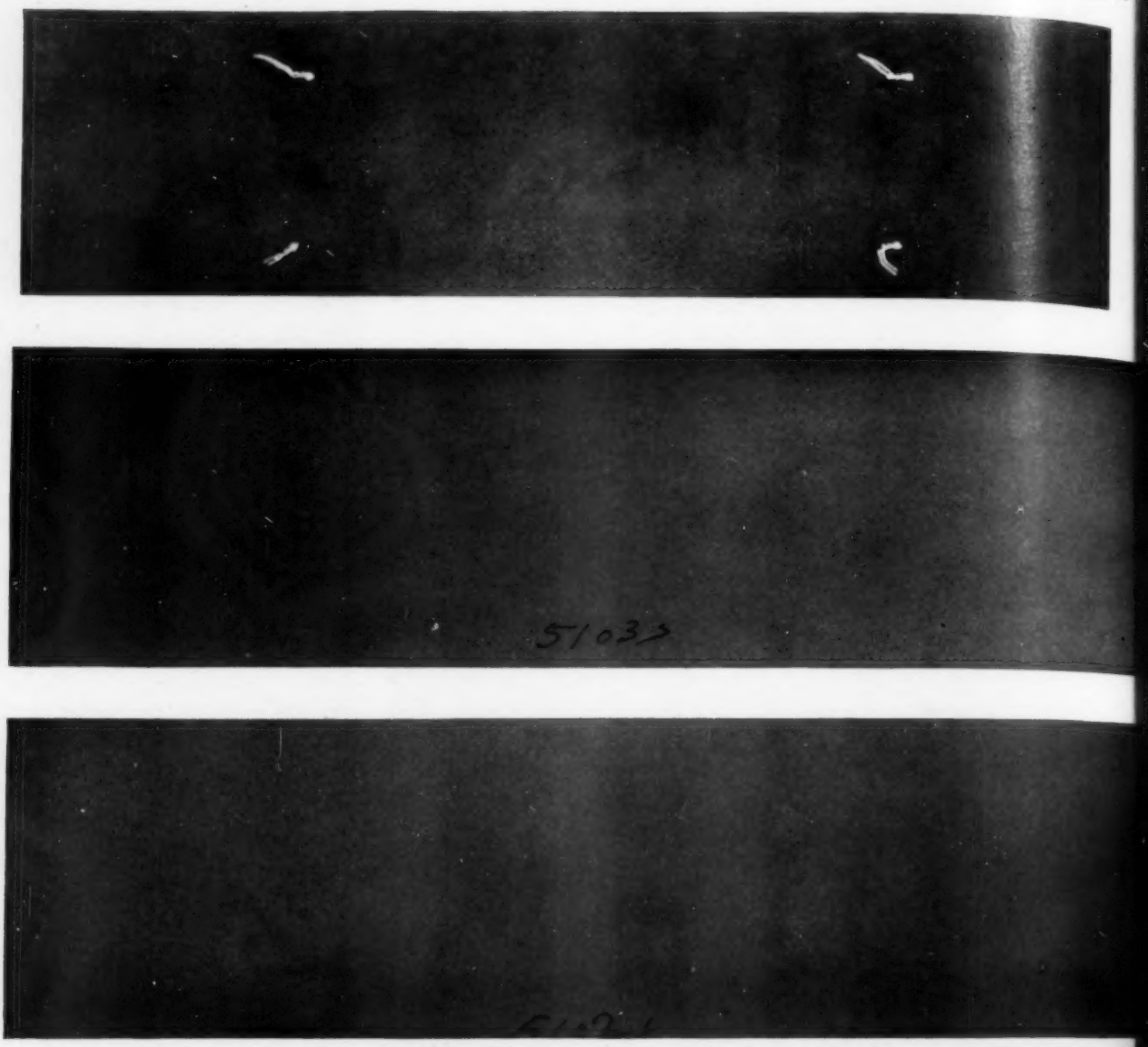


Fig. 44—Radiographic comparison of three panels, each ¾x2x8 in., cast in ASTM-AZ63. Top—Riser ¾-in. thick at the bottom, 1½-in. thick at the top, 2 in. high, 8 in. long. Unchlorinated ASTM-AZ63 alloy superheated for 15 min at 1650 F, and poured at 1450 F in green sand. Center—Same as above, except that the riser was ¾-in. thick at the base, 2½-in. thick at the top, 3 in. high, and 8 in. long. Bottom—Same riser size as employed on the casting illustrated at the top, but the melt was degassed by chlorination. These three radiographs show how microporosity is accentuated by (1) inadequate feeding and (2) the gas content of the melt. In general, the maximum soundness is obtained both by good feeding and melt quality.

tionably as described in the following paragraphs.

The melt poured into the mold having high moisture and low permeability generated enough mold gas pressure that the gas was able to penetrate into the melt in the mold before the metal solidified. These mold gases consist largely of steam, the hydrogen from which is rapidly absorbed by the melt.

When the latter solidifies, this gas is reprecipitated in the form of pinholes but due to the fact that relatively limited time is involved, the gas has only slightly penetrated to the center section of the casting. The amount of pinhole porosity increases rapidly as the surface of the casting is approached. It would be erroneous to assume that these pinholes are trapped mold gases. On the contrary, they are formed by hydrogen which was dissolved or absorbed from the moisture in the mold gases and then reprecipitated during solidification of the casting.

The pinholes in the smaller sections are quite distinct; whereas some other unsoundness is associated with the pinholes in the two heavier sections. Under these conditions some sand reaction occurs, producing rather dirty metal adjacent to the mold surface and causing rather spongy metal in some locations in the casting.

In contrast to this, the metal poured into the mold having low moisture and high A.F.A. permeability did not develop sufficient mold pressure to cause the mold gases to

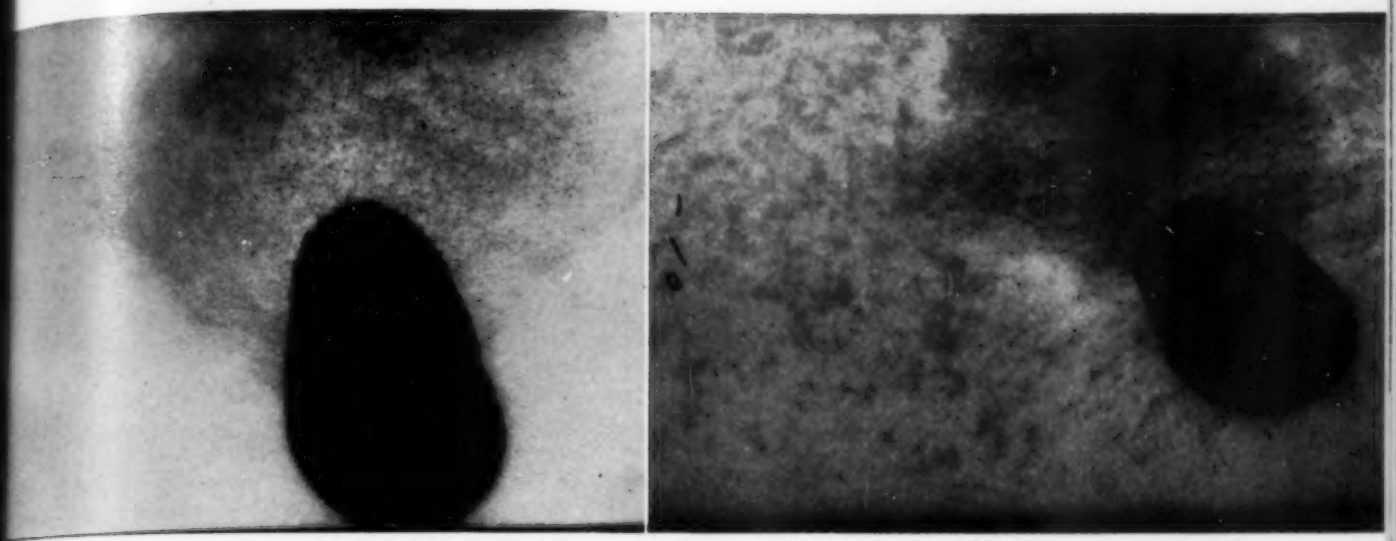


Fig. 45—Radiographs of a ½-in. thick section cut from a riser 1 in. thick, showing the relationship of shrinkage cavity, surface set, spongy shrinkage, and microporosity to the gas content of the melt. Left—ASTM-AZ63 alloy superheated at 1700 F for 15 min, cooled at 1500 F, and then degassed by chlorinating with 0.02 lb of chlorine per lb of melt. Right—Similar to radiograph at left, except that the casting was poured from the melt after passing one cu ft of hydrogen through 250 lb of metal at 1500 F. The radiograph (left) shows the shrinkage cavity surrounded by spongy shrinkage and a surface set at the top. When the melt is gassed with hydrogen, the amount of surface set decreases, the localized shrinkage remains but decreases in size, and there is a large amount of microporosity generally distributed throughout the section.

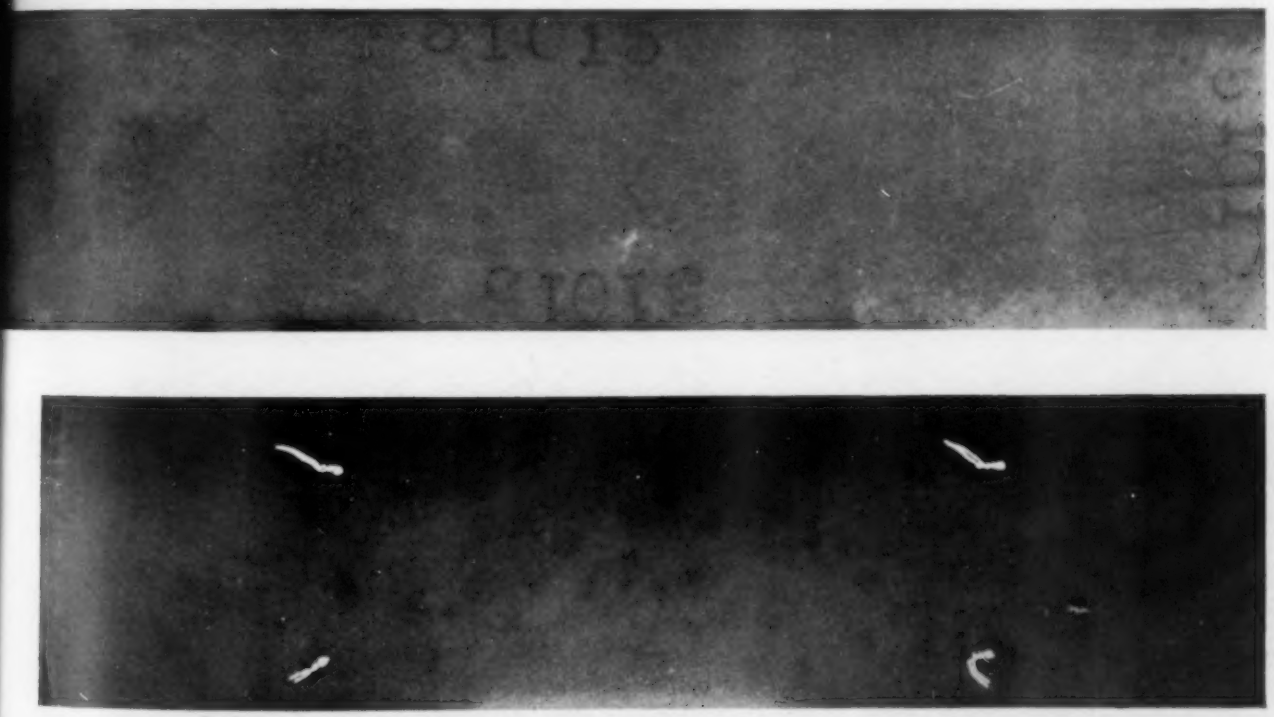


Fig. 46—Radiographic comparison of two similar panels of different thicknesses poured from ASTM-AZ63 alloy unchlorinated, superheated 15 min at 1650 F and poured at 1450 F. Top—⅜-in. panel. Bottom—¾-in. panel. These radiographs clearly indicate the oriented microporosity in the thin section and the unoriented microporosity in the heavier section. These differences probably arise from differences in solidification rates and in degrees of direction solidification, and not in the microporosity formation mechanism.

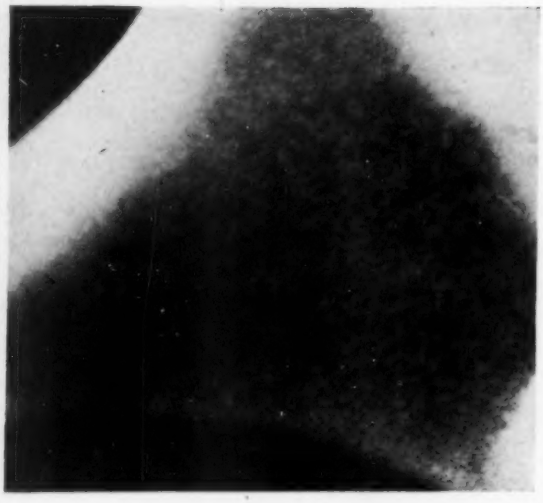
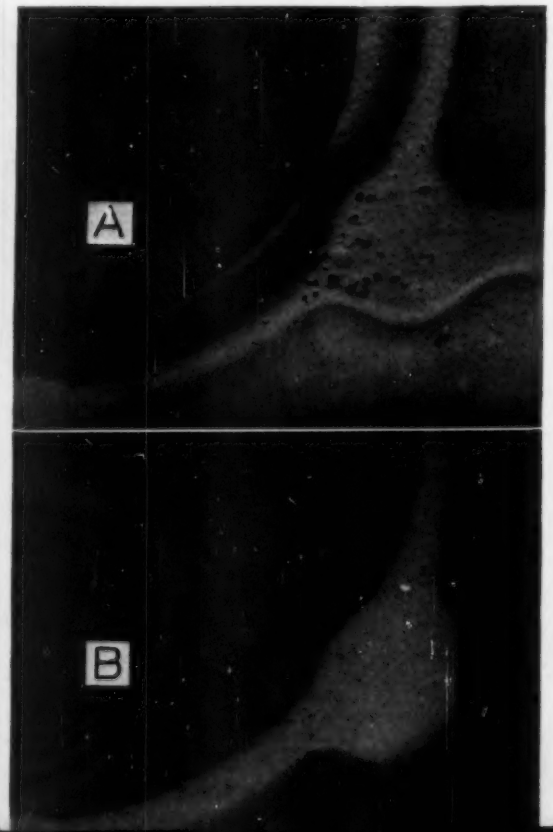


Fig. 47—Radiograph of a thin section of a No. 220 casting showing mottling caused by the coarse-grained structure. This mottling should not be confused with microporosity or pinhole porosity. ×1.

Fig. 48—(A) Photograph and (B) radiograph of a section of a No. 355 alloy crankcase showing "cope dross" on the cope surface. This defect generally is observed after sand blasting and is not related to shrinkage or to gas evolution by the melt. The term "gas holes" should not be applied to this defect because it is misleading. ×1.



penetrate the melt. As a result, there was no gas absorption because the skin surrounding the melt is good protection against hydrogen absorption unless it is broken. The lesson is obvious. The A.F.A. permeability must be relatively high, at least 15 at the minimum, and the moisture content should be kept low, consistent with sand workability.

Relation of Pinhole Porosity to Microporosity. The accentuation of microporosity by the gas content of the melt in magnesium alloys has been firmly established [2, 3, 4, 5, 6, 7], and it does not seem necessary to consider this phase of the problem further. As already illustrated by Fig. 1, magnesium alloys are not subject to the formation of pinholes but rather to large gas holes which seldom occur under commercial conditions. When they do occur, however, it will be shown that they may grade into angular slit-like cavities and into microporosity.

Solid Solution Type Alloys

Aluminum alloys form pinholes as the result of the gas content of the melt produced under commercial operation. It will be shown also that this pinhole porosity may grade off into microporosity, especially in alloys which are largely the solid solution type, but which have a small amount of eutectic. This is well illustrated by Fig. 39, which represents an aluminum alloy containing 6 per cent magnesium, 1 per cent manganese, and 1.5 per cent copper.

This alloy is essentially the solid solution type with a small amount of eutectic liquid forming at the end of solidification. This photograph, at ×50, is of a polished cross section of an ordinary tensile test bar. The polished cross section showed the characteristic generally distributed type of pinhole porosity. Examination under the microscope, however, showed that there was a great deal of porosity associated with these pinholes which approached the appearance of microporosity.

In fact, in this series of alloys, all gradations can be found between fairly rounded pinholes to rather large areas of generally distributed microporosity, which taken as a whole has the appearance, to the naked eye, of ordinary generally distributed pinhole porosity. Figure 39 shows a pinhole with associated microporosity, and it also shows one of these areas in which the pinhole is an aggregate of more or less widely distributed interconnecting channels, microporosity, between the grains.

It has already been noted that solid solution type aluminum alloys normally do not have pinhole porosity, though they may and sometimes do have this defect. Instead, they are characterized by the presence of microporosity. On the other hand, alloys which solidify over a considerable range of temperature and have a large amount of eutectic solidifying at the end normally have little or no microporosity but are quite subject to pinholing.

On the other hand, alloys such as that represented by Fig. 33, which are intermediate, are characterized by pinhole porosity grading off into microporosity. Such defects as that shown by Fig. 39 can properly be called "pinholes with associated microporosity." This term, though longer, is more accurate and more descriptive than the common one which is applied, namely, "channel porosity." This term is also decidedly to be preferred rather than reference to such defects by the erroneous term, "shrinkage."

Gas Hole Transition

With decreasing gas content of the melt, the transition of gas holes in magnesium castings from rounded shapes to angular slit-like cavities and then to microporosity has been described[3]. This is further illustrated by Figs. 40, 41, 42, and 43, with their captions. The accentuation of microporosity by the gas content of the melt has also been illustrated in a previous publication[3].

This is further illustrated by Fig. 44, which shows the reduction of microporosity obtained by degassing the melt as well as the improvement obtained by using a larger riser, thus showing that microporosity can be accentuated by either poor feeding or by gas evolution during solidification. Figure 45, with caption, is an illustration of the relationship of shrinkage, microporosity, and gas content in the melt.

Elaborate explanations have been made[6] for the oriented or "layer" appearance of microporosity in magnesium castings, but these seem unnecessary. This oriented appearance is common in thin sections where

solidificat... is rapid and considerably affected by the path of flow of the melt in the mold cavity, giving rise to the oriented appearance of the microporosity. It is not common in heavy sections where the direction of solidification is less dependent upon the path of flow of the melt in the mold cavity. This is illustrated by Fig. 46.

Mottling is a term applied to the appearance of radiographs which show the grain structure. An example of this is shown by Fig. 47, which is a thin section of a No. 220 alloy casting having a nominal composition of 10 per cent magnesium in aluminum. Mottling, showing the grain structure, is not a defect, and should not be confused with pinhole porosity or with microporosity when interpreting radiographs.

However, it must be admitted that when mottling occurs with either of these two defects, proper interpretation of the film becomes difficult. Other examples of mottling were illustrated by composition *A* of Fig. 4, and by Fig. 11. In general, mottling is more apparent the thinner the section radiographed, the coarser the grain, the better the contact of the film with the casting, and with other features which improve sensitivity of the radiographic process.

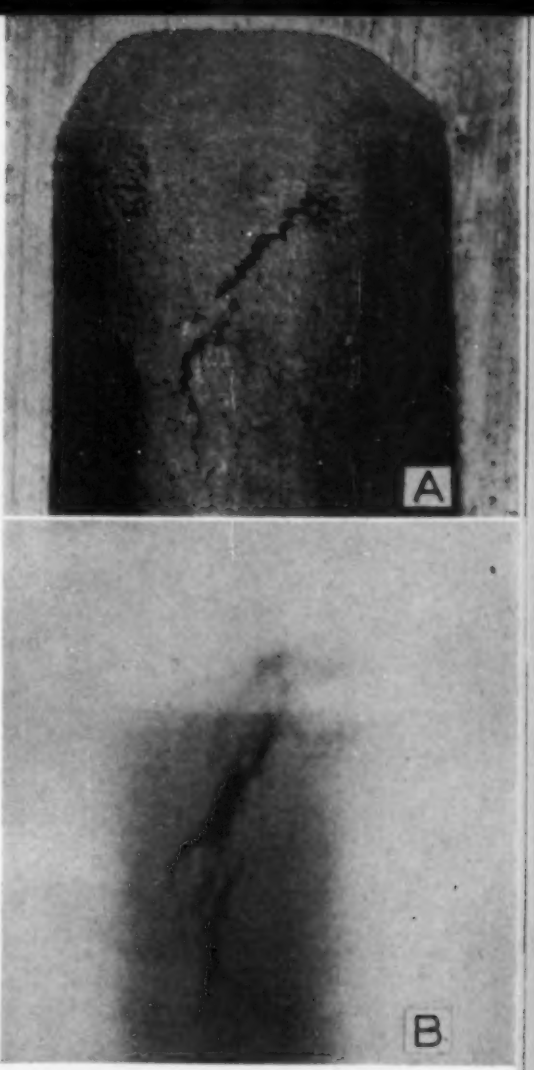


Fig. 50—(A) Photograph and (B) radiograph of a 1-in. thick section of an inadequately fed No. 355 alloy impeller casting. This crack resulted from shrinkage and is properly called a "shrinkage crack," not a "hot crack." ×1.

The exact cause of mottling has never been precisely proven, but it is clearly associated with the grain structure. It does not show a defect in the casting, except insofar as it reveals the existence of coarse grains.

Melt Oxidation

Cope Dross. One illustration of cope dross is included not because it is in any way associated with shrinkage or with the gas dissolved in the melt, but because cope dross is frequently called "gas holes." This defect is, on the contrary, a result of turbulence during the pour, or pouring improperly skimmed melts.

When an easily oxidizable melt, such as aluminum or magnesium, is poured in such a way that some turbulence occurs, some oxidation of the melt takes place. This oxide becomes entrapped in the metal with a certain amount of air, forming a "froth" consisting of these three materials—oxide, air, and melt.

Because of the air present, the over-all density of this froth is low and, as a result, tends to rise to the surface. For this reason, this defect invariably occurs on cope surfaces and may not be entirely visible until the casting has been sandblasted. Figure 48 shows a photograph *A* and a radiograph *B* of a section of a crankcase of No. 355 alloy containing this defect on the cope side.

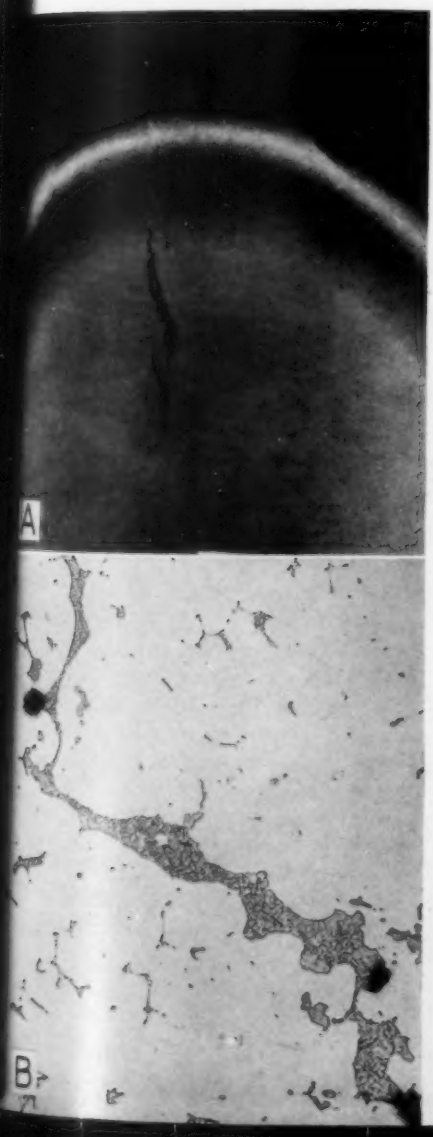


Fig. 49—(A) Radiograph of a small No. 195 alloy, cored cylindrical housing showing a hot crack partly filled with eutectic. ×1. (B) Photomicrograph of the above-illustrated crack showing the eutectic filling in a portion of the defect. The unfilled portion might resemble microporosity but should not be confused with it. ×100.

This defect is frequently referred to as "gas holes." However, it is believed that because the term "gas holes" is used for other types of defects it would be better to refer to this one as cope dross since (1) it occurs mainly on the cope side, and (2) the defect is not caused by gas but by a mixture of metal, gas, air, and oxide, which constitute dross.

Internal Stress

It is frequently stated that internal stresses produce microporosity. Internal stress applied to a hot casting will cause a hot crack, a typical example of which is illustrated by Fig. 49. This figure shows a radiograph *A* of a small No. 195 alloy, cored cylindrical housing showing a hot crack with a eutectic filling in part of the crack. This eutectic-filled portion of the crack appears as a fine white line because it has high density, whereas the unfilled portion of the crack appears as a dark line.

The partially filled portion of the crack is shown by the photograph

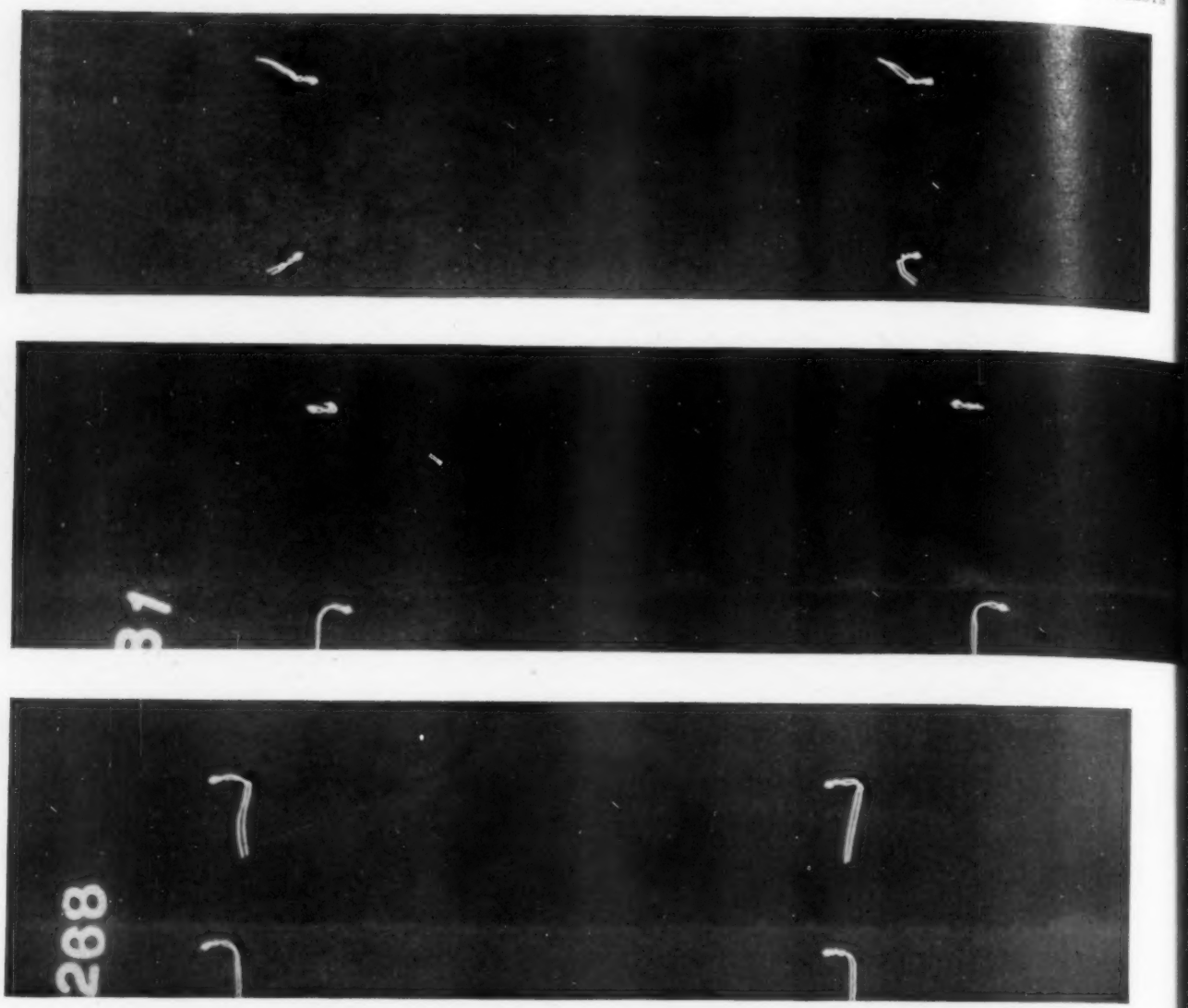


Fig. 51—Radiographic comparisons of panels ¾x2x8 in. having a riser ¾-in. thick at the base and 1½ in. thick at the top, 2 in. high and 8 in. long, poured from ASTM-AZ63 alloy, not chlorinated, superheated for 15 min at 1650 F, and poured at 1450 F. Top—Cast in green molding sand. Thermal gradient, 15 F per in. Center—Cast against a graphite chill. Thermal gradient, 58 F per in. Bottom—Cast against a cast iron chill. Thermal gradient, 171 F per in. These three radiographs clearly illustrate the reduction in microporosity obtained by increasing the temperature gradient. A high temperature gradient reduces the thickness of the mushy zone, thereby reducing the amount of the microporosity as described elsewhere.[2]

(*B*) at ×100 of a section from the casting in which the crack was filled with the copper-rich eutectic liquid. The cavities remaining in this hot crack may resemble those of microporosity. However, it is not desirable to call this microporosity because its origin is quite different. This alloy, No. 195, is quite hot short, and the hot crack shown by Fig. 49 is representative of a condition that can result from the use of an improper foundry technique.

Cracks can also occur in alloys which are not particularly hot short. This is well illustrated by Fig. 50, which is a photograph (*A*) and a radiograph (*B*) of a 1-in. section of an inadequately fed No. 355 alloy impeller casting. However, this defect is associated with obvious shrinkage and, therefore, this crack should not be called a "hot crack" but, on the contrary, it should be referred to as a "shrinkage crack."

The reason is obvious. If the inspector or metallurgist calls such a defect a hot crack, the foundryman would normally proceed to reduce the internal stress of the hot casting, perhaps by making a softer core. It is obvious that this terminology would be misleading to the foundryman, whereas, if the metallurgist or inspector properly labeled the defect as a shrinkage crack, the foundryman would correct it by more adequate feeding of this portion.

All of the illustrations, except Fig. 51, included in this paper are of sand castings. Similar defects may occur in permanent mold castings, excepting, of course, the defects arising from gas absorption from the mold, as illustrated by Fig. 38. In general, permanent mold castings solidify more rapidly and with steeper temperature gradients than is the case with sand castings.

As a result (1) the gas porosity tends to be finer, more rounded, and

more numerous, (2) the localized shrinkage tends to be more localized with a smaller volume of spongy metal adjacent to the main shrinkage cavity, and (3) as a corollary of Item 2, microporosity occurs to a lesser extent.

This is partly a result of increased temperature gradients and partly because aluminum permanent mold alloys invariably form a considerable amount of eutectic. The magnesium alloys selected for permanent mold use are also of the type which are least subject to microporosity.

The effects of solidification rate upon the occurrence of pinhole porosity in aluminum and gas holes in magnesium alloys have been mentioned. Figure 51 is an illustration of the relationship of the degree of directional solidification as affected by the steepness of the temperature gradients in the solidifying casting and occurrence of microporosity.

A number of defects which may occur in aluminum and/or magnesium castings have been illustrated. In general, the terms applied to these defects as nearly as possible convey to the foundryman the exact cause of the trouble. Accordingly, the terms applied to these defects have been of a nature which describes their appearance and indicates their cause. Where the cause is not definitely known, a term is used which is descriptive but which does not give the foundryman an erroneous impression as to the cause of the defect.

In general, the type of defect produced by shrinkage and/or the evolution of gas in light-alloy melts depends mainly upon (1) the alloy composition, (2) the gas content of the melt, and (3) the manner of solidification and the degree of directional solidification as determined by casting design, section thickness, and gating method, etc. Depending mainly upon these factors, localized shrinkage grades into microporosity, and gas porosity may also grade into microporosity so that no hard and fast lines of demarcation are formed.

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DISCUSSION

Chairman: W. E. Martin, National Smelting Co., Cleveland.

Co-Chairman: A. W. Stolzenburg, Aluminum Co. of America, Detroit.

M. C. Corson[1] *(written discussion):* The writer hoped that Dr. Eastwood might introduce new and fundamental concepts in the understanding of the causes of unsoundness and the mechanics of its development. Unfortunately, his hopes did not materialize. Quite to the contrary.

Dr. Eastwood has shown a number of pictures taken from extremely porous samples. Not having read his paper in preprint, the writer can not see how he obtained them. The writer can say, however, that not in the poorest aluminum foundry could one encounter such degrees of porosity, except perhaps in a fraction of one per cent of all castings made.

To illustrate the point, while an ordinary aluminum alloy No. 12 if properly treated would show 21,000 psi. tensile strength, as cast, it happens but rarely that a test bar from a very poor heat would run below 16,000 psi. If, however, the porosity of the test bar were such as corresponds to Dr. Eastwood's pictures it would hardly last beyond 10,000 psi. In short, the writer thinks that studies on excessively porous laboratory samples can be of no assistance to the practical foundryman; he just does not get such bad castings.

The above, however, is a point of secondary importance as far as the writer's discussion of Dr. Eastwood's paper goes. Far more important is the fallacy of his statement that pure aluminum and some eutectic and near eutectic alloys of aluminum are less subject to porosity because they solidify at a point, while the other ones solidify through a range.

Dr. Eastwood forgets evidently that a point in temperature is not a point in time, and that it actually takes a longer time for pure aluminum to solidify than it takes in the case of aluminum-copper alloys so that aluminum stands a better chance to expel the absorbed gases than all of its alloys because it will remain liquid much longer.

In fact, aluminum does not solidify at *a point* of temperature (1220° F.). It is one of the metals most subject to undercooling and it can remain liquid even below 1100° F.

The following examples shall be mentioned to refute the fallacious concept of "the range of freezing":

1. Pure copper has no range of freezing and is extremely sensitive to absorbed gases. It is very easy to obtain copper castings that look as if they were worm eaten. The same is true for pure nickel, iron, and silver.

2. An aluminum alloy with 13 per cent silicon treated with sodium is a eutectic and should solidify at a point, hence be fairly immune to porosity, but as a matter of fact it is one of the most likely to be porous. That's why the tensile strength of the modified alloy can run from 25,000 to 40,000 psi. and its elongation from 5 to 18 per cent.

In short, we still do not possess a clear understanding how porosity develops and this pertains both to gas porosity and intergranular shrinkage. We know that the first exists and is usually due to the absorption of hydrogen either directly or from the water vapors present. It can be taken care of by a thorough degassing with a non-absorbable gas or vapor. This can bring about a complete absence of pinholes.

Intergranular shrinkage assumedly can be taken care of by proper gating and risers of a proper size, but the extent of its elimination is by no means large and depends on the size of the casting, particularly its cross-section. It is that non-eliminable shrinkage that makes almost impossible to obtain a strength higher than 45,000 psi. in an alloy containing 4 per cent copper and 2 per cent silicon, while the same alloy as forged can reach a strength of 60,000 psi.

In short, Dr. Eastwood's paper throws no light on the nature of unsoundness and therefore does not indicate the ways to its elimination. On the other hand, it introduces the concept of the constitutional solidification range which has nothing to do with unsoundness.

L. W. Eastwood *(answer to Mr. Corson's written discussion:)* In reply to Mr. Corson's first three paragraphs, it is merely necessary to point out that a few typical examples of unsoundness in commercial castings are shown, but most of the principles illustrated are based on experimental castings. These principles are applicable to commercial castings. Obviously, commercial castings are unsuited to illustrate the principles involved, just as they would be unsuited for many other types of laboratory investigations.

His fourth paragraph is adequately covered in the paper since the effect of solidification rate and steepness of the temperature gradients has been described briefly. It is questionable that pure aluminum solidifies so much more slowly than an aluminum-copper alloy. Indeed, because of the higher conductivity of the commercially pure aluminum, it

[1] Metallurgical Consultant, New York.

would be expected that, if anything, the reverse would be true.

In any event, the differences in the time factor would be small. As indicated by the marked columnar grains in commercially pure aluminum, it solidifies in a very directional manner and the mushy zone is narrow because the melt solidifies at nearly constant temperature, giving ample opportunity for the gas to escape ahead of the solidifying casting.

Regarding the two points intended to prove the "fallacious concept of the range of freezing," the first is irrelevant because it pertains to copper-base alloys in which the most important gas precipitation is in the form of steam from the hydrogen-oxygen reaction. The subject of our paper is light alloys in which gas unsoundness results from precipitation during solidification because of the decreased solubility.

Regarding his second point, the writer is sure no one would suggest that any aluminum cast alloy is "immune from pinhole porosity." Those alloys solidifying at constant temperature, such as a modified 13 per cent silicon alloy for example, is less prone to form porosity than one which solidifies over a range in temperature such as the No. 122 alloy for example. The data in the paper clearly illustrate this.

Mr. Corson attributes to the presence of porosity all the variations in properties of the eutectic silicon alloy and the differences in properties obtained in a forging, as compared with a casting. Actually, porosity is only one of the factors and this is probably less important than the structure, i.e., differences in the size, shape, and distribution of the constituent phases in the alloy.

The last two paragraphs in Mr. Corson's discussion have been adequately covered in many of the papers published in this country and in England in the last few years. Several of these are listed in the bibliography. While the last word probably has not been written on this important subject, these papers will clearly indicate to the foundrymen the fundamentals involved in the formation of microporosity, incorrectly referred to as "intergranular shrinkage" in Mr. Corson's comment.

MEMBER: Would any amount of heading or anything you could have done given you a good sound picture of the 4-in. cube in the step casting to make that 4-in. cube look similar to the 1½-in. cube in soundness? Could you have headed or gated it to gain good metal there?

DR. EASTWOOD: In that step casting, the heavy section could be made sound, insofar as a localized shrinkage is concerned, by better feeding. However, the heavy section invariably exhibits more unsoundness resulting from gas evolution, than a light section. Heavier feeding would not eliminate the defects arising from gas evolution. Heavier feeding would eliminate the defects arising from shrinkage.

CENTRIFUGAL DUST COLLECTORS

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Buell Engineering Co., Inc.
New York, N. Y.

The writer discusses trap type, common cyclone type, and the volute or spiral type of centrifugal dust collectors. Centrifugal type dust collectors utilize radial acceleration of a gas stream for seperating solid particles suspended therein. The fundamental operation is separation of dust particles from the main air flow into a quiescent zone, settling out by gravity into a receiving hopper.

THE basic principle employed in centrifugal type dust collectors is the utilization of radial acceleration of a gas stream for separating solid particles suspended therein. Various designs are in use ranging from the simple trap, in which the particles are thrown out of the gas stream by virtue of a single change in direction, to the high efficiency cyclone wherein the change in direction of the air stream is continuous throughout its passage through the cyclone.

In all such collectors the fundamental operation involved is the separation of the dust particles from the main air flow into a quiescent zone, where they can be settled out by gravity in a receiving hopper.

Dust Particle Separation

This requires that the particles move across the stream, but such motion is opposed by the aerodynamic resistance on the particle, the magnitude of which is a function of the particle size, shape and surface qualities. The velocity of the particle relative to the air stream will depend upon a separating force imposed on the particle against this resistance.

In an ordinary settling chamber (Fig. 1) the separating force is the gravitational force and in the absence of turbulence the horizontal distance of travel required to settle out a particle is directly proportional to both the air velocity and the height of the particle above the bottom and inversely proportional to the settling velocity of the particle.

It is apparent from a consideration of practical values of the foregoing factors that separation of dust by this means usually involves units of impractical size. In the centrifugal collector the force imposed on the particle is a function of the radius and velocity of rotation and may be of much greater magnitude than the gravity force, so that the distance of travel required for separation of the particle to the collecting surface is reduced.

The maintenance of a minimum radial width of the entering gas stream will also require a minimum length of gas path. In the high efficiency cyclone type centrifugal collector of proper design the downward spiral flow pattern produces the maximum length of gas path in the minimum space.

An analysis of these fundamental factors, will, in general, indicate the relative separating ability of the various types of centrifugal collectors. Actually the problem is more complicated due to velocity requirements in the region of turbulent flow and the presence of secondary flow components. However, consideration of these factors is beyond scope of this paper but it can be stated that various designs have been developed to minimize the effect of these conditions.

Trap Type Collector

Simplest of centrifugal collectors is the trap (Fig. 2). It should be noted that the length of the gas path is extremely short which is a particular reason for its being ineffective except on particles of large mass. Modifications of this design have been developed using multiple vanes in an attempt to improve the efficiency but the inherent disadvantage of a single change in direction of flow would indicate a rather definite limitation to the gain that might be attained.

Common Cyclone Type Collector

So-called common cyclone (Fig. 3) can be considered next in order, with reference to separating ability. These units are of relatively large dimensions on the basis of gas volume handled as indicated by the large ratio of cyclone diameter to inlet diameter. While the inlet velocity is generally of the order of 50 ft. per sec., the true cyclonic action is not in evidence because of the large gas volume within the cyclone compared to through-put volume.

The large dimensions provide a long holding time but the low rotational velocity in these units means a low separating force on the particles. The application of these units is therefore usually limited to rather coarse dust.

Volute or Spiral Type Collector

Next design to be considered is what we might call the volute or spiral collector (Fig. 4). This usually takes the form of a scroll similar to a fan casing except that the flow is in the opposite direction. Usually at about 270° F. from the inlet a slot or slots are provided for passing the dust thrown to the periphery into a hopper or secondary collector.

The gas follows a spiral path

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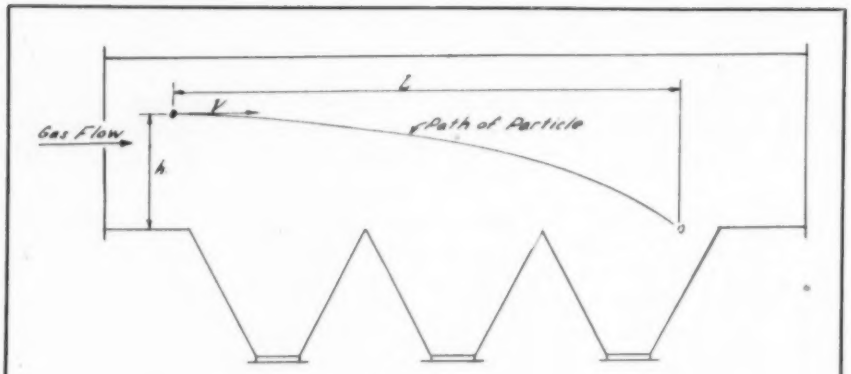


Fig. 1—Sketch of settling chamber illustrating settling of dust particles.

toward the center and leaves through a central opening. Vanes are sometimes provided to form a cylindrical baffle so that the gas is made to reverse its direction before leaving as a means of further improving the separation.

This design has in some cases been used to form the inlet boxes on fans thus incorporating a fan and collector as an integral unit. It is essential that a proper withdrawal of the separated dust is attained to assure a definite flow through the slot into the collecting hopper or secondary collector. This gas flow must necessarily rejoin the main stream and the required pressure differential may be obtained by connection to a point of low pressure or the use of a small fan capable of handling the shunt flow.

Essentially, the length of path is limited to less than one revolution although some designs are proportioned to provide for recirculation (Fig. 5), in which case the distance is increased with a corresponding increase in efficiency. However, the comparatively large dimensions required means a low separating force so that here again the application is limited to the coarser dust.

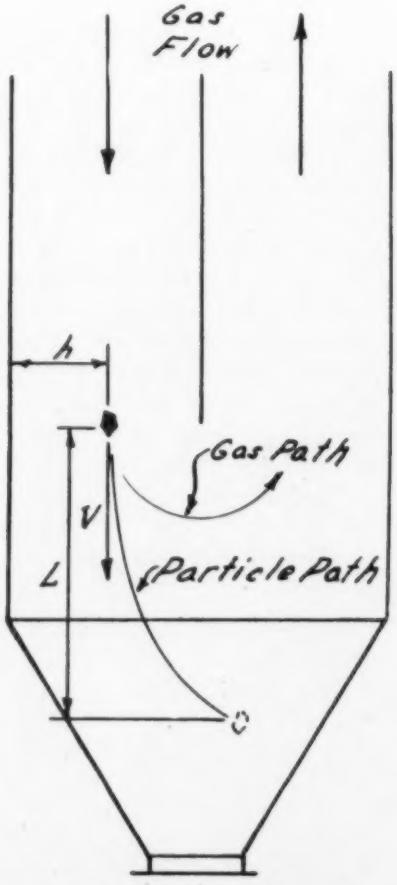


Fig. 2—Sketch of trap type centrifugal collector showing settling of dust particles.

High Efficiency Cyclone

At the present stage of the art the high efficiency cyclone (Fig. 6) is the most effective centrifugal collector available. There are many variations of this type but in general they have an inlet at or near the top with the greatest diameter at this point and taper down to a small diameter dust discharge opening at the bottom.

The gas outlet consists of a concentric tube extending down into the cyclone at various distances from the top. The flow pattern takes the form of a descending spiral at the periphery and an ascending spiral in the center.

The gas rotation is achieved either by an inlet tangential to the cyclone body and the outlet tube. In the latter case directing vanes are disposed in the annular space to impart a rotational flow to the entering stream.

There have been many attempts toward mathematical analysis of cyclone operation and establishment of basic equations. However, the complexity of the many factors involved has prevented any great degree of success along these lines. At best any equations that have been evolved have been limited to units of particular design proportions.

Most of the work done in the development of high efficiency cyclones has therefore been of an empirical nature. In the case of designs, with which the author is concerned, individual experiments have run into the thousands.

In view of the effect of dimensional proportions on the performance of cyclonic collectors, it is understandable that various investigators have evolved different concepts as to the relative importance of the various factors and have concentrated their efforts on designs in which one in particular is stressed.

Various Factors Considered

However, a proper balance of these proportions is essential since extreme values of one particular factor, while tending to produce higher efficiency may effect other factors to a greater extent in the direction of lower efficiency. By way of illustra-

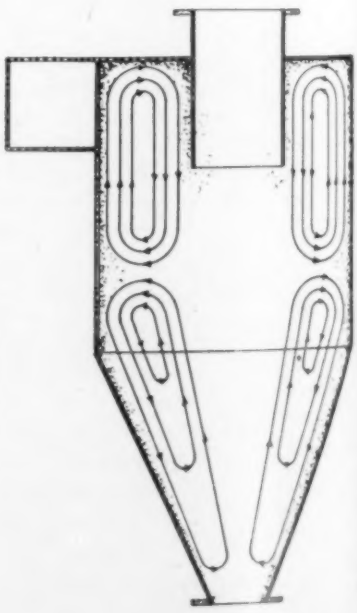


Fig. 3—Sketch of common cyclone type of dust collector.

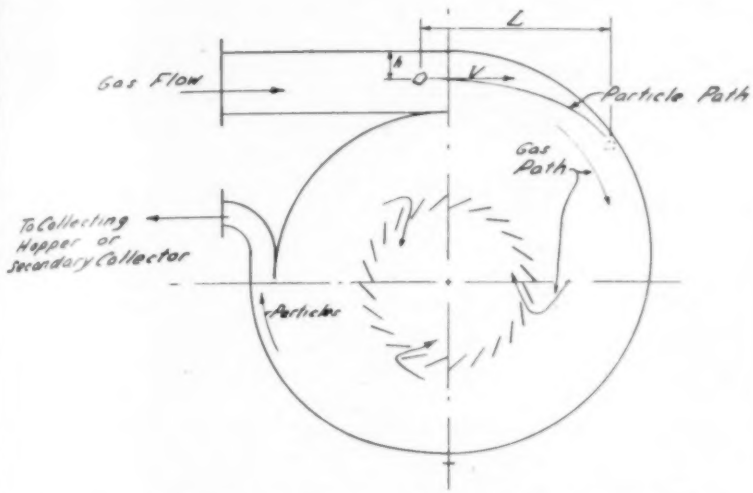


Fig. 4—Sketch of volute or spiral type dust collector showing path of dust particles in settling.

tion let us consider some of the conditions which promote good performance in a cyclone.

It is fundamental that at a given tangential velocity the separating force on the particle is greater at the smaller radii. However, this does not mean that a small diameter unit is inherently more efficient than one of large diameter, although it does mean that a smaller diameter unit of particular proportions is inherently more efficient than a large diameter unit of the same proportions.

The distinction lies in the fact that in a unit of diameter D tapering down to a dust discharge of $1/2\ D$ the separating force at the discharge is about eight times that at the inlet, whereas in a unit of diameter $2D$ and dust discharge diameter $1/2\ D$, the separating force at the discharge is about 64 times that at the inlet or about eight times that in a unit having a diameter only $1/2$ as large. This would indicate that a small ratio of dust outlet diameter to cyclone diameter is essential.

However, it is also necessary to have a large ratio of overall height to cyclone diameter. It is readily apparent when we consider that if the cyclone tapers too sharply the vertical component of the reaction of the conical wall to the centrifugal force on the particle may prevent its downward motion to the collecting hopper.

The radial width of the inlet is also an important factor since both theory and practice indicate that with a decrease in width an increase in efficiency can be expected. There is of course a limit to this decrease where other proportions remain the same since it thus approaches the case or the ordinary cyclone having a large ratio of cyclone volume to through-put volume.

The position and diameter of the gas outlet pipe will effect the efficiency of a given unit. The depth to which these pipes should extend below the top for best performance cannot generally be stated since it is more commonly related to the overall design but a decrease in its diameter, will, in general, mean an increase in efficiency as well as increase in pressure drop.

The foregoing discussion has been confined to the design of single elements. In practice, it is generally found expedient to use multiple elements, having common inlets and outlets and discharging to a common hopper. It is essential in such

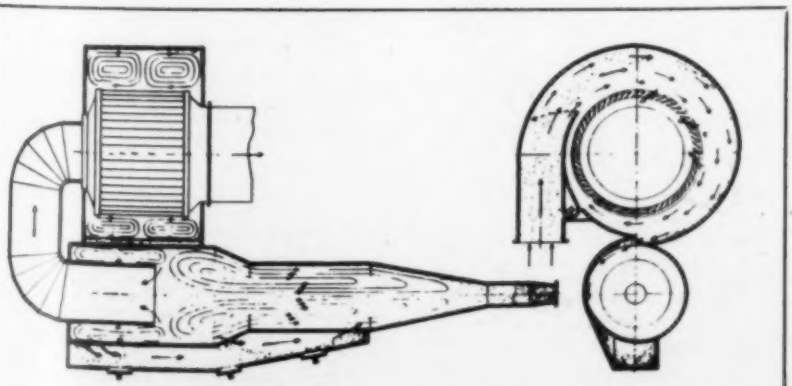


Fig. 5—Another sketch of volute or spiral type dust collector provided with recirculation.

Fig. 6—Diagram of a high efficiency cyclone type dust collector.



cases that the individual elements equally share the load, since any inequality will result in circulation between elements by way of the hopper, thus reducing efficiency. The design of the inlet and outlet should therefore insure proper distribution to the elements.

In the high efficiency type of collector the action of the vortex wherein high spin velocities obtain at the dust discharge, the static pressure in the hopper is usually lower than the static pressure in the gas outlet. This means that even with the collector on the discharge side of the fan a negative pressure can exist in the collecting hopper.

It is therefore essential that the hopper be properly sealed at all times against any inleakage since admission of air to the hopper creates an upflow into the cyclones which has the effect of materially reducing the efficiency.

With regard to applications in the foundry the centrifugal type collector has a definite place on various operations. They have been successfully used on knock-out and shake-out screens, on sand blast and shot blast machines, as well as grinding and polishing operations.

It must be borne in mind however that in view of its inability to separate completely all the dust from the air stream the exhaust air should not be returned to the working area but should be exhausted outside the building.

In conclusion, it may be stated that progress is still being made in improvements in design and performance of centrifugal type collectors and while gains can be made to improve the inherent characteristic of decreasing efficiency with decrease in particle size, it is believed that complete collection of the very small particle sizes of the order of five microns will have to be achieved by some other means.

WELDING HEAVY GRAY IRON CASTINGS

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WELDING APPLICATIONS on gray iron differ from those on most other metals. Welding on gray iron has been largely limited to repair and salvage operations. In some cases pads, bosses, and also other more complicated parts, such as brackets, have been added to iron castings by welding. However, the manufacture of gray iron structures by welding together two or more castings is not common practice and, as a result, production procedures have not been developed.

Gray Iron Sensitive

Limited use of welding on gray iron is due, in part, to the difficulty of achieving ideal conditions for making satisfactory welds. Because of its composition, gray iron is sensitive to the effect of heat treatment. If it is heated above its critical temperature and cooled at a rate exceeding about 600 F per minute it becomes hard and brittle, and is easily cracked by stresses caused either by loading or by temperature differentials.

In all of the welding processes, the material adjacent to the deposited metal is heated above the critical temperature for gray iron; hence, the only method of preventing the hardening of this zone is to retard the cooling rate. The most effective way of doing this is to preheat the adjacent portion of the casting.

Temperature to which the metal must be preheated and the size of the preheated area depend upon the welding process to be used and the size and shape of the casting. Although specific rules governing all cases cannot be given, the following general principles will serve as a guide when repairs of maximum strength are desired:

1. Heat the surface on which metal is to be deposited sufficiently to obtain a good bond between the weld and the parent metal.

2. Heat the area adjacent to the repair sufficiently to prevent hardening in the weld zone.

3. Heat other parts of the casting in such a manner that excessive stresses will not be caused by the welding operation.

A general preheat to a high temperature, about 1000 F, would meet all of the foregoing requirements, but such a preheat is seldom possible on large castings. Furthermore, the welder would not be able to work on castings heated to this temperature. Therefore, a compromise is necessary, and either a lower general preheat or a local high temperature preheat is used.

Many successful repairs have demonstrated the feasibility of welding heavy gray iron castings, the essential factor being the proper selection of a welding process and a method of procedure for the particular application.

For repairing gray iron castings, a number of methods have been successful. Among these are:

a. Metallic arc welding.
b. Oxyacetylene welding.
c. Carbon arc welding.
d. Bronze welding.
e. Thermit welding.
f. Burning.

Each method has certain advantages and certain disadvantages, and the particular process to be used depends upon circumstances. Sometimes two or more processes may be used on the same casting.

Metallic Arc Welding. For metallic arc welding, either ferrous electrodes (low carbon steel or cast iron), or non-ferrous electrodes (generally monel metal or an equivalent composition) may be used.

Repairing

Preparation for making a repair consists of chipping away any unsound metal and, if the defect is a crack, of chipping a V-groove from one or both sides of the casting. For castings up to about 1½-in. thickness a single V-groove may be used, but on heavier sections a double V-groove is desirable if both sides of the casting are accessible.

The groove should be sufficiently wide at the bottom to permit of proper manipulation of the electrode and should have an included angle of from 60 to 90 degrees. For best results, back chipping to sound metal and welding the second side are essential for both single and double V-grooves. A shear V-groove, which is a V-groove with a wide top, is sometimes used to increase

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the bonding area between the weld metal and the parent metal.

In most applications of arc welding it is desirable to use threaded studs in the groove. These studs vary in size from ⅛-in. for the thinner sections to ½-in. for heavy sections. The studs should be well distributed over the surface of the groove or weld area, spaced at about ½-in. for the smaller studs and up to 2 in. for the larger studs.

If the studs are to be fully effective care must be taken to obtain good fusion around and onto the studs, thus attaching the weld metal securely to the studs which, in turn, are anchored into the cast iron below the heat-affected zone. Without the use of studs, arc welds may pull away from the parent metal, as shown in Fig. 1. Cracks such as those shown in this specimen generally occur in the heat-affected area and are not due to lack of fusion between the weld metal and the cast iron.

With the metallic arc process, preheating of the base metal in order to obtain fusion between it and the deposited metal is not required, but hard spots in the weld area will occur if preheat is not used. Figure 2 shows the results of hardness surveys on single-bead welds made with two ferrous and two non-ferrous electrodes, with no preheat, and with 500 and 1000 F preheats. The curves at the left show the hardness in the heat-affected zones, and those at the right the hardness of the weld deposits.

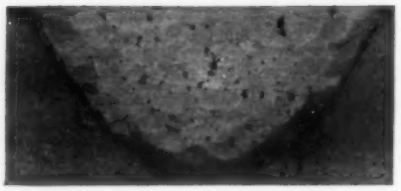


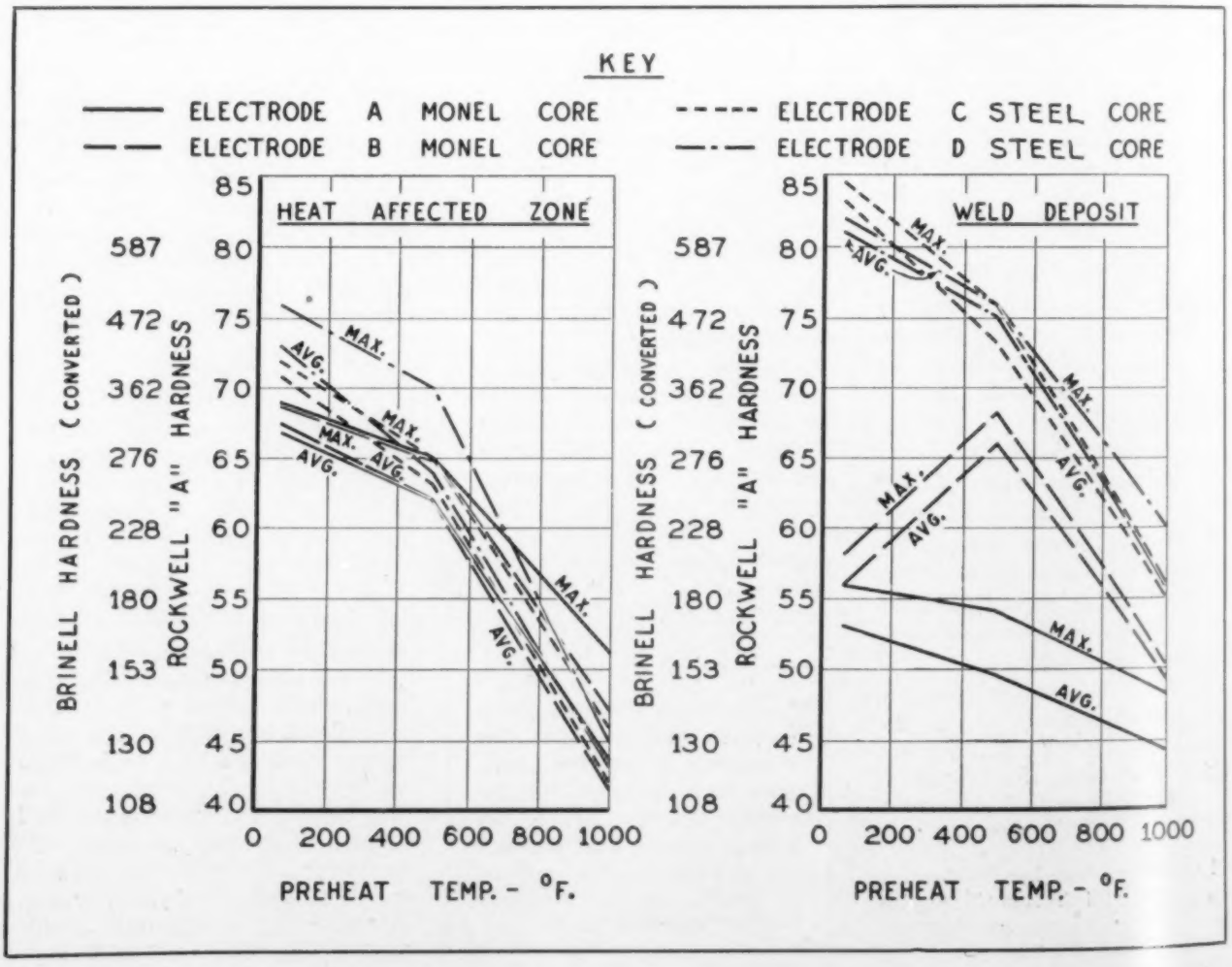
Fig. 1—Metallic arc weld with coated steel electrodes and no preheat. ×1.

As might be expected, all of the electrodes produce high hardness in the heat-affected zones when no preheat is used. The values are somewhat higher for the ferrous than for the non-ferrous types of electrodes, but the difference is not great. At a preheat of 500 F the hardness values, with one exception, are between 245 and 275 Brinell for all electrodes, and at the 1000 F preheat all hardness values are below 160 Brinell.

Hardness values of the weld deposit show a marked difference between the ferrous and non-ferrous types of electrodes. With steel electrodes the hardness is above 400 Brinell for preheats of 500 F or less, and between 190 and 225 Brinell for a 1000 F preheat. The monel type electrodes show values below 210 Brinell for all preheats except one electrode at 500 F, which has a maximum hardness of 320 Brinell.

With multipass welds, both the

Fig. 2—Curve showing hardness values of single-bead metallic arc welds.



weld deposits and the heat-affected zones are softened materially by the heat of subsequent layers, but this does not eliminate the danger of cracking in the hardened areas before the next layer is deposited. Even with a preheat of 500 F, multipass welds made with either ferrous or non-ferrous electrodes may develop cracks, as is shown in the photomicrographs (Figs. 3 and 4).

Weld Strengths

To determine their strengths, welds of several types were made in standard gray iron castings such as that shown in Fig. 5. This shape of casting was chosen because it offers some restraint to the shrinkage of the weld. The strength values for the welds were obtained from tension specimens located as shown in the upper part of Fig. 6. The numerical values shown in the lower part of Fig. 6 apply only to a particular bronze weld.

Metallic arc welds were made with ferrous and with non-ferrous electrodes with no preheat and with a preheat of 500 F. Since it was desired to determine the strength of the junction of the weld to the cast iron, no studs were used.

With no preheat the welds pulled away from the cast iron, as shown in Fig. 1, and no tension tests were made. With a preheat of 500 F the weld made with the ferrous electrodes had a tensile strength of 11,000 psi across the junction and 49,000 psi for the deposited metal. The corresponding values for the weld made with the non-ferrous electrodes were 12,200 and 70,000 psi, respectively.

Structure Distortion

Metallic arc welds were not made with a preheat of 1000 F since other welding processes are generally preferred if this amount of preheat is permissible.

The fact that fusion onto the base metal can be obtained without preheat with the metallic arc process leads to its use on structures in which distortion must be kept at a minimum, and for which maximum strength of the repair is not required. Cast iron cylinders of low pressure steam turbines are structures of this type.

After years of service the material between the blade grooves

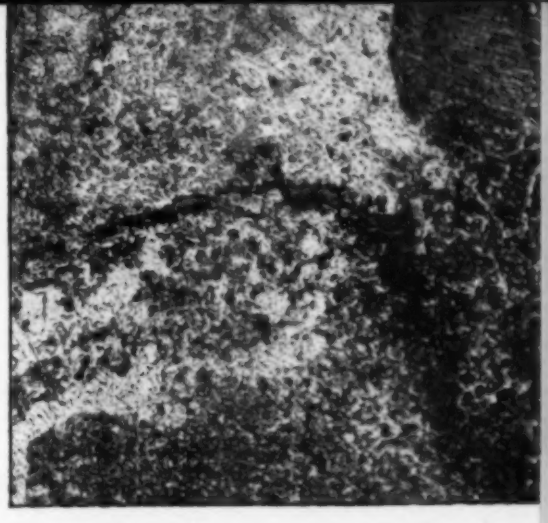


Fig. 3—Crack in heat-affected area and weld deposit (right) of metallic arc weld with steel electrodes and 500 F preheat. ×100.

often corrodes sufficiently to impair the support for the blades. The corroded areas are built up by welding without preheat since distortion of the cylinder would render it useless for further service.

The areas to be repaired are studded with ⅛-in. screws, and lightly coated monel electrodes are used. By depositing only short beads and skipping around to different parts of the cylinder it is kept sufficiently cool to touch with the hand. Cylinders have been repaired by this method for more than 20 years and rewelding has never been necessary. The improved life after welding is due to the superior corrosion resistance of the monel metal deposit.

When the distortion resulting from a moderate preheat is not serious but a high preheat is objectionable, metallic arc welding may be used with a preheat of about 500 F. This results in a repair with sufficient strength for many structures, especially if studs are used. Such welds, made with non-ferrous electrodes, are machineable.

Gas Welding

Oxyacetylene Welding. Oxyacetylene welding or gas welding, as it is frequently called, is done with the oxyacetylene torch adjusted to a neutral or slightly carburizing flame. Cast iron filler rods having about 3.5 per cent carbon and 3.0 per cent silicon are suitable, and a cast iron welding flux is used.

For repairing cracks in castings, a V-groove similar to that used for arc welding is prepared. The groove need only be wide enough to permit proper manipulation of the welding torch, and studs are unnecessary.

Parent metal around the weld area is generally heated to 1000 F or more to speed up the welding operation. This may be done with a charcoal fire or with large gas or oil burners. Hardening is not a problem in the case of gas welding since the heat required to obtain proper fusion will raise the temperature of the adjacent metal sufficiently to prevent the weld area from cooling faster than the critical rate.

Size and shape of the preheated area are determined by the shape of the structure being repaired, the object being to heat it in such a manner that the structure will not be subjected to excessive thermal stresses during heating or cooling.

Flux is sprinkled in the groove and the local area at which the welding is to be started is heated to the melting temperature. The filler wire, dipped in flux, is heated to a red heat and rubbed onto the fused area. With continued heating a pool of molten metal is formed, and the tip of the filler wire should be kept submerged in this pool.

If the filler wire is melted and allowed to drop into the pool, slag inclusions and porosity are likely to result. Working the filler wire in the molten pool helps to float out the slag. Care must be taken to melt the parent metal ahead of the molten pool in order to obtain good fusion. Flux is added as the welding progresses.

After the welding is completed the structure should be covered so

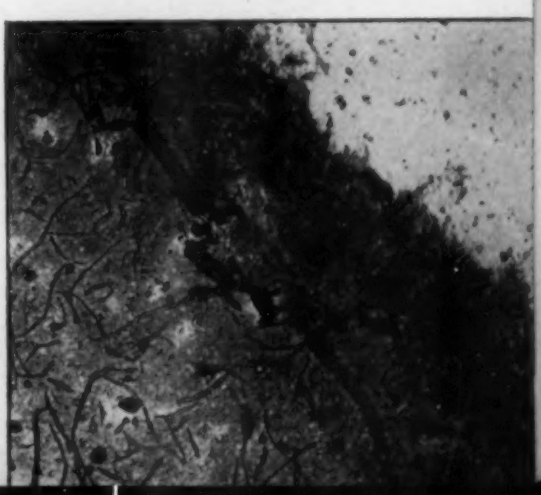


Fig. 4—Crack in heat-affected area of metallic arc weld with monel electrodes and 500 F preheat. ×100.

as to cool slowly. Stress-relieving at temperatures of 1100 to 1200 F is beneficial, especially if the structure is rigid.

An oxyacetylene weld, properly made, will have a structure similar to that of the parent metal except that the grain size will be smaller. Figure 7 shows the junction between an oxyacetylene weld and

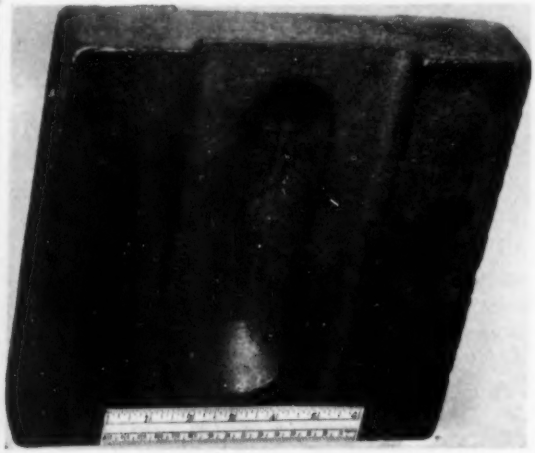


Fig. 5—Cast iron plate used for welding and brazing tests. This type of casting was selected because it offers some restraint to shrinkage of the weld.

gray iron. The strength of the deposited metal corresponds to that of cast iron of a similar analysis and, if the welding procedure has been carried out so as to obtain proper fusion, the joint will have a strength approximately equal to that of the parent metal.

Oxyacetylene welds with cast iron filler rod have the same color as the parent metal, which is sometimes important, especially when the repair is to be machined.

Carbon Arc Welding. Although this is an arc welding process it resembles oxyacetylene welding since the same types of filler wire and flux are used. However, an arc between carbon or graphite electrodes and the work is used to furnish the heat necessary for making the weld.

Inasmuch as the heat of the arc is sufficiently intense to melt the parent metal without difficulty, preheating is not required in order to obtain fusion. However, preheating is necessary to prevent the heat-affected area from hardening and cracking, just as in the case of metallic arc welding. The weld area should be heated to about 1000 F, and the surrounding area heated sufficiently to prevent excessive temperature stresses. After the weld is completed the casting should be covered so as to cool slowly.

Welds made by this process will be easily machineable and the color of the weld metal will be the same as that of the casting. If the welding is properly done the strength of such welds will be satisfactory, but it is more difficult to control this welding procedure than it is that of the gas welding method.

Bronze Welding. This welding process or brazing is similar to gas welding except that a brass filler wire of approximately 60 per cent copper and 40 per cent zinc composition is used in place of the cast iron filler wire. The oxyacetylene flame adjusted to be slightly oxidizing is used as a source of heat. It is recommended that two types of flux be used, a "tinning" flux for coating the surface of the cast iron and a welding flux for the remaining brazing operation. These fluxes are commercially available.

For repairing cracks in castings, a V-shaped welding groove, similar to that used for arc and gas welding, is

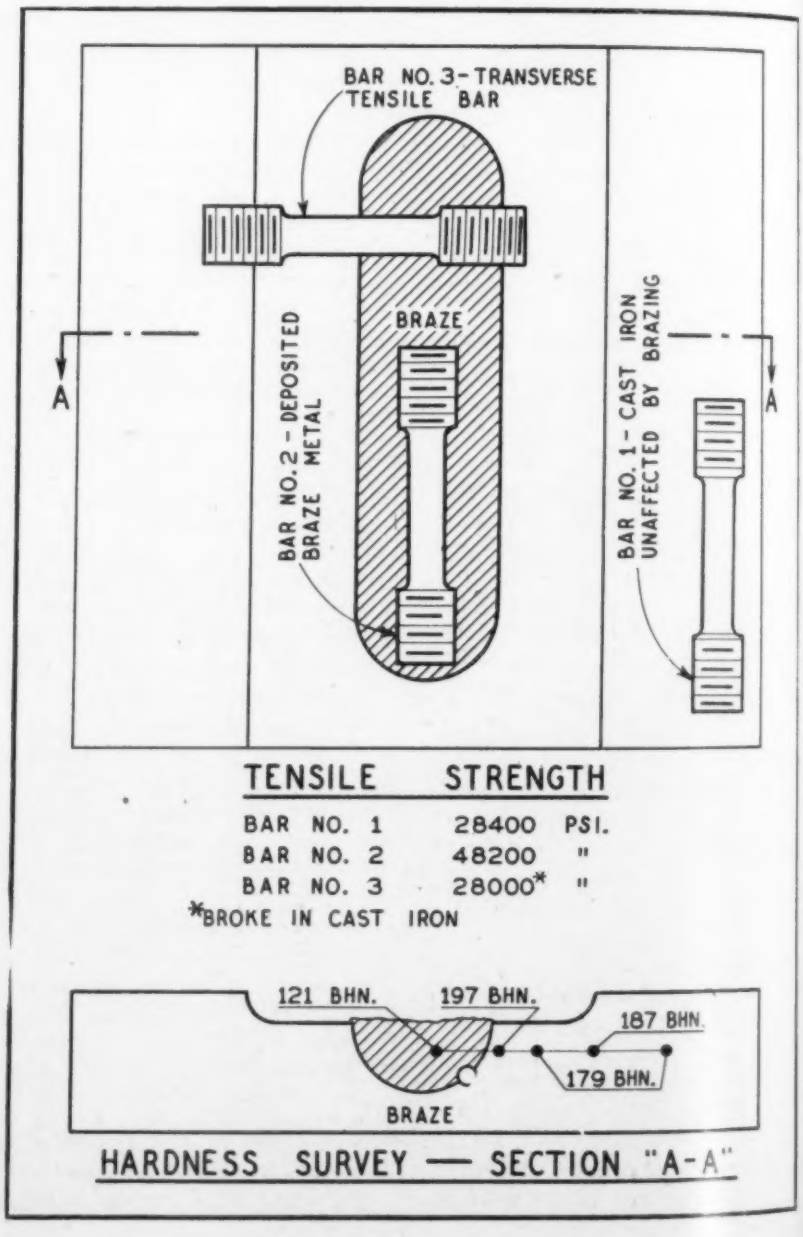


Fig. 6—Layout showing location of tension specimens cut from test plates, and physical properties of brazed joint.

used. This groove should be chipped rather than ground because grinding tends to coat the cast iron with graphite which interferes with the brazing process. Cleaning with a wire brush is helpful, and excess graphite may also be burned off with an oxyacetylene torch adjusted to an oxidizing flame.

Preheat Developed

The weld area is preheated by any convenient means in order to speed up the welding operation. As in the case of oxyacetylene welding, the bronze welding operation develops sufficient preheat in the material adjacent to the welding groove to prevent hardening of the heat-affected area.

Some of the tinning flux is sprinkled into the groove and a small area at the point where the weld is to be started is heated to a temperature slightly above the melting temperature of the brass filler rod, which is about 1650 F.

Behavior of the brass on the cast iron indicates when the proper temperature has been reached. If the cast iron is too cold, the brass does not spread out or "tin" the surface, and if the temperature of the iron is too high the brass collects in little balls on the surface of the cast iron. The fact that bronze welding is done at a temperature of several hundred degrees below the melting point of cast iron is an advantage in favor of this method of welding.

The welding groove must be tinned with brass ahead of the weld to insure that the deposited metal will not be cast in the groove without fusion. When tinning is properly done the brass penetrates between the grains of the iron, as shown in Fig. 8 at ×25, and in Fig. 9 at ×100.

Improper "Tinning"

The result of improper tinning is shown in Fig. 10 at ×25. In this case will be noted not only a lack of intergranular penetration of the brazing material but the presence of an oxide film between the two surfaces. Obviously, under this condition a satisfactory joint is not obtained. After the groove is tinned the weld is built up in layers until complete, and the casting is then covered and allowed to cool slowly.

A bronze weld, properly made, results in a repair which is as strong as the casting. As may be seen from values given in Fig. 6, the bronze deposit had a tensile strength considerably greater than that of the gray iron, and the transverse specimen had a strength equal to that of the parent metal. Figure 11 shows the transverse specimen, with the weld at the right and the fracture at the left about 2 in. from the edge of the weld.

From the standpoint of strength, bronze welding is one of the best methods of repairing gray iron, and the chief objection to this method of repair is that the color of the weld does not match that of the parent metal.

Thermit Welding. The thermit welding process for gray iron is similar to that for steel except that a cast iron thermit mixture is used for gray iron. The process is used for making repairs on many types of structures, but is not suitable for welding long cracks in thin sections or for welding cracks which extend only partially through a section.

However, in the latter case, it may be feasible to cut the remaining portion of the section and make a complete weld. According to the manufacturer's recommendations, the process should not be attempted if the length of the crack exceeds about eight times the thickness of the material because the shrinkage of thermit iron is about twice that of gray iron.

Transverse cracks are likely to de-

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Fig. 7—Oxyacetylene weld (right) with cast iron filler wire in gray iron test plate. ×100.

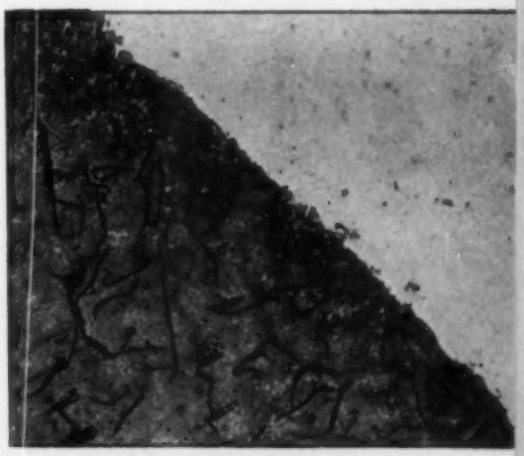
Fig. 8—Junction of bronze weld in gray iron. Good brazing procedure. ×25.

Fig. 9—Junction of bronze weld in gray iron. Good brazing procedure. ×100.

Fig. 10—Junction of bronze weld (right) in gray iron. Improper tinning procedure. Note oxide film at junction. ×100.









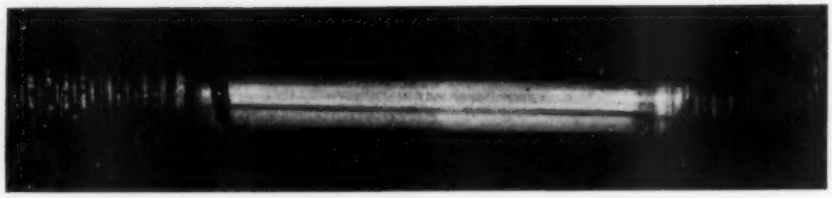


Fig. 11—Fractured transverse tension specimen from bronze weld.

velop when the length of the section is too great in proportion to its thickness. If the section is compact there is practically no limit to the size of sections which can be successfully welded.

Preparation for thermit welding is simple. The ends of the broken member are machined or otherwise cut so as to form a parallel-sided gap at the section which is to be welded. The width of the gap increases with the size of the section, and tables are available showing the proper gap for various cross sections. After the preparation is completed, the broken pieces are set up in alignment with proper allowance for shrinkage and a wax pattern is made to conform to the desired shape of the casting after completion of the weld.

A mold box of suitable size is then set in place around the broken member, and this is filled with thermit molding material with the various gate and riser patterns in their proper positions. After the molding material has been rammed into place the patterns are removed. Figure 12 shows a typical setup for making a thermit weld.

Preheating Weld Area

When the mold has been completed the ends of the casting next to the gap are preheated to a red heat with a kerosene-air burner through the gates provided for that purpose. The preheating melts out the wax pattern. In the meantime the pouring crucible is set in position above the mold and the correct amount of thermit mixture is placed in the crucible.

When the preheating is completed, the preheating gates are plugged and the thermit mixture is ignited with a special powder. The time of the thermit reaction is independent of the amount of thermit metal in the crucible, and is about 30 sec.

As soon as the reaction is completed, the crucible is tapped and the molten metal flows into the space formed by the wax pattern, while the slag from the reaction fills the slag basin in the mold. The superheat in the thermit metal is sufficient to melt some of the iron adjacent to the gap, thus insuring a bond between the added metal and the parent metal of the original casting.

Weld Metal Properties

The thermit metal is slightly harder and tougher than the original gray iron, but the joint can be machined. Welds made by this process have low lock-up stresses and stress-relieving is not necessary unless the repaired member is a part of a frame or rigid structure.

Thermit welds made on gray iron sections suitable for the process have shown satisfactory service records. Figure 13 shows a thermit repair weld in a press frame.

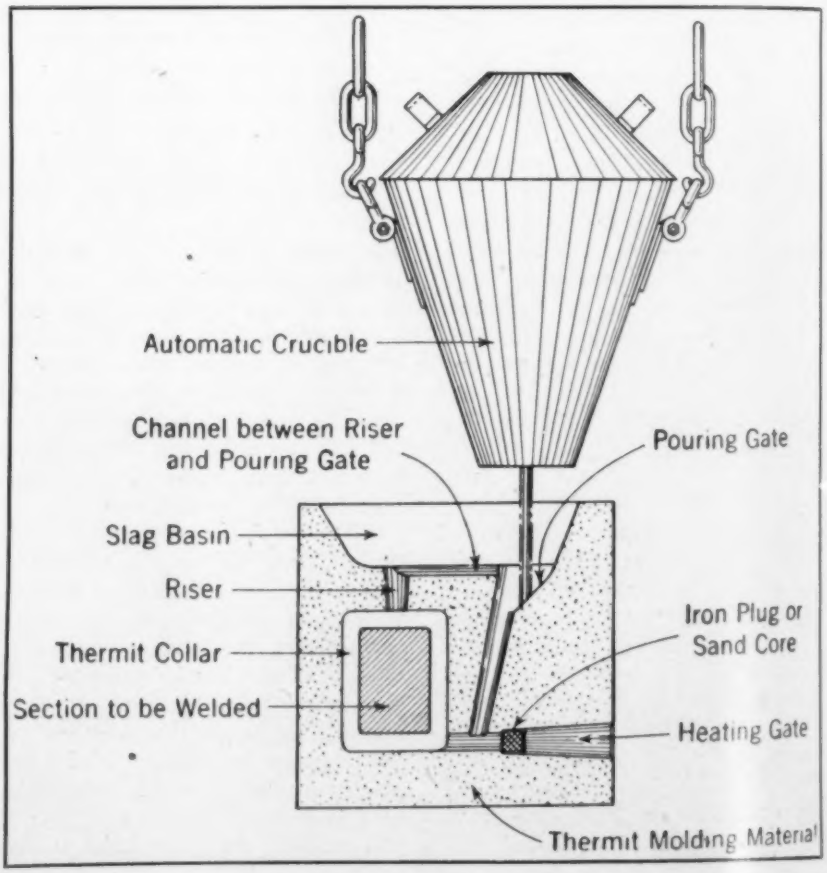
Burning. Although "burning" is not generally classified as a welding process, it is one of the accepted methods of repairing gray iron castings, and also of adding parts to finished castings. Basically, it is quite similar to thermit welding except that the molten metal for making the burn is obtained from a cupola.

Casting Preparation

The casting is prepared by chipping or other means and placed in position so that the crack or area to be repaired is in a horizontal plane. A dam of loam sand is built up around the area so as to form a well with an opening at one end to provide flow-off for the excess metal. If the repair extends through the full thickness of the casting, refractory material is built up below the opening to act as a retainer for the molten metal.

The area of the casting around the repair area is preheated to a temperature of 400 to 500 F by a charcoal fire or other convenient

Fig. 12—Setup for making thermit weld.

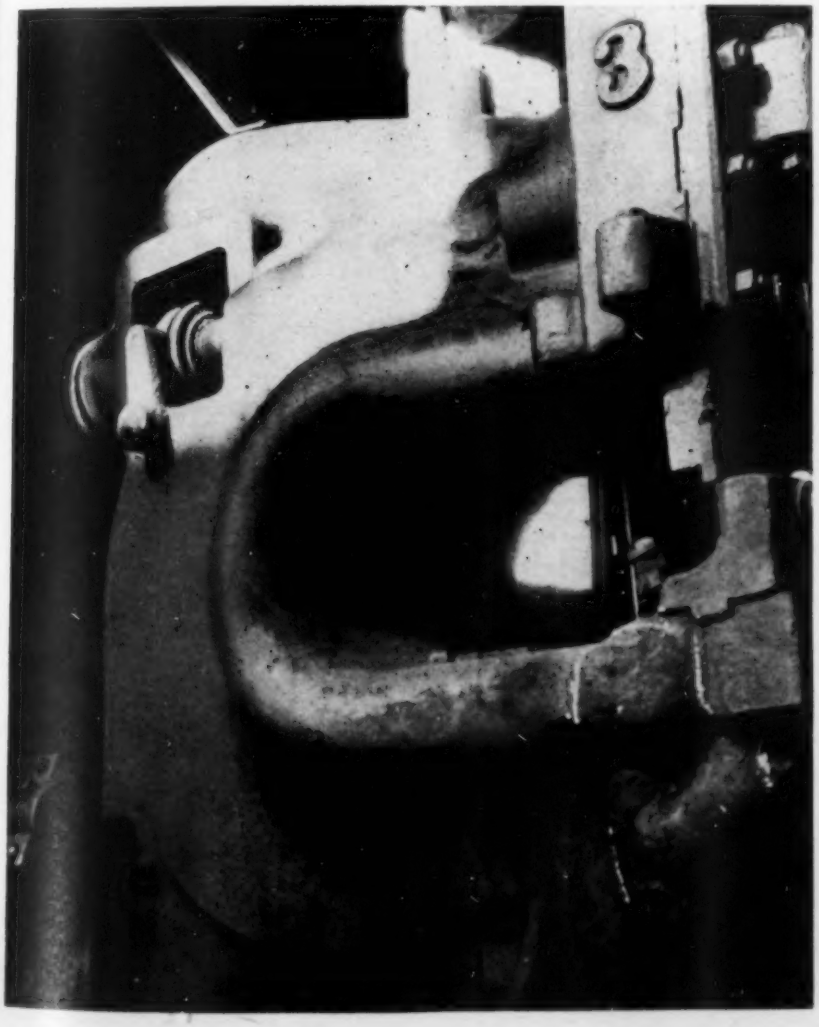


means. Molten cast iron of a suitable composition is taken from a cupola and poured onto the weld area or crack in such a manner as to heat the casting in the entire area to be repaired.

Pouring is done slowly and continued for one to two minutes, allowing the excess metal to run out through the flow-off in the dam. Pouring is then interrupted for a short time, perhaps a minute, to allow the heat to soak into the casting, and then resumed for a short time. This alternate pouring and waiting is continued until the area to be repaired is heated above the melting temperature and proper bonding between the added metal and the metal of the original casting takes place.

The metal used for pouring should be at a temperature of about 2800 F, and the pouring should not be done after the temperature drops below 2500 F. Otherwise, too much molten metal is required to heat the casting to the fusion temperature. After the pouring is completed, the casting is covered to permit slow cooling.

Fig. 13—Cast iron punch press frame repaired by thermit welding.



Added Metal Properties

Physical tests of metal produced by burning show that the properties of the added metal are practically the same as those of cast material of the same analysis. The microstructure of the added metal is similar to that of gray iron except that the added metal has a finer grain structure.

Figure 14 shows junction of burn metal at right and parent metal at left. If the process is properly carried out there will be no hard spots in the added metal or in the casting adjacent to the repair.

The burning process results in a repair which is similar in composition to the original casting. Therefore, the repair matches the casting in color after machining. Since the burning process requires molten metal from a cupola, its use is generally limited to repairs made in the foundry.

After completing the repairs with any of the foregoing processes, the question often arises, "Is stress-relieving necessary?" The answer depends upon several factors. Structures which are welded with the "cold" arc welding process to prevent distortion, such as the turbine cylinder discussed, cannot be stress-relieved because to do so would cause distortion. Rigid structures which are welded with a high local

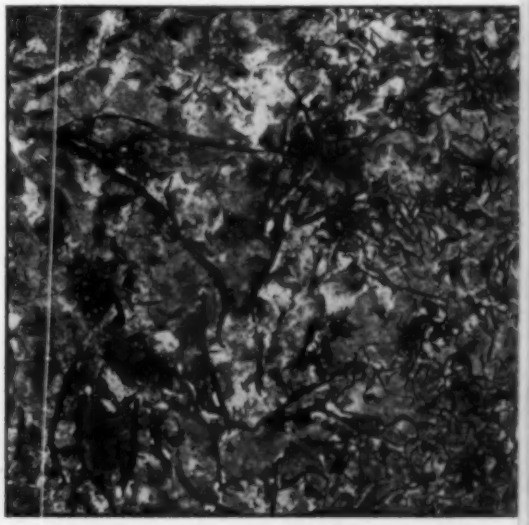


Fig. 14—Junction of "burn" metal (right) and gray iron. ×100.

preheat and which require high strength should be stress-relieved if possible.

If a structure is not rigid and the entire heated area can expand and contract freely, then stress-relieving is not necessary. Many structures repaired in the field cannot be stress-relieved; therefore, the metallic arc process and a moderate preheat are used even though the resulting welds do not have a strength equal to that of the casting.

An example of a large cast iron structure which was successfully repaired by welding is the four-section, 252-in. water box shown in Fig. 15. While machining the flanges on these castings, four cracks were found at the outer corners of the castings comprising the water box.

Three of the cracks extended

through the thickness of the flange, and one of these also extended about 14 in. into the side wall along the inlet opening. In the faces of the flanges were a number of shrink cavities, some being as great as 2 in. in depth.

Method Selected

It was decided to use bronze welding for the cracks in the flanges, since this would insure repairs of high strength and color match was not important. Inasmuch as the cracks were at the corners of the sections the expansion due to high preheat could take place without causing high stresses in the casting. Figure 16 shows the crack which extended into the side wall after chipping for welding. Figure 17 shows the outside and Fig. 18 the inside of the completed bronze weld.

The shrink cavities in the flanges were welded with the metallic arc, using monel electrodes and a low preheat. Threaded studs of ⅛-in. diameter for the smaller cavities and ¼-in. diameter for the larger cavities were used.

This method was chosen because these repairs are not subjected to high stresses and, therefore, metallic arc welds have sufficient strength. Also, the preheat required for the other welding processes would have produced high stresses in the side wall. Figure 19 shows one of the cavities after chipping and studding.

Stress Relief After Repairs

The sections were stress-relieved after all the repairs had been completed. No difficulty was encountered in machining these sections and no trouble has developed in service.

Use of welding to repair these castings was of prime importance in several respects. Without welding it would have been necessary to re-

Fig. 15—(below)—Four-section cast iron water box on 40-ft mill for machining.

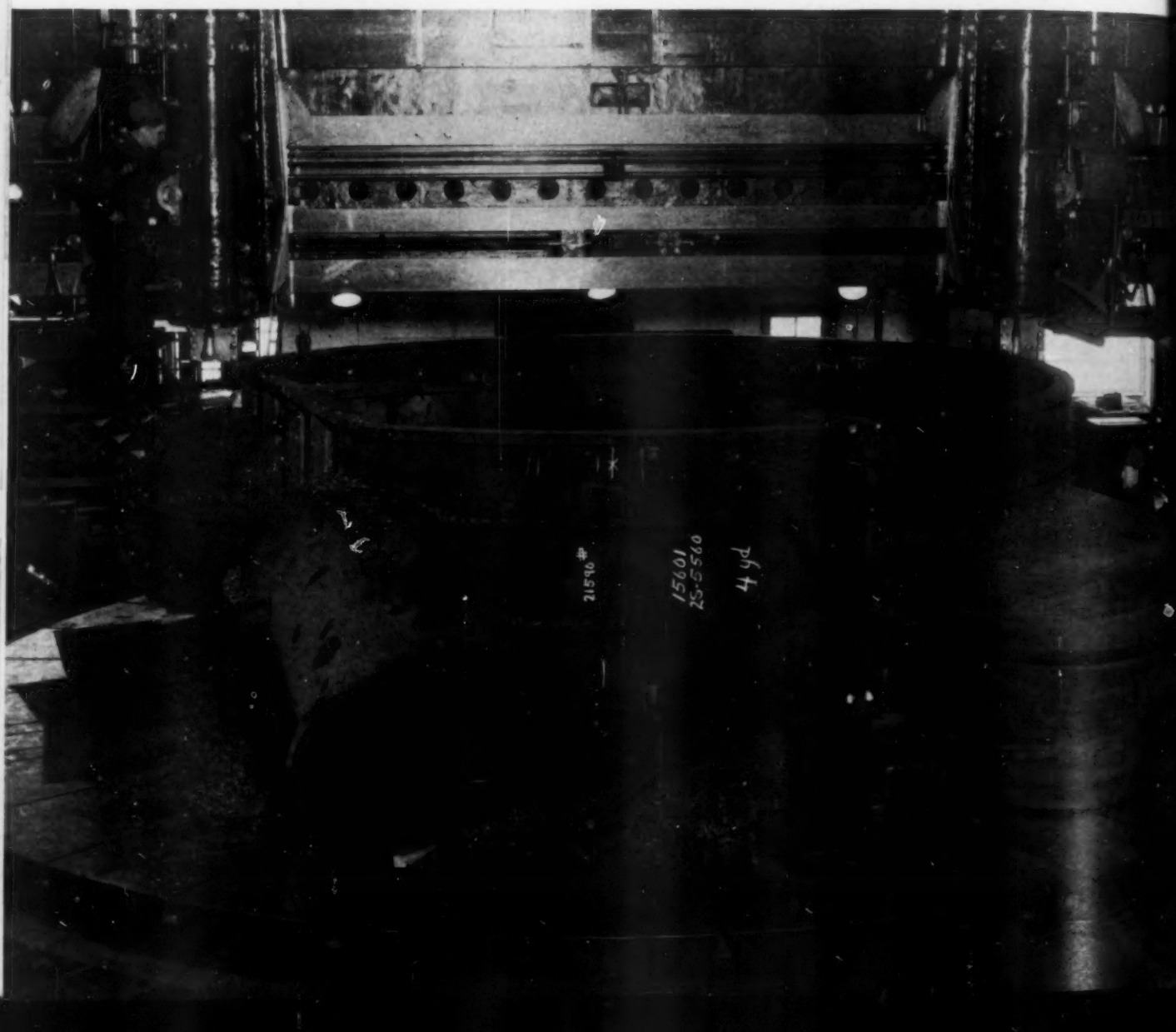
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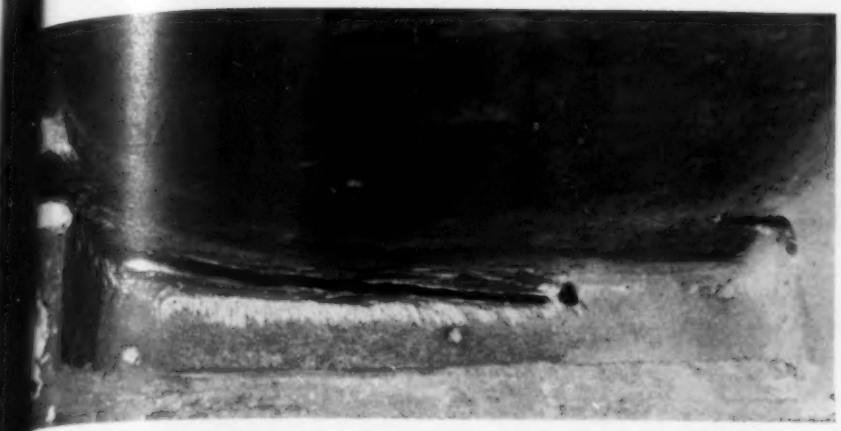
Fig. 16—Crack in flange and side wall of water box. Chipped for bronze welding.

Fig. 17—Bronze repair weld of crack shown in Fig. 16.

Fig. 18—Bronze weld shown in Fig. 17 viewed from inside of water box. Inlet opening appears at right of repair.

Fig. 19—Cavity in water box flange chipped and studded for metallic arc welding.











place three of the sections at considerable expense. More significant, however, is the fact that the replacement would have tied up foundry facilities needed for war production and delayed the installation of equipment which was urgently needed.

Acknowledgment

The author wishes to acknowledge with thanks the photographs furnished by the Metal and Thermit Corp., and the assistance of his colleagues, particularly H. O. Quartz, W. A. Hambley, and G. W. Leupold, in furnishing information and preparing specimens.

DISCUSSION

Chairman: E. C. JETER, Ford Motor Co., Dearborn, Mich.

Co-Chairman: JOHN CROWE, Air Reduction Sales Co., New York.

CO-CHAIRMAN CROWE: How do you clean out the castings for welding? In general, what preparations are made for welding?

L. J. LARSON[1]: In preparation of castings for welding we generally remove the metal by chipping. It is possible to remove most of the metal by "burning out" with a metallic arc. In some cases the groove may be burned out with a metallic arc and then chipped to finish the preparation. In view of the value of the casting we generally consider the extra time required for chipping is well worth while in order to obtain the best preparation for welding.

CO-CHAIRMAN CROWE: Do you know whether anyone is removing the defects with the oxygen jet? If not, why not? Is that not a practical means of removing defects?

MR. LARSON: It is my understanding that oxygen cutting and gouging do not work successfully on cast iron because cast iron is not readily oxidized.

CHAIRMAN JETER: Discussing welding of castings in a general way, there has been much misunderstanding between the foundries and the consumers about welding. As a result, some of the foundries perhaps tried to hide whatever repairs they might make to castings. It would have been better to bring it out in the open and show what was being done.

The A.F.A. is trying now to publish the facts about welding of castings so that foundries and consumers know what is being done and understand each other. The S. A. E. War Engineering Board, together with the Ordnance Department, have accomplished a great deal in this direction, but it is up to the A.F.A. to carry on and make available to every foundry and consumer information as to correct procedure for welding and repairing of castings.

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[1] Allis-Chalmers Manufacturing Co., West Allis, Wis.

BEHAVIOR OF MOLDING MATERIALS IN "OWN ATMOSPHERE" AT ELEVATED TEMPERATURES

Harry W. Dietert
and
Robert L. Doelman
President and Foundry Consultant, Respectively
Harry W. Dietert Co.
Detroit

A PRACTICAL CLASSIFICATION of gas activity may be made by recognizing that gas may be oxidizing, neutral, and reducing.

Each type of gas may materially influence the functioning of a mold or core surface during the pouring, solidifying, and cooling stages of a casting, since the gas may influence the hot strength, retained strength, refractoriness and rate of collapsibility of the mold surface.

In the beginning of the investigation, elevated temperature properties of sand and core specimens were studied with the three mentioned types of gases. However, it was soon realized that such atmosphere tests could be far removed from the exact gas in which the sand or core was performing its work in a mold. A mold contains a great variety of different gases in varying concentrations; it is thus an endless task to endeavor to subject the test specimen to the proper mixture of gases by adding the gases to the test furnace.

Assuming that the quantity of gases given off by the metal in the mold is small, the authors concerned themselves with gases given off by the molding materials. Under these conditions, the gas present then comes mainly from the various ingredients of the mold surface; and thus the sand surrounds itself with its own gases, which the authors have termed "own atmosphere."

In a mold the gases generated by the ingredients of the mold surface layer are held within the mold cavity, along the metal-mold interface and also in backing sand. A goodly portion of the gases escape but, as more gases are generated continuously, the gas composition changes. During the pouring operation, the gases are diluted with air that is within the mold cavity and sand pores. As the oxygen of the air is quickly consumed, the gases formed by the mold material become the major constituents of the mold atmosphere.

In order to reproduce the mold-like conditions in a dilatometer furnace as nearly as possible, the test specimen of sand or core is placed within a metal or ceramic hood. Three types of hoods used are metal, quartz and ceramic post hood. These are illustrated in Fig. 1.

A metal hood made of carbon-free nickel works well for temperatures below 2000° F., and also for hot strength, retained strength and collapsibility tests. For temperatures above 2000° F., a platinum hood is required. Since the life of a platinum hood may not exceed 50 tests under a reducing atmosphere as generated by many sands, this is expensive. The actual cost of a platinum hood is $100.

For retained strength tests where no load is placed on the test specimen at elevated temperature, the quartz hood suspended in the dilatometer furnace proved satisfactory.

A ceramic hood formed by a special top post of the dilatometer will undoubtedly be perfected, so that it will be economical and easy to use.

When a sand or core specimen is placed within a hood, some air surrounds the sand as it does in a mold. The confinement of the hood traps the gases from the test specimens. Excess gases may escape from bottom of hood. Thus, each sand mixture tested is allowed to create its own atmosphere as it does in a mold.

All core and molding sands were prepared in accordance with A.F.A. specifications. For elevated temperature testing, a test specimen of 1⅛-in. diameter and 2-in. length was used. This specimen was rammed with three drops of a 7-lb. rammer weight falling 2⅝ in. in a floating specimen tube producing double-end ramming.

A high temperature testing furnace employing a 2-in. inside diam-

Developments in foundry art have revealed that molding materials at the surfaces of a mold or core will behave differently in different sections of a mold or in different molds where identically prepared sand and molding methods are used. This different behavior of mold or core surface cannot always be ascribed to difference in composition or physical properties of the material nor in their mechanical placement, but in a possible mold surface failure due to a peculiar gas atmosphere. The gases that surround the mold face, while it acts as a container for molten metal and while the metal is solidifying and cooling, will determine to a large extent whether or not the cast surface is of the desired quality.

Presented at a Sand Research Session of the Fiftieth Annual Meeting of the American Foundrymen's Association at Cleveland, May 8, 1946.

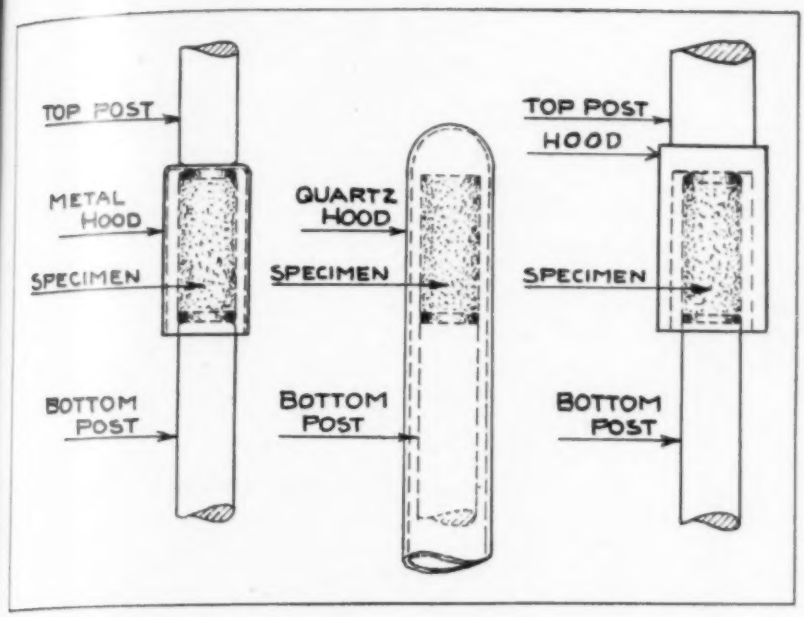


Fig. 1—Various types of hoods used to produce own atmosphere around sand or core specimen under test; (left) metal hood; (center) hood made of quartz; (right) a ceramic hood formed at the lower end of the top refractory post.

eter heating element furnace was used for all high temperature testing.

Core Collapsibility. A core which derives a substantial portion of its strength from a binder that will oxidize at elevated temperatures in the presence of oxygen will lose its strength so rapidly in an oxidizing atmosphere that one would conclude that no metal could be cast against its surface.

In a mold, as soon as heat from the metal strikes the core, volatile gases will be freed from the core, and these will consume the available oxygen. After the initial oxygen is consumed, the core surrounds itself with its own atmosphere which retards the rapid oxidation, allowing the core to slowly disintegrate.

When a core specimen is placed in a dilatometer furnace without a hood the atmosphere is oxidizing, and the core specimen will collapse quickly at high temperatures should it derive a substantial portion of its strength from combustible binders. Under identical conditions, however, with a hood over the core specimen, the core will show a much slower rate of collapsibility.

As an example, Table 1 shows that core collapsed in 43.7 sec. in an oxidizing atmosphere, and in 121 sec. in its own atmosphere when subjected to a temperature of 1500° F. The test data of core collapsibility under own atmosphere correlates well with foundry experience.

Own atmosphere of most cores form a protective shield at elevated temperatures and avoids a too early collapse of the surface of the core.

Table 1

CORE COLLAPSIBILITY

(90 psi. at 1500° F.)

Mixture per cent	*Collapsing Time, sec.* — *Oxidizing Atmosphere*	*Own Atmosphere*
Michigan City Sand, 87.0 Linoil, 1.5 Cereal Flour, 1.0 Silicon Flour, 10.0 Bentonite, 0.5	43.7	121.0

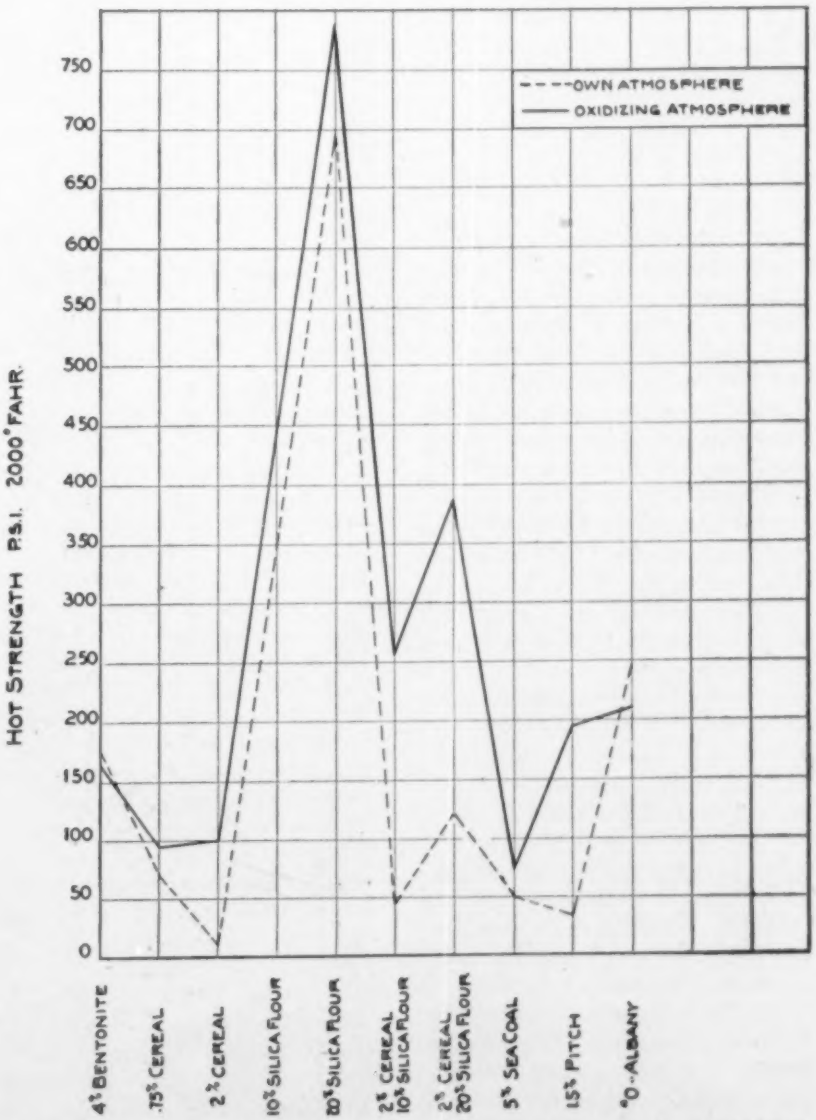


Fig. 2—Effect of oxidizing and own atmosphere on hot strengths at 2000° F., 12-min. soaking.

Table 2

EFFECT OF ATMOSPHERE ON CORE HOT STRENGTH

(2500° F.—12 min.)

Core Mixture No.	Hot Strength, psi.		Binders		Sand	
	Own Atmosphere	Oxidizing Atmosphere	Type	Per Cent	Type	Per Cent
1	3	0	Core Oil	1.0	Lake Sand	99.0
2	5	4	Oil	1.2	Lake Sand	70.0
			Cereal	0.5	Bank Sand	28.3
3	7.8	9	Oil	1.5	Lake Sand	87.0
			Bentonite	0.5	Silica Flour	10.0
			Cereal	1.0		
4	250	9	Resin	3.0	Lake Sand	15.0
			Cereal	0.5	Heap Sand	81.5
5	55	3	Pitch	2.0	New Sand	49.0
					Heap Sand	49.0

Table 3

SAND FINENESS TEST

(No. 17 Ottawa, washed and dried)

Mesh No.	Retained per cent
6	—
12	—
20	—
30	—
40	6.6
50	35.2
70	34.8
100	19.2
140	3.8
200	0.4
270	—
Pan	—
Fineness No.	51.26

Core Hot Strength. A reasonably close relation exists between core collapsibility and hot strength. In the core collapsibility test, the specimen is preloaded with a predetermined load. For example, an 85-psi. load is applied and the time in minutes that the specimen is able to support at some chosen elevated temperature is recorded; for example, 1500° F.

In the hot strength test, the specimen is first allowed to remain (soak) in an elevated temperature zone; for example, 12 min. at 2500° F., and then crushed. The load in psi. is recorded as the hot strength.

Many of the oil, resin or cereal bonded cores will disintegrate before termination of the 12-min. soaking period when the atmosphere is oxidizing. The hot strength under such conditions would be recorded as zero.

All cores possess a hot strength when they are heated in their own atmosphere; for example, when a hood is employed to reproduce mold conditions.

Test data which the authors secured on core mixtures, as used by the A.F.A. Subcommittee on Physical Properties of Iron Molding Materials at Elevated Temperatures, are good examples to show the effect of oxidizing atmosphere as compared to own atmosphere on hot strength of cores. Test data are tabulated in Table 2. It may be noted that the hot strength at 2500° F. in own atmosphere is much greater than that in oxidizing atmosphere in all cases except one. For example, core mixture No. 4 possesses 250 psi. hot

Table 4

HOT STRENGTH OF MOLDING SANDS

Sand Mixture		2000°F. 12 Min.		2500°F. 12 Min.	
Component	Per Cent	Oxidizing Atmosphere	Own Atmosphere	Oxidizing Atmosphere	Own Atmosphere
No. 17 Ottawa, W. & D.	96.0				
Western Bentonite	4.0				
Moisture	2.0	164	175	7.0	7.25
No. 17 Ottawa, W. & D.	95.25				
Western Bentonite	4.0				
Cereal	0.75				
Moisture	2.6	94	72	7.0	10.0
No. 17 Ottawa, W. & D.	84.0				
Western Bentonite	4.0				
Cereal	2.0				
Moisture	2.6	101	8	4.6	3.0
No. 17 Ottawa, W. & D.	86.0				
Western Bentonite	4.0				
Silica Flour	10.0				
Moisture	3.75	488	322	10.3	16.4
No. 17 Ottawa, W. & D.	76.0				
Western Bentonite	4.0				
Silica Flour	20.0				
Moisture	5.0	783	696	152	137
No. 17 Ottawa, W. & D.	84.0				
Western Bentonite	4.0				
Cereal	2.0				
Silica Flour	10.0				
Moisture	4.6	255	45	66	54
No. 17 Ottawa, W. & D.	74.0				
Western Bentonite	4.0				
Cereal	2.0				
Silica Flour	20.0				
Moisture	5.5	382	118	220	199
No. 17 Ottawa, W. & D.	91.0				
Western Bentonite	4.0				
Sea Coal	5.0				
Moisture	2.5	52	69	8.5	25
No. 17 Ottawa, W. & D.	94.5				
Western Bentonite	4.0				
Pitch	1.5				
Moisture	2.0	191	38	7.0	10.8
No. 0 Albany Sand	100.0				
Moisture	9.0	211	243		

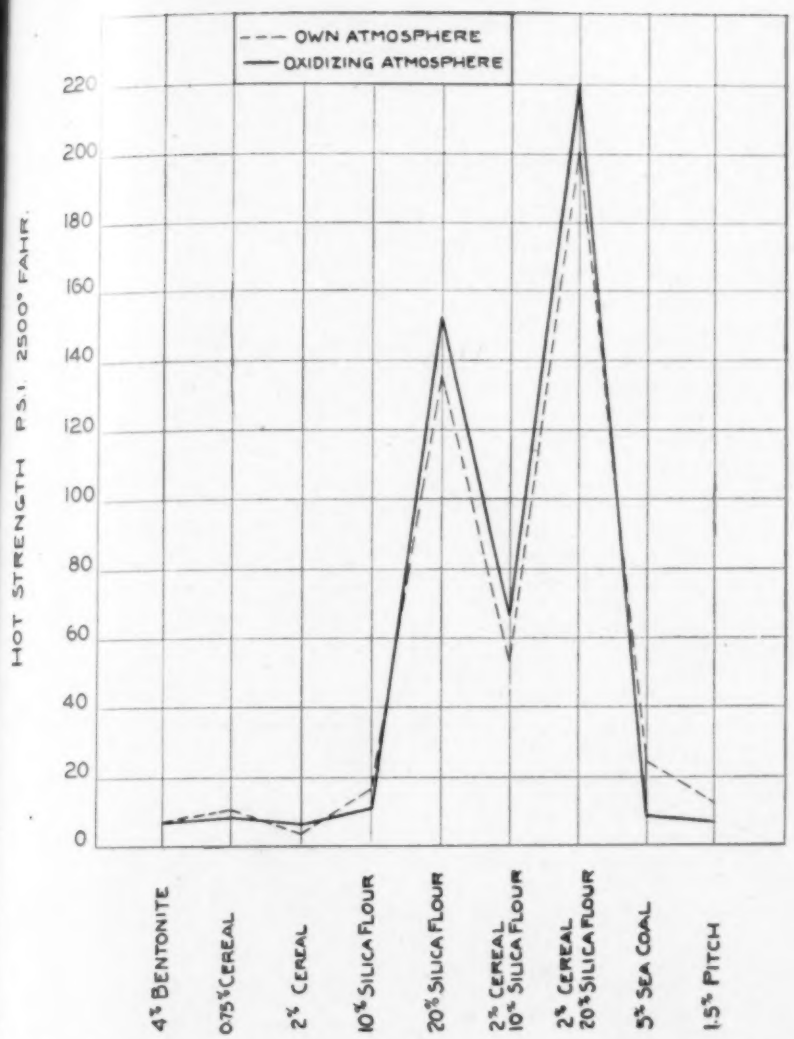


Fig. 3—Effect of oxidizing and own atmosphere on hot strengths at 2500° F., 12-min. soaking.

strength at 2500° F. in own atmosphere as compared with 9 psi. in oxidizing atmosphere. The bond of the No. 4 core mixture is 3.0 per cent resin and 0.5 per cent cereal binder (corn).

An important fact to note is that in the case of core mixture No. 3 the hot strength in reducing atmosphere is 7.8 psi., and in an oxidizing atmosphere, 9.0 psi. This higher hot strength at 2500° F. in oxidizing atmosphere could be caused by the presence of silica flour, which possesses greater strength in oxidizing atmosphere.

Hot Strength of Molding Sands. The majority of the hot strength tests on molding sands for this investigation were made on bentonite bonded sand, using 4 per cent of western bentonite and a No. 17 Ottawa washed and dried sand of the fineness shown in Table 3. This synthetic sand was prepared with 0.75 per cent and 2.0 per cent cereal binder. Other mixtures contained 10 per cent and 20 per cent silica flour with and without cereal binder. This sand was also prepared with 5 per cent sea coal and 1 per cent pitch.

Only one natural sand (No. 0 Albany) was used in this study. The test results are tabulated in Table 4 for 2000 and 2500° F. using a 12-min. soaking period. A platinum hood was used to obtain own atmosphere.

A graphical presentation of this data is shown in Fig. 2. A comparison of hot strength at 2000 and 2500° F. shows that the effect of atmosphere is of greater magnitude at 2000 than at 2500° F.

Referring to Fig. 2 and Table 4, note that the hot strength at 2000° F. in own atmosphere is appreciably lower than in oxidizing atmosphere when cereal binder, silica flour, or pitch is present, while it is greater when sea coal is present. At a test temperature of 2000° F., the own atmosphere within the test specimen undoubtedly is reducing and the interior remains unsintered.

Silica sand bonded with western bentonite and the natural bonded Albany sand showed a higher hot strength in own atmosphere than in oxidizing atmosphere at both 2000 and 2500° F. These sands bonded by clay substance prefer their own atmosphere. When secondary bonds such as cereal, silica flour, and pitch are present in appreciable percentages, they do not prefer their own atmosphere at 2000° F.

Sand containing pitch is quite soft in own atmosphere at 2000° F., while sand containing sea coal is firm, showing a greater hot strength in own atmosphere. When the test temperature is raised to 2500° F., the difference found between hot strength in oxidizing and reducing atmosphere is interesting (Table 4 and Fig. 3).

When the molding sand is bonded principally with western bentonite, pitch, sea coal, nominal percentages of cereal or silica flour, then the hot strength at 2500° F. in own atmosphere is greater than in oxidizing atmosphere; but, as the additions are increased abnormally, e.g., cereal to 2.0 per cent or silica flour to 20 per cent, then the hot strength at 2500° F. reduces in its own atmosphere.

This reduction in own atmosphere in the case of high silica flour addition does not appreciably reduce the over-all great increase in hot strength due to silica flour addition. In the case of cereal binder, an abnormally high addition might prove detrimental.

Value of the sea coal addition as a means of improving the toughness of a mold face is well demonstrated by the large increase in hot strength in own atmosphere.

Referring to Table 4, it may be noted that the addition of 2 per cent cereal binder to a sand containing either 10 or 20 per cent silica flour increases the hot strength at 2500° F. for oxidizing or own atmosphere. At 2000° F. the hot strength is decreased when cereal is added.

More research work must be done

before one can draw a complete set of conclusions on the effect of atmosphere on the behavior of molding materials at elevated temperatures.

From the test data at hand, the following effects have been indicated. Conclusions are grouped separately for molding sand and cores.

Molding Sand. For western bentonite bonded sands, and also natural Albany sand, the hot strengths were found to be higher in own atmosphere than in oxidizing atmosphere. The same holds true for sands where 0.75 per cent cereal binder was added.

The addition of cereal binder in high percentage, e.g., 2.0 per cent to a molding sand causes the hot strength at 2000 and 2500° F. to be lower in own atmosphere than in oxidizing atmosphere.

An addition of 20 per cent silica flour caused the hot strength at 2000 and 2500° F. to be lower in own atmosphere than in oxidizing atmosphere. A 10 per cent addition showed an increase in hot strength at 2500° F. in own atmosphere.

An addition consisting of a combination of silica flour and cereal binder causes the hot strength of the sand to be higher in its own atmosphere than in oxidizing atmosphere at 2500° F.

Sea coal in molding sand causes the hot strength to be materially higher in own atmosphere than in oxidizing atmosphere, which correlates well with foundry experience.

Pitch in molding sand will show a higher hot strength at 2500° F. in own atmosphere than in oxidizing atmosphere, and a lower hot strength at 2000° F. acting as a cushioning agent for the mold face.

Molding sands, with or without normal secondary binder addition in conservative amounts, generally prefer their own atmosphere for elevated temperatures such as 2500° F.

Core Sand. The hot strength of cores bonded with combustible binders, such as oil, is increased by own atmosphere and reduced by oxidizing atmosphere.

Addition of silica flour to a core mixture seems to cause the hot strength to be lower in own atmosphere than in oxidizing atmosphere. The difference is of no great practical value, since silica flour additions step up the hot strength so greatly.

Addition of pitch to a core mixture causes the hot strength at 2500° F. to be much higher in own atmosphere than in oxidizing atmosphere.

Own atmosphere is recommended when making the retained strength test to evaluate core binders for ease of knockout.

Further research on the subject of gas activity control will undoubtedly help to a better understanding of the behavior of molding material at elevated temperatures in molds and to uncover some practical applications of mold atmosphere control.

DISCUSSION

Chairman: B. H. BOOTH, Carpenter Bros., Inc., Milwaukee, Wis.

Co-Chairman: R. E. MOREY, Naval Research Laboratory, Washington, D. C.

DR. H. RIES: How many tests do those figures represent on the average?

MR. DIETERT: The data shown on our tables was the average of at least 25 readings. We tested at least five different mixtures, taking five test readings per mixture. On some of the test data shown we have more than five mixtures tested at different times over a longer period of time.

DR. RIES: Did you get much variation?

MR. DIETERT: Yes, we did obtain variations. Some of these variations were due to different operators. We found that the mixing of the sand was quite critical. It had to be done precisely, and the moisture content in the sand had to be held very closely before you could duplicate results. We were very conscious of the fact that the green sand properties had to be duplicated and the preparation of the sand had to be done carefully and accurately. We could do a nice job of reproducing our results, provided the platinum hood did not have a leak in it. Just a little pinhole would throw us off.

DR. RIES: I noticed some of the test figures for hot strength were close together, and I wondered if the different tests might overlap. One of the difficulties with the dilatometer is reproducibility of results, and until this matter is satisfactorily settled, we must be careful what interpretations we make.

MR. DIETERT: There might have been some overlapping over the earlier tests that we made. On recent tests we found that taking the average results secured by several different operators' work, the same trend was exhibited.

DR. RIES: You will not get that much difference in specimens.

MR. DIETERT: Yes, I believe you can if you would consider the many variables that may creep in due to mixing of the sand and in ramming the specimen.

MR. BREWSTER[2]: Would you care to elaborate on some of the more glaring mistakes that can be made in preparing sand?

MR. DIETERT: Yes, there are many. One of the greatest mistakes we make in test work is that we never lay aside sufficient amount of sand to complete an investigation. We may lay aside 200 lb. of sand in two sacks. We assume that the sand in the two sacks is the same because both have the same trademark and come from the same producer. We do not mix the two sands. We use up the first sack of sand and then open the second one. The sand might have the same fineness number, but that means very little. Just a slight difference in one sieve size or a difference in the grain distribution will account for differences in test results.

After using up those two sacks, we order more. The new sand may be quite different. We then have test data which we are unable to use because we cannot duplicate the sand used for previous tests. Therefore the fineness of sand is of first importance. Make certain that you have a sufficient quantity of identical sand to complete the entire testing project.

The same thing is true of the type of bond used. We may decide on bentonite. We may keep it in a 2-qt. jar or in a gallon container. The container is left open. The clay absorbs moisture. We use the same weight of bentonite but do not get the same amount of bentonite in the sand mixture due to variation in moisture content. Soon the 2 qt. of bentonite is used up and we have to get more from another source. The new bentonite will probably have different bonding characteristics.

The amount of sand and bond used in each batch must be exactly the same. We use dummy batches to get the mixing unit conditioned. The adjustments of the mulling wheels and the rapidity with which one dumps the mixture are important. If one just dumps the mix and lets the mixer run on a little layer of sand in the bottom of the pan this layer of sand will have more mixing and will be different from the original batch.

Following all A.F.A. test procedures of mixing and timing the batch is important. It makes a difference if the muller is not kept closed. Wait 2 min. after stopping muller with a dry mix before removing the muller cover.

A bentonite mix or even a fire clay mix must be weighed accurately in the preparation of small batches. Condition of prepared sand storage, such as temperature, makes a difference. We have experienced water migration in the sand, also condensation of the water on the surfaces of the container. These cause errors in test results.

The water will migrate differently in sand that is packed tightly as compared to a sand that is packed loosely. Whether the sand was riddled before placing in container and whether it was riddled before we used it again makes a difference. Frequently before sand is removed from the container, it has been handled a lot and is jolted down. The sand on the bottom is packed together. Although it may weigh 200 grams on the balance, the lumps when put into the specimen con-

tainer will cause a non-uniform specimen.

The problem of preparing the ram specimen is a big one, but no greater than the problem of ramming in the foundry. In the foundry the sand is allowed to roll into the flask from one side, or the sand is permitted to pile up, in the center of the flask. The sand which is rammed first or which is piled high is rammed harder and uniform mold hardness is not obtained. The same thing applies in ramming the A.F.A. 2-in. and the 1⅛-in. specimen.

L. A. BATEMAN[3]: When testing Albany sands in their own atmosphere, do you attain the characteristic spalling action which occurs when these sands are tested in an open or oxidizing atmosphere?

MR. DIETERT: I did not notice that because we could not see the specimen. It was underneath the hood.

MR. BATEMAN: Our own observations of spalling were made in an open atmosphere. However, we did use the following method of testing specimens in their own atmosphere. A porous alumina extraction cup was cut off to a length of 1⅜ in. This cup was used to cover the specimen during the dilatometric test.

In an open atmosphere, Albany sands spall badly during the 2-min. test period. When the cup is placed over the specimen, allowing the specimen to form its own atmosphere, spalling is completely retarded. We are at a loss to interpret these results with plant practice.

In regard to spalling, we might add that having recently finished testing 19 different natural molding sands, we have found that all Albany sands, in fact all New York and Ohio sands, exhibit a characteristic spalling during the 2-min. spall test.

CHAIRMAN BOOTH: Do you mean in the oxidizing atmosphere?

MR. BATEMAN: Yes, with the porous cup over the specimen, spalling is retarded to practically nothing.

D. C. WILLIAMS[4]: I would like to ask Mr. Bateman where is the controlling thermocouple on the 2-min. test?

MR. BATEMAN: It is outside the cup. I might add that we do extend the time of that test to 5 min. rather than 2 min. In other words, we use the color of the sand as a guide to temperature. We ran so many tests on Albany sands that we feel we can determine temperature visually by the color of the specimen.

MR. WILLIAMS: What is the material the cup is made of?

MR. BATEMAN: Aluminum oxide. We tried to measure the permeability and it can be measured on the machine. It runs from 40 to 50 for various cups, which is somewhat the permeability of our Albany sand.

H. E. HENDERSON[5]: Have you found that any particular clay or combinations of clays in your test seem to aggravate the spalling condition?

MR. DIETERT: I do not believe that I can name one. Sands which contain very low percentages of clay substance by volume, generally have a tendency to spall more than sands containing a larger volume of clay, because clay acts as a cushion medium, retards flowability, allows more void spaces between the sand grains and clay, and also has a tendency to decrease in volume when heated, while silica sand grains grow in size upon heating, thus the clay would somewhat minimize the over-all growth of the sand specimen.

I think that clay volume probably has more effect on spalling than the specific type of clay. You do have certain clays that may have more flowability than others, and if that is the case the sand with the higher flowability would spall more than the one with the lower flowability. Clays with high organic substance content may show less spalling due to combustion of the organic materials.

MR. HENDERSON: That corroborates my experience.

W. A. SPINDLER[6]: Would the density be lower on a specimen with higher clay content?

MR. DIETERT: I do not know exactly. I cannot recall figures to answer that.

A. E. MURTON[7]: Our experience is that the more of certain clays you add, the less the density is. The flowability decreases.

Referring to that cup, would there be a severe shock applied to the specimen? If you could not put the cup over it and raise it into the furnace rapidly enough, it is not going to have the severe shock and to make it spall.

MR. BATEMAN: We preheat our cup in the furnace 15 min. before test. While the dilatometer furnace is brought down one operator places the test specimen on the post while the other raises and quickly drops the cup over the specimen. We have noticed no cracking.

MR. DIETERT (authors' closure): In the research tests discussed in this paper we placed the furnace temperature controlling thermocouple inside the hood which surrounded the sand specimen. Thus the temperatures shown are actual temperatures at the outer surface of the sand specimen.

It is a known fact that temperature rises more slowly within a specimen when a reducing atmosphere is present than when an oxidizing atmosphere is present. When a hood surrounds a sand specimen and reducing gas is generated by the sand it is true that the sand specimen receives less shock under like temperature, therefore less spalling may be expected. In a mold the sand may be compounded to generate a reducing gas which would reduce or eliminate spalling.

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[2]Dow Chemical Co., Bay City, Mich.
[3]National Sanitary Co., Salem, Ohio.
[4]Cornell University, Ithaca, N. Y.
[5]Macaulay Foundry, Berkeley, Calif.
[6]University of Michigan, Ann Arbor, Mich.
[7]Metallurgy Laboratories, Dept. of Mines and Resources, Ottawa, Ont.

CRUCIBLE FURNACE MELTING COMBUSTION FACTORS

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Joseph Dixon Crucible Co.
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ALL METALS AND ALLOYS in the molten state have an affinity for certain gases, some metals more so than others. Absorption of gas may take place during melting, particularly if the metal comes into direct contact with the products of combustion in the melting furnace. It may also occur at any point where moisture comes into direct contact with the molten metal, such as moisture entrapped in the ingots or charging metal, in the air on a particularly humid day, or in a green sand mold.

When moisture comes into contact with molten metal a disassociation, or breakdown, of the moisture (H_2O) occurs, and the metal dissolving the hydrogen (H_2) thus liberated. The amount of pickup depends to a large extent on the type of metal and the temperatures involved. The affinity for gas increases as the temperature of the metal increases.

Another important source of gas is that originating from improperly vented cores. Organic core binders break down and liberate combustible gases, as evidenced by the fact that these gases ignite and burn at the core vents when the molten metal comes into contact with the cores. Quite frequently a casting made in a mold containing numerous cores will show pinhole porosity, whereas one made from the same melt in a mold without cores shows no porosity.

Improper combustion during the melting, particularly as it relates to the melting of non-ferrous metals in gas- or oil-fired crucible furnaces of the stationary or tilting type, is unquestionably one of the most serious contributors to the "gassing" of metal. Therefore, too much emphasis cannot be placed on the importance of properly proportioning fuel (gas and oil) and air in order to obtain not only maximum furnace efficiency but quality castings as well.

Air:Fuel Ratio

In the typical crucible furnace for the melting of brass and bronze, the crucible sets on a pedestal, the flame from a tangential burner swirls around the crucible, and the products of combustion vent through the furnace cover hole. The metal and crucible are entirely surrounded by the flame and the gases, making it apparent that a proper air: fuel ratio setting of the burner is important in order that complete combustion take place in the furnace. If combustion is complete no reducing gases will be present, particularly H_2 (hydrogen) and CO (carbon monoxide), to be dissolved by the metal. From the standpoint of crucible life and metal quality, the best burner ratio setting is one producing a neutral atmosphere inside the furnace.

Maintaining a proper fuel to air ratio and combustion rate in crucible melting is an important factor in eliminating conditions which may contribute to the "gassing" of metal and consequent scrap losses due to pinhole porosity.

Combustion processes are classified into three categories, the first of which is a condition where an insufficient quantity of air is used to burn the fuel, resulting in products of combustion containing no O_2 (oxygen), but containing the reducing gases H_2 and CO. This is referred to as incomplete combustion (reducing atmosphere), and is mainly responsible for the "gassing" of metals due, particularly, to H_2 pickup.

The second condition is where just enough air is used to burn the fuel completely, without excess, resulting in products of combustion containing practically no O_2 or H_2. This represents perfect combustion, or a neutral atmosphere. The third condition is where an excess of air is used to burn the fuel, producing flue products that are oxidizing, containing O_2 with an absence of CO and H_2. This is referred to as an oxidizing atmosphere, a condition not conducive to long crucible life due to oxidation of the crucible.

Manufactured gas (city gas) is quite widely used as a crucible furnace fuel, but varies considerably in analysis depending on location and method of manufacture. However, for any given locality the analysis is fairly constant. Curve "A" (Fig. 1) is typical of a manufactured gas and illustrates the variation in flue gas constituents as the quantity of air used for combustion is changed. For

Presented at a Brass and Bronze Session of the Fiftieth Annual Meeting, American Foundrymen's Association, at Cleveland, May 8, 1946.

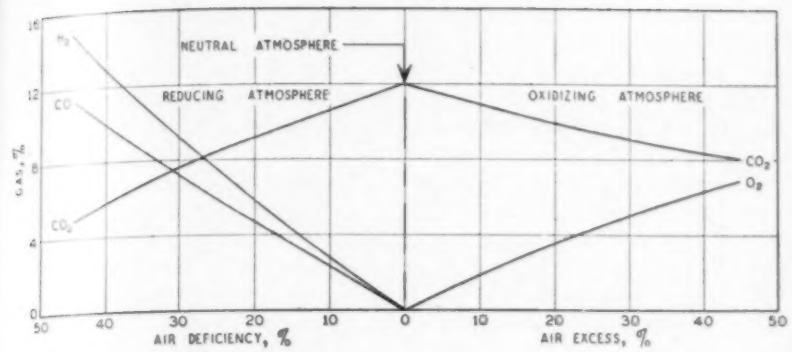


Fig. 1—(Curve A) Ratio of constituents in flue gas of manufactured gas-fired crucible furnace.

example, if 20 per cent less air is used than is required to complete combustion, an analysis of the products of combustion would show approximately 5 per cent CO, 6 per cent H_2 and 9 per cent CO_2 with no O_2, and is definitely reducing. This is not considered good melting practice from the standpoint of combustion efficiency or gas pickup by the metal. On the other hand, if 20 per cent excess air is used to burn the gas the products of combustion will show approximately 3½ per cent O_2 and 10 per cent CO_2, with no CO and H_2, definitely oxidizing.

Neutral Flame

This, likewise, is not considered good melting practice, resulting in excessive oxidation of the metal, fuel waste and oxidation of the crucible. Therefore, a neutral flame, the closest practical approach to perfect combustion, is the only alternative remaining in an effort to keep the O_2, CO and H_2 in the flue gas to a minimum. This is the midway point between a reducing flame and an oxidizing flame.

Curve "B" (Fig. 2) illustrates the ratio of the constituents in the flue gas of an oil-fired crucible furnace and is based on a No. 2 furnace oil with a C to H_2 ratio of 85 to 15. This follows exactly the same pattern as the manufactured gas curve but, inasmuch as two widely different fuel analyses are involved, the relationship between air supplied for combustion and flue products will be different in each case. What is common to both, however, is that a deficiency of air for combustion will result in incomplete combustion, producing a reducing atmosphere, whereas an excess of air for combustion will result in complete combustion, or an oxidizing atmosphere.

Maintaining a correct fuel to air ratio in the combustion of oil or gas is important, but this should be coupled with a proper combustion rate. Rate of combustion is the amount of fuel burned in a given period of time. In starting a cold crucible melting furnace, the rate of combustion may be the same from the time the furnace is started until the crucible is pulled, although the efficiency of combustion may vary somewhat until the furnace reaches temperature.

Curves "A" and "B" (Figs. 1 and 2) are based on a furnace operation at near maximum temperature. Combustion rate is largely a matter of furnace size and design commensurate with what is established as a normal melting time for the size of crucible the furnace has been designed to accommodate. Burner size and capacity are also important factors in determining the combustion rate of a furnace of any particular size. These factors usually are determined and established by the furnace manufacturer.

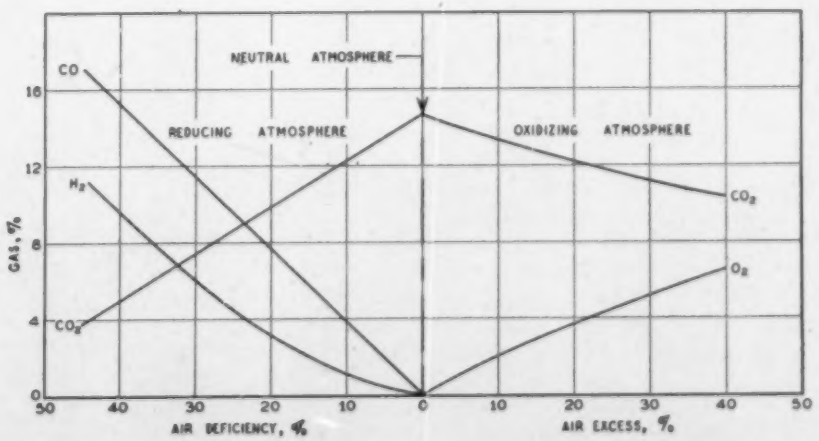
At the present time combustion control normally is left in the hands of the crucible furnace melter, who attempts to judge the proper fuel and air setting by visual analysis of the color of the flame exhausting through the cover hole of the furnace. At best this is a haphazard method, although some melters, with sufficient experience, become fairly proficient in their interpretation of flame color in terms of combustion efficiency. The author, having chemically analyzed the products of combustion from many crucible furnaces in order to check combustion, has found more often than not that flame color yields an erroneous answer.

The oxidation, or non-oxidation, of a piece of bright copper or zinc inserted through the cover hole of the crucible furnace may give a rough idea as to whether the furnace atmosphere is reducing or oxidizing, but it is doubtful if this method can be used to set the fuel and air ratio so as to obtain a critically neutral atmosphere.

Gas Analyzers

For the analysis of vent gases to determine the efficiency of combustion various types of gas analyzers are available, from the relatively simple type of portable instrument for obtaining periodic readings, to the permanently installed continuous automatic indicating and recording type of analyzer. The principle of operation may be electrical, mechanical, or chemical. For general foundry checking, probably the comparatively inexpensive, easy to operate, portable type of analyzer is most practical. A fast operating portable electrical analyzer has been devel-

Fig. 2—(Curve B) Ratio of constituents in flue gas of oil-fired crucible furnace.



oped which will quickly indicate whether the furnace atmosphere is reducing or oxidizing, but the readings from this type of instrument must be properly interpreted. It is questionable whether the continuous automatic recorder or indicator is a practical investment for the average non-ferrous foundry.

Exhaust Gas Analysis

High initial cost and difficulty of maintenance, particularly the gas sampling tube in the furnace, make the use of this equipment impractical. However, for laboratory and development work to determine combustion characteristics of a furnace this continuous automatic analysis equipment undoubtedly is of value.

Probably the most accurate method of analysis of the exhaust gases, and the most practical, is the determination of CO_2, CO, and O_2, by volume, by the chemical Orsat method. This involves the chemical absorption of each gas, samples being obtained through the furnace cover hole with a quartz or silica sampling tube. When making an analysis, it is important that an O_2 reading supplement the CO_2 reading, as a CO_2 reading alone may be misleading.

For example, referring to curve "B" (Fig. 2), a CO_2 reading of 12 per cent could be interpreted as 20 per cent excess air (oxidizing) or 10 per cent air deficiency (reducing), whereas if this is followed by an O_2 determination and the O_2 reading is 4 per cent, it definitely proves that the air is in excess. This indicates that either more fuel, or less air, is required in order to approach a neutral atmosphere, or a state of perfect combustion.

Burner Adjustment

While it is a simple matter to make a flue gas analysis and a burner adjustment, the stability of this adjustment is short-lived in the average foundry, particularly where both air and fuel are variable. A correct setting can be made during any particular melt, but when the metal is melted and ready for pouring the air and fuel usually are turned off.

When they are turned on again the new setting may be incorrect. Checking the furnace for correct combustion during each melt, or after each disturbance of the fuel and air valves, is impractical. A practical solution would be the development of a burner, by the burner or furnace manufacturer, simple and foolproof in design, that would enable the melter to turn off both fuel and air without disturbing the ratio setting.

In addition, it should maintain the fuel and air ratio through the entire combustion range, from low flame to high flame. In fact, considerable work has already been done along these lines and ratio controlled burners, for crucible melting furnaces, are beginning to appear on the market. This undoubtedly will help in maintaining uniform combustion, and aid in the elimination of gas absorption during the melting cycle.

DISCUSSION

Chairman: D. FRANK O'CONNOR, American Saw Mill Machine Co., Hackettstown, N. J.

Co-Chairman: HERMAN SMITH, Federated Metals Div., American Smelting & Refining Co., Pittsburgh, Pa.

GEORGE DALBEY[1]: I would like to confirm Mr. Schmid's statements about the difficulty of having the furnace man judge the composition of his furnace gas by visual means. I had a similar experience with the furnace man. He described the flame issuing from the furnace as a blue-green flame and decided that it was an oxidizing atmosphere. However, when we analyzed the gas we found considerable CO.

[1]U. S. Navy Yard, Mare Island, Calif.

TECHNOLOGICAL PRINCIPLES of CASTING DESIGN

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CASTING is one of the most important processes in the manufacture of all machines. By weight castings form 85 to 90 per cent of machine tools, 50 to 55 per cent of tractors, 70 per cent of escalators, 25 per cent of crane bodies, trucks, etc.

More than 20 to 25 per cent of the total cost price of a machine goes in the manufacture of castings.

The development of more durable types of cast iron has led to the more extensive use of iron castings. In a number of new designs cast iron has been successfully used in place of steel for the manufacture of crankshafts, spindles, gear-wheels, etc.

With the more extensive application of casting, greater importance must be attached to the design of the components from the viewpoint of the technological processes involved in their fabrication. Better designs make it easier for a factory to master the production of a new machine, reduce cost prices and make it possible to run the foundry shop with workers less highly skilled.

The technological principles of casting design make themselves most strongly felt in plants engaged in repetitive and mass production of machines, where molding is done mechanically and expensive pattern equipment is employed.

The design of the whole casting as well as its elements and member junctions must take into consideration the phenomena which occur during the solidification and cooling of the metal.

Unless this principle is followed it is impossible to obtain castings that are free from cracks, flaws and internal stresses without employing special methods, complicating the whole technological process.

The size and intricacy of the castings should not only be in accordance with the general technological principles of casting processes but also in accordance with the specific conditions obtaining in the particular foundry where they are cast.

General Design of Castings. Simplicity in production demands simplicity in form. Every casting should be based on the simplest possible geometrical figure. The whole design should be based on the idea of obtaining this figure in the casting and all the appendages such as ribs, flanges, bosses, etc., should be subordinate to it.

In designing castings every effort must be made to obtain the simplest possible patterns and core boxes. The shape of the various surfaces of the casting should be made as close as possible to plane or cylindrical surfaces in order to simplify the making of the pattern. (Fig. 1).

The design of the casting should provide for the easy withdrawal of the pattern from the mold without the use of loose pieces and irregular parting line, etc. The more complicated the design of the casting, the greater the number of cores, bosses, etc., the more difficult it is to achieve accuracy of dimensions.

It is particularly desirable that there should be no loose pieces on jobs molded mechanically and intended for machining in fixtures.

To make it easier to remove the pattern from the mold it is desirable that the casting should be designed with the necessary draft. If this is

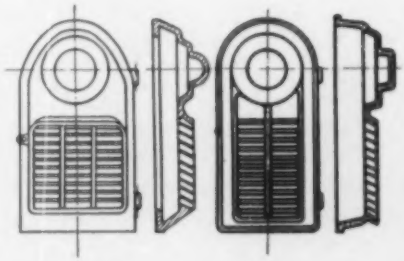


Fig. 1—Drawing of door casting showing poor design (left) and improved design (right).

Greater importance must be focused on design of castings from the viewpoint of technological processes involved in their fabrication. Technological principles of casting design makes themselves more strongly felt in mass production foundries, where molding is done mechanically and expensive pattern equipment is employed. The writer presents herein fundamental principles which should interest the casting designer.

This paper was scheduled for presentation at a Steel Session at the Fiftieth Anniversary Meeting of the American Foundrymen's Association, May 10, 1946, at Cleveland.

Table 1

PATTERN DRAFT ANGLES

$\frac{a^*}{h}$	Value	β	Remarks
$\frac{1}{5}$		11°30′	where h is less than 1 in.
$\frac{1}{10}$		5°30′	intermediate h
$\frac{1}{20}$		3°	intermediate h
$\frac{1}{50}$		1°	where h is more than 15 in.

*See Fig. 2.

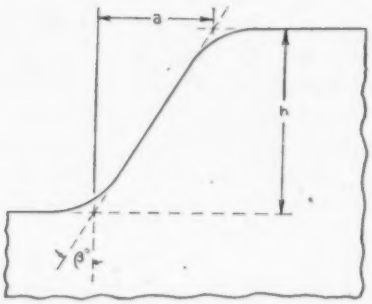


Fig. 2—Sketch showing recommended tapering practice (Table 1).

not done the pattern-maker will have to taper the walls of his pattern thus deviating from dimensions of the job.

U. S. S. R. Government Standard No. 2670-44 planned by the Stalin Machine-Tool Institute recommends the draft a/h=1/5=1/20=1/50 (Fig. 2) which corresponds to the angles β. See Table 1.*

If the slope is obtained from the cast dimension a the quantity $\frac{a}{h}$ may deviate from the figures given in Table 1 in order to avoid fractions. The tapered surfaces of the main part of the casting or its separate components (bosses, ribs, brackets, bores, etc.) must coincide with the parting line of the mold.

Intricate and cumbersome castings are best designed in parts and after casting may be fastened together by welding, bolting or some other method of jointing. This applies in particular to parts of a casting which form large projections and give rise to internal stresses later on. The cost of making molds increases very rapidly and not in proportion with their dimensions.

In addition to this, the division of the casting into a number of parts makes the molds simpler to build up, obviates the use of heavy cranes, and permits the employment of less skilled workers.

The design of the casting must be such as will allow the liquid metal to flow freely into the mold.

Large horizontal surfaces parallel to the top of the mold should be avoided on account of the fact that the gases which form in the mold and are given off from the metal may be retained on these surfaces. It is desirable to design that element of the casting which is uppermost in the mold in such a way that it will assist free flow of the gas and produce a casting without gas cavities.

Solidification of Casting

The design of castings should foresee a regular progression in the process of solidification. This means that the solidification should begin at a definite, known point in the casting, usually the point farthest from the gate where the liquid metal is poured and should spread from this point in direct sequence ending with the head or riser.

In the majority of cases the solidification process begins in that part of the casting which is at the bottom of the mold and ends with the part nearest the top, the whole process progressing gradually. This rule must always be borne in mind when handling those metals whose volume decreases considerably as they cool, such as steel, white iron, etc.

Every cross-section of the casting, especially at points where there are masses of metal, must be ensured a regular supply of liquid metal while the solidifying process is under way if cavities and porosity are to be avoided. The isolation of heavy cross-sections or their disposition in parts of the mold where they are not constantly supplied with liquid metal is particularly dangerous to the quality of the casting.

The design of the casting must allow for shrinkage and the checking of shrinkage caused by mechanical and thermal factors. In order to prevent the so-called "mechanical" stresses castings should allow for free contraction.

Thermal Stresses

To prevent "thermal" stresses the design should provide for all cross-sections to cool as evenly as possible. The approximate criterion of the rate of cooling of a cast member is the relative thickness of the cross-section, that is, the area of the cross-section divided by its perimeter.

The designer must remember that internal stresses which arise are directly proportional to the elasticity modulus of the metal.

If the elasticity modulus of steel averages 31,460,000 psi., white iron 24,310,000 psi., and gray iron 14,300,000 psi., then, all other factors being equal, the ratio of stresses in steel, white and gray iron will be: 2.2: 1.7: 1.0.

In designing the beds of lathes and similar machines it must be remembered that the great difference in relative thickness between the way and the walls of the bed will result in internal thermal stresses which may lead to warpage of the casting.

The following ratios must be maintained for cast iron fly-wheels

Table 2

PERMISSIBLE DEVIATION IN DIMENSIONS OF GRAY IRON CASTING

Dimensions of Castings, in.	Degree of Accuracy I (+)	I (−)	II (+)	II (−)	III (+)	III (−)
	Permissable Deviation in in.					
Up to 3.94	0.04	0.04	0.059	0.04	0.079	0.04
3.94— 7.88	0.04	0.04	0.079	0.04	0.079	0.079
7.88— 11.8	0.04	0.04	0.079	0.059	0.12	0.079
11.8 — 19.7	0.059	0.04	0.12	0.079	0.16	0.12
19.7 — 35.5	0.079	0.04	0.16	0.079	0.2	0.12
35.5 — 47.3	0.12	0.059	0.2	0.12	0.24	0.16
47.3 — 71.0	0.16	0.079	0.24	0.16	0.32	0.2
71.0 —102.4	0.2	0.12	0.32	0.2	0.4	0.24
102.4 —149.7	0.24	0.16	0.4	0.24	0.47	0.32
149.7 —212.5	0.32	0.24	0.47	0.32	0.59	0.4
Over 212.5	0.4	0.32	0.59	0.4	0.79	0.59

*All dimensional and tensile values in this paper were originally in the metric system, and were converted to the decimal system.

and pulleys:

$$\text{Flywheels: } 0.5D = 21 \times 0.5\frac{K_s}{K_r} + d \times 0.5\frac{K_h}{K_r}$$

$$\text{Pulleys: } 0.5D \times \frac{K_r}{K_h} = 21 \times 0.5\frac{K_s}{K_h} + d \times 0.5$$

Where D is the diameter of the wheel
l = length of the spokes
d = diameter of the hub
K_r, K_s, K_h = coefficients equalling inverse ratio of cross-section to the perimeter of rim, spokes and hub.

The following empirically obtained ratios between the structural elements of pulley wheels are also recommended:

$$\frac{Fs}{Ps} : \frac{Fr}{Pr} = 0.9; \frac{Fh}{Ph} = 04; \frac{Fr}{Pr} = \frac{1}{1 - \frac{100\,A}{d}}$$

where Fs, Fr, Fh = the area of the cross-section in mm² and Ps, Pr, Ph the perimeter of the cross-section of the spokes, rim and hub,

d—diameter of the hub

A—a coefficient depending on the diameter of the pulley D mm:

*D, mm.**	*A*
300	0.39
400	0.52
500	0.55
600	0.73
700	0.51
800	1.0
900	1.18
1000	1.3

The inevitable deformation of wooden patterns, the imperfect nature of sand molds[1] and variations in the shrinkage of the metal may cause certain deviations from the required dimensions and weight of the casting. The extent of permissible deviations for cast iron and cast steel are laid down in U. S. S. R. Government Standards 1855-42 and 2009-43 and are given in Table 2-5.

These figures are the highest permissable and do not prevent severer standards from being set, often for whole branches of industry where the foundries have achieved a higher all-round standard.

Up to date there are no U. S. S. R. Government Standards of permissible deviation in dimensions and weight for other metals.

Tables 6-9 show the recommended permissible deviations in dimensions and weight for castings for malleable iron and copper alloys worked out by the Central Research Institute of the Heavy Machine Building Industry.

When there are rough castings in a machine moving one against another the minimum clearance recommended is given in the following Table 10[1].

The drawings should show the locating surfaces to be used in machining so that they may serve as a starting point in making the pattern and in checking the patterns and the castings.

In designing the casting, care must be taken to ensure the possibility of removing heads and risers, knocking out the cores and core irons.

The design of the casting should also take into consideration the extent to which later use of the part will depend on the method of casting. If, for example, the casting is intended to work under pressure its design should be such that the cores can be held in the mold by core prints without any recourse to chaplets.

By distribution of loads in the construction, the design should consider the properties of the metals used. For instance, a gray iron works better under compression than at a tension (Fig. 3 and 4) while steel, on the contrary, works equally well under both stresses.

[1]This paper does not deal with specific conditions of die casting, centrifugal casting, etc.

*1 in. = 25.4 mm.

[1]E. Geiger, "Designing for Foundry Production," *Transactions*, American Foundrymen's Association, vol. 50, p. 149 (1942).

Table 3

Permissible Deviation in Dimensions of Cast Steel Castings

Dimensions of Castings, in.	*Degree of Accuracy* — *I* Permissable Deviation in in. (+)	(−)	*II* (+)	(−)	*III* (+)	(−)
Up to 3.94	0.04	0.079	0.079	0.04	0.12	0.079
3.94— 7.88	0.04	0.04	0.079	0.04	0.16	0.079
7.88— 11.8	0.04	0.04	0.12	0.079	0.16	0.12
11.8 — 19.7	0.059	0.04	0.12	0.079	0.2	0.12
19.7 — 31.5	0.079	0.04	0.16	0.079	0.24	0.16
31.5 — 47.3	0.12	0.079	0.2	0.12	0.28	0.2
47.3 — 71.0	0.16	0.079	0.24	0.16	0.35	0.24
71.0 —102.4	0.2	0.12	0.32	0.2	0.47	0.32
102.4 —149.7	0.24	0.16	0.4	0.24	0.55	0.4
149.7 —212.5	0.32	0.24	0.47	0.32	0.63	0.47
Over 212.5	0.4	0.32	0.59	0.4	0.79	0.63

Table 4

Permissible Deviation in Weight in Gray Iron Castings

Weight of Casting in lb.	Permissible Deviation in Weight, Per Cent — Degree of Accuracy — I	II	III
220	4.5	6.0	10.0
220-2200	5.0	8.0	10.0
2200	3.0	5.0	8.0

Table 5

Permissible Deviation in Weight in Steel Castings

Weight of Casting in lb.	Permissible Deviation in Weight, Per Cent — Degree of Accuracy — I	II	III
Up to 220	6.0	8.0	10.0
220-2200	5.0	7.0	10.0
Over 2200	4.0	6.0	8.0

Thickness of Casting Walls

Selection of minimum permissible thickness of casting walls is based on the dimensions, weight, molding conditions, the use of the walls in the machine, etc.

Sections thinner than the following are not recommended:

for cast iron.....0.12—0.16 in.
for steel.....0.20—0.24 in.
for bronze.....0.12—0.16 in.
for aluminum.... 0.16—0.20 in.

Efforts to reduce the weight of a casting should not lead to excessively thin sections which may cause scrap, require great fluidity to cast and consequently excessively high temperatures.

The thickness and cross-section of walls should concur with the general laws for obtaining a good casting: harden progressively and cool regularly.

In tall castings the thickness of the walls should as a rule increase from top to bottom as they stand in the mold.

The difference in the relative thickness of the walls of castings, especially those to be joined should be a minimum and should not exceed the ratio 1:2.0.

In cases where a great difference

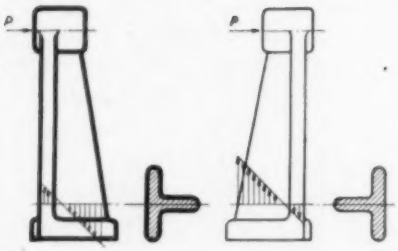
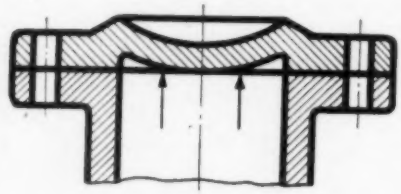


Fig. 3 (above)—Sketch of bracket showing poor design (right) and improved design (left).



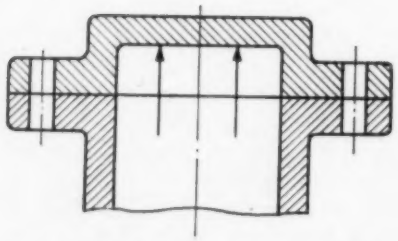


Fig. 4—Sketch of piston showing poor design (below) and improved design (above). Arrows indicate stresses.



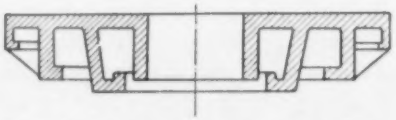


Fig. 5—Sketch of a plate showing poor design (below) and improved design (above). Inside walls should be made thinner than the outside walls.

in cross-section is unavoidable it is recommended that the heavy section be designed with detachable parts, for instance, ways of machine tool beds can be bolted.

The thickness of the inner walls, in view of worse cooling conditions, should be approximately 0.8 of the outer walls (Fig. 5).

The permissible deviation in thickness of casting walls is the following:

Condition	Per Cent +	Per Cent −
For jobs with a size up to 1½ ft	6	2
For jobs with a size up to 6 ft	8	3
For jobs with a size over 6 ft	10	3

In places where chaplets are inserted thin walls should have local thickening so that the mass of the metal is sufficient to melt the chaplet down. If this is not done there is a danger of local defects being caused.

In designing it must be remembered that one and the same cast iron melted at the same temperature acquires different structures when cooled at different rates so that different parts of same casting may have different mechanical properties.

Test bars cut from different parts of a cast gas cylinder showed the following differences in strength:

Sections in in.	Tensile Strength in psi.
0.31	52,500
0.47	48,250
0.71	45,400
0.87	41,100

The working drawings of the casting should show the minimum permissible strength at the most important sections so that they are known to the foundrymen.

Angles, Tapering Sections, Junctions. The chief condition for obtaining castings without defects or stresses is the gradual tapering of sections and the avoidance of sharp angles. Sharp changes in the sections of walls from one thickness to another are not permissible.

The correctness of the proportions of adjacent cross-sections may be checked by the method of inscribed circles.

When the change in thickness is less than 1:2 it may take the form of a fillet; where the difference is greater the form used is that of a wedge.

Wedge-shaped changes in wall thickness are to be designed with a tapering not exceeding 1:4 (Fig 6).

Recommended tapering is given in Table 11.

The radii of fillets should be from 1.6 to 1.3 of the cross-sections which it connects. Larger radii in the fillet (equal to the thickness of the walls or greater) are not to be recommended as this leads to local thickening.

U. S. S. R. Government Standard No. 2716-44 recommends the following series of raddi for cast fillets, depending on general dimensions. This standard was produced by the

Table 6

PERMISSIBLE DEVIATION IN DIMENSIONS OF MALLEABLE IRON CASTINGS

Dimensions of Castings in in.	Degree of Accuracy I Permissible Deviation ± in.	Degree of Accuracy II Permissible Deviation ± in.
Up to 3.94	.02	.059
3.94- 9.85	.04	.079
9.85-15.75	.04	.098
15.75-25.6	.059	0.12
25.6 -39.4	.059	0.14
39.4 -63.0	.059	0.16

Table 7

PERMISSIBLE DEVIATION IN DIMENSIONS OF COPPER ALLOY CASTINGS

Dimensions of Castings in in.	Degree of Accuracy I Permissible Deviation, in in. ±	Degree of Accuracy II Permissible Deviation, in in. ±
Up to 5.9	0.02	0.04
5.9 - 9.85	0.02	0.059
9.85-23.6	0.04	0.059

Table 8

PERMISSIBLE DEVIATION IN WEIGHT PER CENT OF MALLEABLE IRON CASTINGS

Weight of Casting in lb.	Degree of Accuracy I Permissible Deviation in in. (+)	(−)	Degree of Accuracy II Permissible Deviation in in. (+)	(−)
Up to 1.1	0.24	0.2	0.433	0.39
0.22- 1.1	0.24	0.2	0.35	0.35
1.1 - 5.5	0.2	0.2	0.32	0.32
5.5 - 27.5	0.2	0.16	0.28	0.28
27.5 -115.5	0.16	0.16	0.24	0.24
Over 115.5	0.16	0.12	0.2	0.2

Table 9

PERMISSIBLE DEVIATION IN WEIGHT PER CENT OF COPPER ALLOY CASTINGS

Weight of Casting in lb.	Degree of Accuracy I in in. (+)	(−)	Degree of Accuracy II in in. (+)	(−)
0.22	0.24	0.2	0.43	0.43
0.22-0.44	0.2	0.2	0.39	0.39
0.66-0.88	0.2	0.16	0.35	0.35
0.88-3.96	0.16	0.16	0.32	0.32
1.76-3.52	0.16	0.12	0.28	0.28
3.52-7.05	0.12	0.12	0.24	0.24
Over 7.05	0.12	0.079	0.2	0.2

Table 10

MINIMUM CLEARANCE IN IN.

	Dimensions of Castings Up to 1½ Ft.	Up to 6 Ft.	Over 6 Ft.
One surface as cast, other machined	0.39	0.59	0.985
Both surfaces as cast	0.39	0.79	1.18

Table 11

RECOMMENDED TAPERING OF WALL THICKNESSES IN IN.

a	*b*	*c*	*r*
0.394-0.59	0.59	0.12	0.2
0.59 -0.79	0.79	0.16	0.2
0.79 -0.985	0.985	0.2	0.2
0.985-1.18	1.18	0.24	0.31
1.18 -1.38	1.38	0.28	0.31
1.38 -1.57	1.57	0.31	0.31
1.57 -1.77	1.77	0.35	0.39
1.77 -1.96	1.96	0.39	0.39
1.96 -2.16	2.16	0.43	0.47
2.16 -2.36	2.36	0.47	0.47
2.36 -2.56	2.56	0.51	0.59
2.56 -2.75	2.75	0.55	0.59
2.75 -2.95	2.95	0.59	0.59

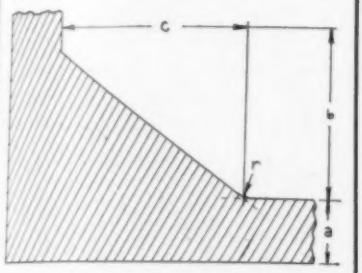


Table 12

ALLOWANCES FOR MACHINING IN CASE OF SINGLE BORE

Greatest Diameter in in.	*Maximum Allowances in in. by Degrees of Accuracy* I	II	III
3.94	.....0.12	0.16	0.2
3.94- 7.88	.....0.12	0.2	0.24
7.88-11.8	.....0.2	0.24	0.32
11.8 -19.7	.....0.24	0.32	0.39

Table 13

ALLOWANCES FOR MACHINING IN CASE OF SEVERAL BORES

Greatest Dimension of Casting in in.	*Maximum Allowances for Machining in in. by Degrees of Accuracy* I	II	III
Up to 7.9	.....0.16	0.2	0.24
7.9- 11.8	.....0.2	0.24	0.28
11.8- 19.7	.....0.24	0.28	0.32
19.7- 31.5	.....0.24	0.32	0.39
31.5- 47.2	.....0.32	0.39	0.47
47.2- 71.	.....0.39	0.47	0.55
71. -102.3	.....0.39	0.55	0.63
102.3-149.5	..... —	0.55	0.63
149.5-213.	..... —	0.71	0.79
Over 213	..... —	0.79	0.87

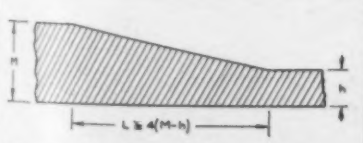


Fig. 6—Wedge-shaped changes in section as shown above in wall thickness should be designed with a taper not exceeding ratio 1:4.

Moscow Machine Tool Institute.[1]

Radius (in.) 0.039; 0.079; 0.118; 0.197; 0.315; 0.394; 0.473; 0.59; 0.79; 0.98; 1.18; 1.57.

The number of radii in fillets used in one pattern should be the minimum possible and where possible should not exceed one.

In order to avoid local accumulation of metals and secure even surfaces at rounded angle junctions between two different sections, the external radius of the junction should equal the thicker member, and the inner should follow the U. S. S. R. Government Standard No. 2716-44.

Cross coupled walls (ribs) should preferably be designed as double T-forms (Fig. 7).

Sharp inner angles should be avoided as they form a focal point for the accumulation of local stresses. Acute angles are particularly dangerous in designs that may suffer fatigue.

In order to lessen the mass of metal in those angles where a perpendicular strengthening rib meets two walls the design should allow for an opening in the rib which does not weaken the construction but makes for a good casting.

In designing changes in cross-sections the engineer must remember that some of the walls will have a bigger thickness due to machining allowances.

[1]The author, with a view to discovering the diameter of fillets actually used in practical work in various types of machine, measured 150 iron castings in 14 machines: six presses, four machine tools and four tractor engines.

A total of 742 fillets were measured and the following frequency obtained.

Radius of Fillet, in.	*Frequency of Occurrence Per Cent*
0.12	1.7
0.16	0.7
0.2	24.5
0.24	1.5
0.28	0.1
0.32	4.7
0.35	0.1
0.39	48.8
0.47	0.1
0.59	1.9
0.63	0.4
0.79	6.9
0.91	0.4
0.95	0.4
0.99	1.5
1.1	0.3
1.18	1.3
1.38	0.8
1.42	0.1
1.58	0.5
1.97	1.1
2.05	0.5
2.95	0.1

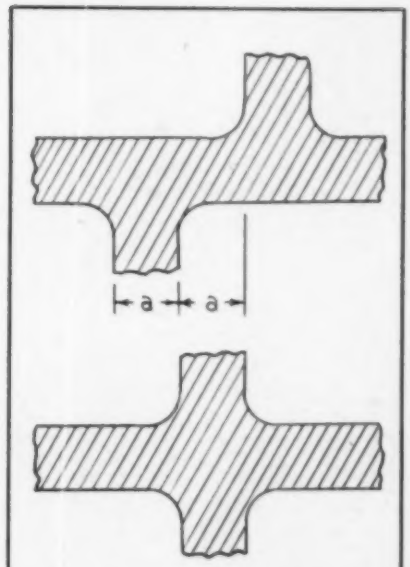
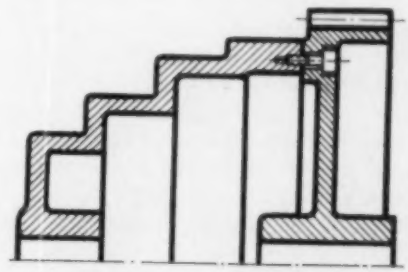


Fig. 7—Sketch of wall junctions showing poor design (below) and improved design (above).



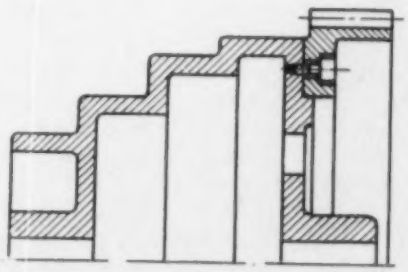


Fig. 8—Sketch of pulley showing poor design (below) and improved design (above). In correct design no cores are required.

Hollow Castings and Bores. In designing castings all shapes which make the use of cores necessary should be avoided to the greatest possible extent as they complicate molding and increase its cost. Wherever possible dry cores should be abandoned in favor of green mold (Fig. 8).

The question of whether or not a bore should be made by casting must

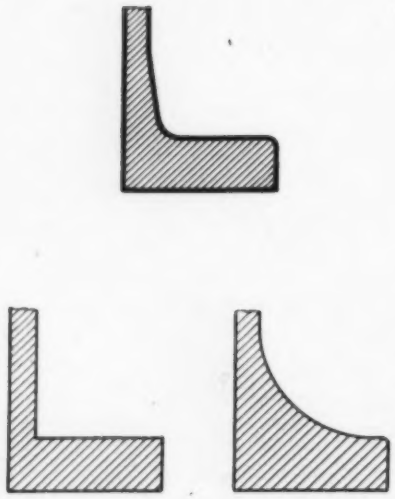


Fig. 9—Sketch of flanges showing poor design (below) and improved design (above).

be decided individually for each case and depends on the nature of the job and technology of fabrication.

Allowances for machining bores, whose size is the factor determining whether the bore shall be made or not, are to be calculated in accordance with existing practices. Tables 12 and 13 show the recommended allowances.

When a large number of cores is required for a casting, such as the cylinders of air-cooled engines, for example, the question of unification must be considered. Design should favor mechanical core molding.

The design of the interior of a hollow casting should be such that the core print and the shape of the core enable it to be accurately fixed in position. The core print should however possible be a continuation of the body of the core. The use of chaplets is extremely undesirable.

When the length of a core is more than double its diameter, the designer should allow for it to be fixed at both ends.

The thickness of the core should be sufficient to allow it to be strengthened by cast or wire core irons; it should not be easily broken during transportation or under the pressure of the liquid metal when the mold is filled.

The design of the casting should allow for the escape of gases from the core preferably from the top of the casting.

It is recommended that walls be strengthened by means of flanges wherever there is an opening in the casting.

The design of the casting should allow for the easy removal of the core and core iron from the casting and for the cleaning of the interior surfaces. If the job is one that requires entirely enclosed internal cores the design must provide openings for the removal of the core, and for their closing after removal.

Ribs and Brackets. As far as possible ribs should be flat in shape. If the ribs are narrow and high stresses may occur at the outer edges which may lead to cracks.

The junction between the ribs and the main casting should prevent any local accumulation of metal.

For the same reason criss-cross ribs should be arranged checkerboard fashion.

When the thickness of flanges differs greatly from that of the body of the casting the change in thickness should be gradual and tapered 1:4.5 (Fig. 9).

Grooves should be made by casting if the unmachined dimensions are not to be less than 0.2 to 0.24 in. depth and 0.79 to 0.98 in. width.

Bosses and Lugs. Boss, lugs, brackets and all other protruding elements of castings should be so designed that they can be removed from the mold together with the main pattern.

To obtain good, strong metal in bosses and lugs their thickness should not differ greatly from that of the main casting.

When there are several lugs and bosses on one surface they should be joined.

Bosses should not be included in the casting design when the surface to support bolts, etc., may be obtained by milling or countersinking.

Lugs should be thick enough to ensure that they can be machined without the main casting being touched.

Minimum recommended heights for bosses on a casting are:

Condition	*In.*
For jobs up to 1½ ft. in length	0.2
For jobs up to 6 ft. in length	0.6
For jobs over 6 ft. in length	1.0

To avoid frequent change of position in planing or milling the various lugs, all those on one side of the casting should, if possible, be in one plane.

The width of lugs, supports, etc., should consider some slight distortion in casting.

Some Examples of the Technological Analysis of Casting Designs. Machine Tool Bed. The base of the machine tool bed casting is an open box with two sleeves which are a continuation of the ways of the bed. In two walls perpendicular to each other there are openings that are to be closed with covers. The dimensions of the bed are 51×30×18 in.

The original thickness of the walls was 0.47 in. In view of the somewhat intricate shape of the bed, however, and especially the change in direction of the members hindering the flow of the liquid metal, when poured and also to ensure even cooling in all cross-sections of the casting, thickness of the outer walls is preferably increased to 0.63 in.

The irregular shape of the walls forms a labyrinth through which the metal must flow. This may cause warp due to the sand being washed away from the sides of the mold. Without changing the outer surfaces of the walls the inner surfaces should be made flat. (Fig. 10).

The tubular section of the members as at first designed did not allow for the fixing of the cores. To fix the cores securely and to make it easier to insert the cores in the mold it would be better to make the tubular members in the shape of an inverted V.

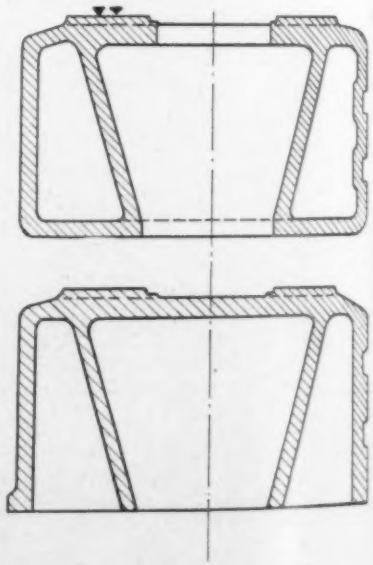


Fig. 10—Sketch of a machine bed showing cross-section of ways. Poor design is shown above while improved design is shown below.

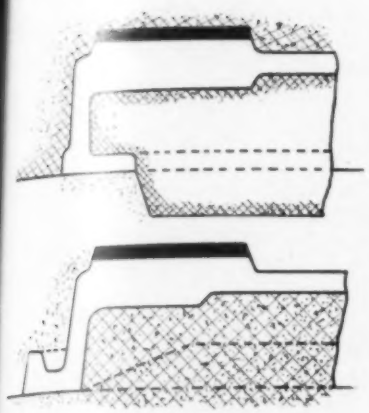


Fig. 11—Sketch of bedplate for motor with external flanges to simplify molding. Poor design is shown above while improved design is shown below.

The design of the lower part of the bed is unnecessarily complicated and can be replaced by ordinary flanges (Fig. 10) thus avoiding the unnecessary use of cores and giving stability to the mold.

Engine Base

The use of internal flanges greatly complicates molding. A number of cores must be used which, although they are of simple shape, are big, heavy and involve great expenditure of labor. In consequence of this, core boxes and drying plates are necessary. Changes in the design of the flange (Fig. 11) enable the foundryman to use green mold.

The strengthening ribs should be made somewhat lower and at the same time thicker to allow for the even cooling of the casting (Fig. 12). Reducing the number of ribs, decreasing their height and increasing their thickness simplifies molding, especially if the whole job is given sufficient taper.

The lugs for bolting the bed to the foundation are weakly joined and may be broken off. By using an outside flange half the length of the lugs merges into the bed which considerably strengthens them.

Spindle Housing

Basic casting of a spindle housing is a rectangular box with an inside arch, dimensions—4 × 36 × 36 in. The casting is open at one side.

The original design required a wall thickness of 0.8 in. but on account of a number of lugs, ribs, etc., it was found that the walls were not strong enough and the box cracked on account of the mechanical hindering of shrinkage; the thickness of the walls was therefore increased to 1.0 in. At the same time an arch and a strengthening rib were introduced into the design which made it possible to cast the job without cracks.

The edges of the internal surfaces should be straight and strengthened by a flange. The core and consequently the pattern and the core box, then form one straight line. This simplifies the cores, makes it easier to manufacture and dry them, and, what is more important, ensures greater accuracy in assembling the mold and checking the dimensions. The producing of the pattern and the core boxes is made simpler and cheaper.

The height of the box should be increased by 0.8 to 1.5 in. in order to strengthen the flange.

The brackets of the box have a width and thickness which are not proportional to the thickness of the box walls. The necessary allowances for machining increase the mass of metal still further giving a ratio of thickness 2:1 in.

The supporting arch with the opening by which the housing is fastened to the bed is subject to big internal stresses due to the difference in the mass of metal in the parts of the casting adjoining it. The cores necessary to make this part are too complicated.

The supports in the bottom of the box have the following two technological defects.

1. The mass of metal is disproportionately great as compared with the thickness of the bottom.

2. The strengthening rib connecting the supports with the vertical walls of box hinders free shrinkage.

It is more rational to design a system of supports joined only with the bottom of the box, with a corresponding simplification of the whole design.

The original shape given to the strengthening ribs does not justify their existence and they can be

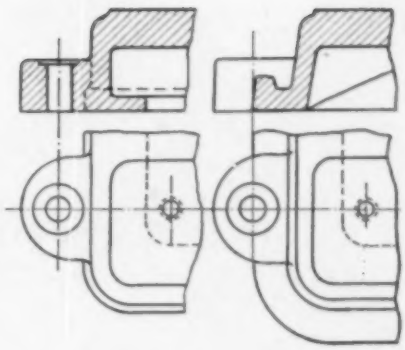


Fig. 12—Sketch of bedplate for motor. External flanges make a stronger joint with the foundation. Poor design is shown at left while improved design is shown at right.

avoided altogether by the change of design previously described.

The beading on cored holes are of excessively massive design especially when we take into consideration the additional thickness that must be given to allow for machining.

The element carrying electrical instruments should be designed separate in the form of a box or a bracket to be bolted on to the walls of the main construction.

The inclusion in one casting of elements of primary importance that are subjected to a heavy mechanical load and those of secondary importance only serves to complicate the design and is incorrect in principle.

The design of this casting is already complicated and requires 20 cores.

The chief economic effect in redesigning castings from the standpoint of the processes involved in the pouring, solidifying and hardening of the metal will be shown by a reduction of scrap.

At the same time simplifying the processes means a reduction in the amount of labor required to make the pattern equipment and molding, also cleaning the castings.

For example, the previously mentioned changes in the design of spindle housing reduces the time required to make patterns and core boxes from 462.2 hr. to 416 hr., that is by 10 per cent.

The time required to make the cores is reduced by 35 min., the assembling of them by 7 per cent and the actual molding by 6 per cent.

MAINTENANCE OF FOUNDRY DUST CONTROL EQUIPMENT

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Peoria, Illinois

NO EQUIPMENT is better than the care it receives. The writer says "care" of equipment because maintenance of equipment as commonly understood is too frequently undertaken only when the equipment has failed. Routine preventive maintenance or "care" is necessary if undesirable interruptions of service are to be avoided.

Routine lubrication, cleaning, replacement of worn parts, and general visual checks of all dust control equipment will forestall the major part of equipment breakdowns during production operations. Preventive maintenance pays by eliminating a large part of the losses incurred when workers and equipment cannot be utilized while essential equipment is being repaired.

Exhaust Hood and Duct Maintenance. Foundry dust control starts with the dust hood and ducts. Once a hood and duct system has been set up properly, every effort should be made to keep it in first class condition. Hood inlet adjustments must be maintained, and dirt accumulations must be removed regularly to provide proper ventilation volume.

Figure 1 shows a typical dirt accumulation in a hood entrance. Dirt accumulations should also be removed to promote the good housekeeping essential to the modern foundry.

Not only do dirt accumulations in hood entrances and ducts cut down the volume of ventilation available, but they frequently constitute a dangerous fire hazard especially where core oil fumes condense and soak into the dust accumulations. Where such fire hazards exist, steam jets should be installed in the duct system ahead of the fan for use in smothering a fire.

Opposed steam jets are necessary to prevent the jet acting as a duct inspirator as well as to fill all sections of duct with steam as quickly as possible. The stop button for the fan should be installed by the steam jet valve. The steam jets should be turned on for a few seconds at regular intervals to insure their proper functioning in emergencies.

Liquids which condense in ducts should be continuously drained out into containers which should be emptied frequently. Figure 2 shows a duct drain and collecting container for core oil residues and water from a mold cooling tunnel exhaust duct. Figure 3 shows the fan section of a similar duct which is badly in need of cleaning. Also the duct entrance door needs a drip deflector on its bottom inside edge.

Routine preventive maintenance or care of foundry dust control equipment is necessary if undesirable interruptions of service are to be avoided. The writer goes into some detail discussing maintenance of fans, cloth type dust collectors, rotary centrifugal collectors, wet collectors, and air filter type collectors. Proper dust control equipment maintenance consists of good routine care and inspection plus replacement of worn parts before equipment failure occurs.

Frequently dust control hoods are severely abused and rendered useless especially at shakeouts. Hood face parts must be rugged. Where experience shows that hoods will be abused, construct them of heavy materials.

When long horizontal ducts are used to convey dirty air to a dust collector, they should be made in flanged sections bolted together to permit turning the duct to a new position several times in the life of the duct. Otherwise, the abrasive wear on the bottom of the duct will cause premature duct failure. Bolted flanged elbows in abrasive dust systems are useful for expediting quick replacement of worn out elbows.

All hoods on grinders should be kept clean and in first-class repair for safety and dust control.

Fan Maintenance. Since the maintenance necessary on a fan is very frequently determined by its design, the proper fan for the dust and fumes to be handled must be chosen with care. For exhaust systems on mold pouring and cooling stations, core ovens, or any system handling condensible fumes, radial blade centrifugal fans will give the most satisfactory results since they are relatively self cleaning by centrifugal force. Figure 4 shows an example of a straight radial blade centrifugal fan impellor in contrast to the multiblade type of fan impellor shown in Fig. 5.

If a multiblade type of fan gives

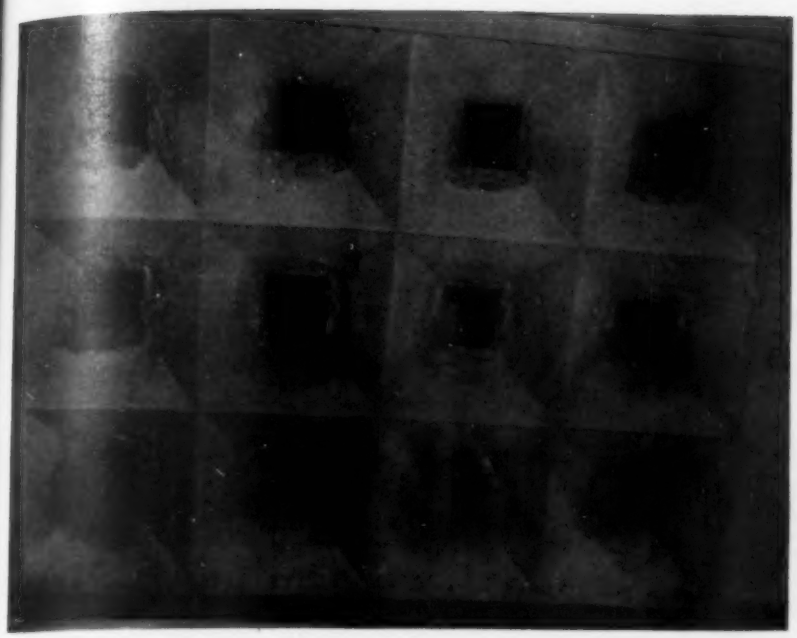


Fig. 1—Typical dirt accumulation at exhaust hood inlets above mold pouring station.

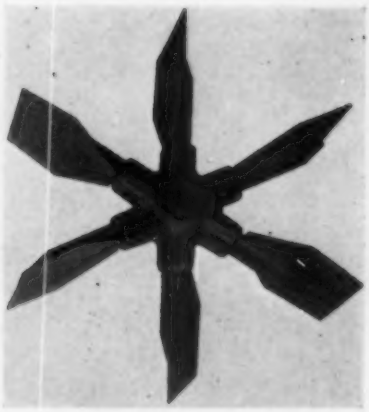


Fig. 4—Straight radial blade fan impellor.

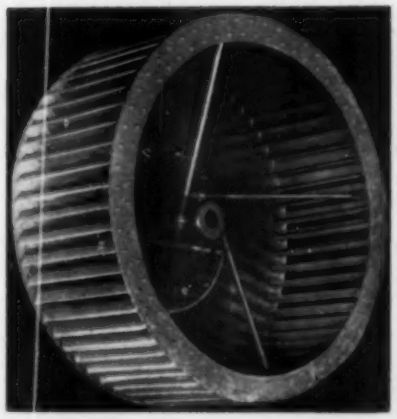


Fig. 5—Multiblade type fan impellor.

Fig. 2—Duct drain and collecting container for duct handling condensible fumes.



Fig. 3—Duct and fan handling condensible fumes from a mold pouring station showing necessity for periodic cleaning.



Fig. 7—Repaired impellor for centrifugal fan.



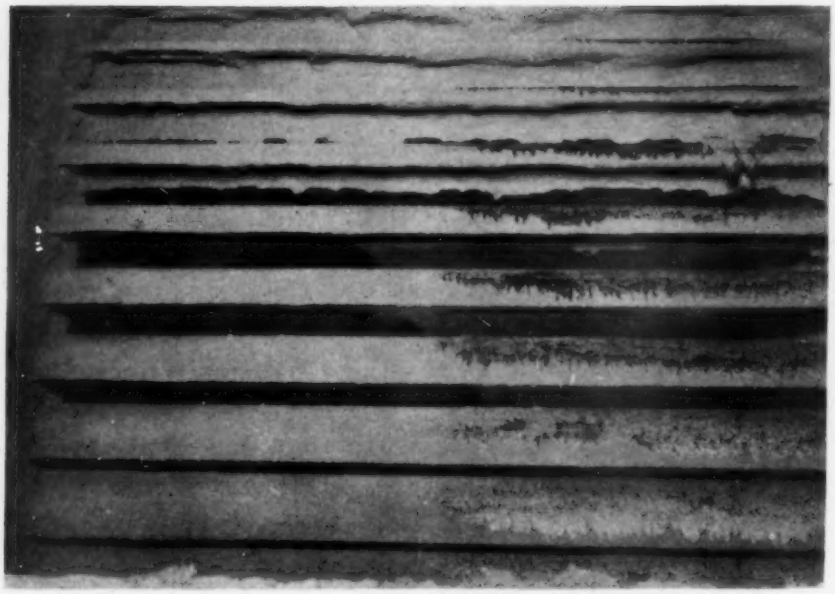


Fig. 6—Multiblade type fan impellor on mold break apart zone exhaust hood showing necessity for regular cleaning.



Fig. 9—Fan drive with one belt missing.

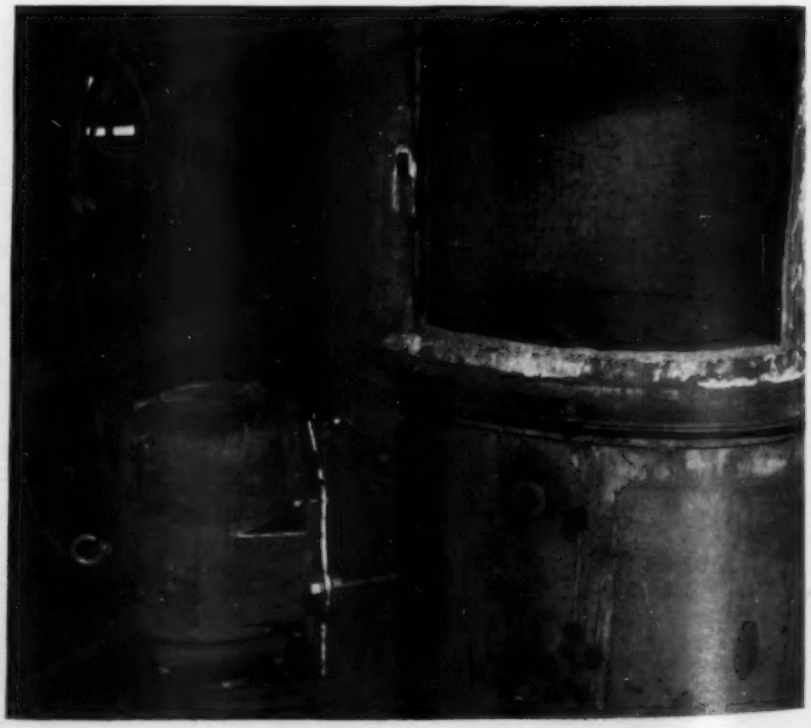
continual trouble from dirt build up and consequent vibration, it should be replaced with a fan of the radial blade type. Figure 6 shows a multiblade type fan impellor on a large mold break apart exhaust system where regular cleaning of the fan impellor is necessary. Cleaning and checking of such fans can be done more easily if a section of the fan scroll sheet is made removable. Suitable sealing strips must be provided on edges of the removable section.

When vibration is detected in a centrifugal fan, it should be checked for worn or missing impellor blades and for material accumulations on the impellor. Material accumulations should be removed and necessary repairs made. Figure 7 shows a repaired fan impellor. Rebuilt fan impellors should be carefully balanced. Vibration build up in a centrifugal fan can be easily detected when the bearings are given their routine lubrication. Centrifugal fans should not be operated at high speeds unless they are given very good care. Vibration from dirty impellors running at high speed can cause serious damage.

On dust and fume control systems where the dust is not collected, propellor type fans are frequently used. When they are used to handle condensible fumes, they should be of the narrow streamline blade type to cut down the blade area which can collect dirt. Even so, they should be cleaned regularly. Convenient access doors should be provided in the ducts (Fig. 8) to permit easy cleaning and changing of fan blades.

The belts on external drive duct fans must be kept tight, and the bearings should be kept well lubricated especially in ducts from mold pouring exhaust hoods which may carry hot gases. Figure 8 also shows an easily adjusted belt tightener for duct fan drives. The drives for centrifugal fans which use multiple V-belts are often abused. Frequently the individual loose or broken V-belt is not replaced until the overload on the remaining belts has damaged them too. See Fig. 9.

Fig. 8—Propellor type fan with ample duct entrance door and convenient belt tightening arrangement.



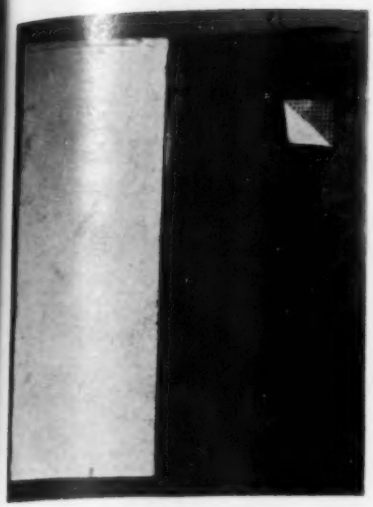


Fig. 11—Dust collector bags showing new bag and well worn bag with dirt accumulation.

Where abrasive dusts may be encountered, the propellor blades must be checked and changed periodically to insure efficient fan performance. Figure 10 shows typical worn propellor blades which need changing in order to insure efficient ventilation. Note the maintenance man's quick method of obtaining an adequate door for fan blade replacement.

Cloth Type Collector Maintenance. Several types of dust collectors are used in the foundry, each having separate maintenance problems. Where there are no condensible fumes, high temperatures, or gummy materials present, the cloth type collector is commonly used.

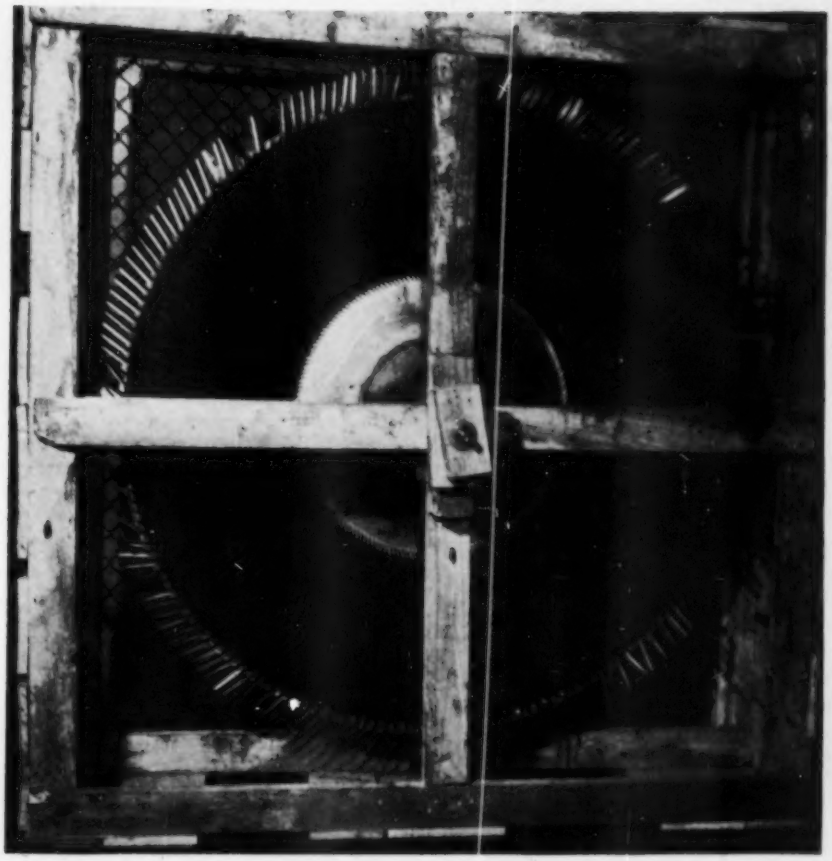
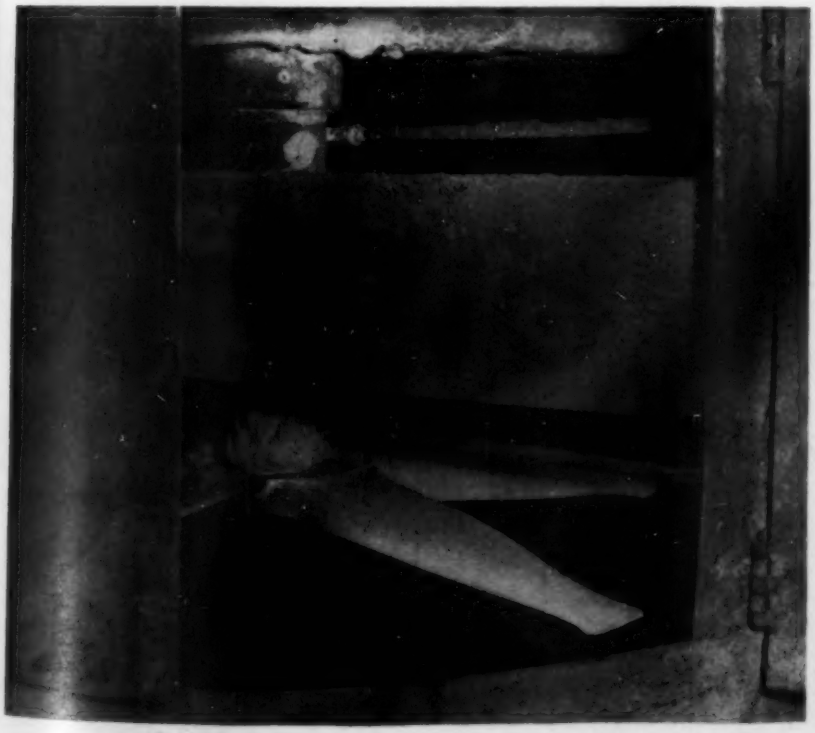


Fig. 12—Worn impellor for rotary centrifugal dust collector as crated for return to manufacturer for reblading.

Fig. 10—Worn propellor blades on fan handling air containing abrasive dust.



Both the tube and the envelope type bag will eventually increase in resistance from the accumulation of lint in the cloth pores. This lint often comes from grinders' gloves and from paper used to clean safety glasses. When the cloth resistance increases sharply, the bags should be replaced. Otherwise, ventilation volume drops and rapping devices wear holes in bags which in turn lets the abrasive dust wear out the fan.

Even if lint does not get into the cloth pores, the mechanical cleaning will wear out the bags in approximately two years of twenty-four hours daily service. Figure 11 shows a worn bag in contrast to a new bag. Continued rapping has worn the old bag thin enough to permit dust to pass through in a few areas.

Rapping devices should be kept in good order to keep the collector bags

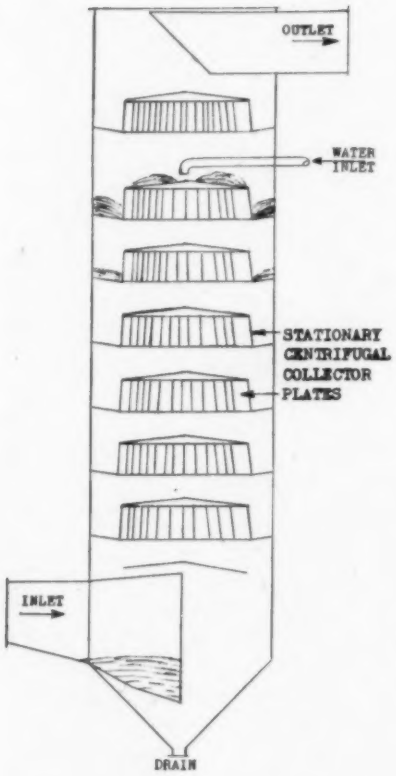


Fig. 13—Multistage wet centrifugal dust collector showing points where deposits may build up.

in full use. Cloth collectors should be shaken down at regular intervals such as lunch hours and at shift changes with fan turned off. The collected material hoppers should be emptied regularly to prevent unnecessary dust recirculation against the cloth sacks. The dirty air inlet must be kept baffled to prevent the air stream from impinging directly upon the collector sacks.

Rotary Centrifugal Collector Maintenance. The rotary centrifugal type of collector combines collector and fan functions. For light dust loads they will operate for long periods with only routine lubrication and general inspection. For heavy abrasive dust loads the rotary centrifugal fan collector should be provided with liners.

These liners should be checked periodically to prevent excessive damage to the main collector housing in case the liners wear through. By the time a set of liners has worn through, the fan impellor probably will need reblading. Prompt reblading service is provided by the manufacturer of this equipment.

Figure 12 shows a worn impellor crated for return to the manufacturer for reblading. For heavy dirt concentrations, a precleaner arrangement reduces the wear on the fan impellor and housing. The dust hopper for the rotary centrifugal collector should prevent any recirculation of the dust. Empty the hopper at frequent intervals.

Wet Collector Maintenance. There are several wet type dust collector arrangements available for foundry use. Their basic principle of operation is to mix the dirty air with a water fog or spray in order to wash the air. Some form of moisture elimination is then employed to separate the water and dirt mixture from the air stream. If rotary impellors are used, the impellors must be kept clean to prevent vibration. If baffles only are used, they should be checked periodically.

The eliminators must be kept in good order to prevent extra carryover to the fan which may get out of balance from solids built up on the impellor. Where a mechanical sludge conveyor is used, the conveyor will need routine checking and occasional repair and replacement.

Do not put core oil fumes through a wet type rotary centrifugal collector. The fumes condense on the fan impellor, where they collect dirt, and throw the impellor out of balance.

Where a series of vertical centrifugal collector sections are used in a wet collector, adequate clean water must be passed through the collector to keep it flushed clean. If the water must be recirculated, adequate settling tanks or basins must be provided. Dirty water deposits sludge in the collector sections, as shown in Fig. 13. It also plugs the water lines and the drain, and wears out the circulating pump. Dirt also tends to collect at the collector inlet as shown in Fig. 13. This dirt must be removed at regular intervals from both the collector sections and inlet. Provide adequate clean water, and this cleaning problem will be reduced to the occasional cleaning of the collector inlet.

Air Filter Type Collector Maintenance. Figure 14 shows a grinding bench equipped with six sectional type air filters. These filters are removed every second day, cleaned in oleum, sprayed with SAE 10 lubricating oil, and reinstalled. This maintenance is done on the night shift when the grinding benches are not in use, but a spare set of filters with the portable dipping and draining equipment would make it possible to do this work at any time.

Summary

Proper dust control equipment maintenance consists of good routine care and inspection plus replacement of worn parts before equipment failure occurs.

Fig. 14—Down draft grinding bench showing filter cleaning equipment.



MALLEABLE FOUNDRY CORE SAND PRACTICE

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ALMOST ALL SAND used in this shop for making cores is obtained from local deposits of bank sand having an A.F.A. fineness number of about 70. Figure 1 shows the sub-angular grain of this sand as compared with the round grained Ottawa silica sand.

Deposits, or pits, of bank sand occur in a relatively thin layer ranging from 5 to 15 ft. in thickness, and are covered with an overburden of topsoil. In order to extract uniform bank sand from a pit, it is essential that topsoil be thoroughly stripped away and that the proper depth and area are not exceeded in removing the bank sand. Excessively coarse sand comes from the edges and bottom of the pit.

Selection of New Pits. When a pit becomes depleted, a suitable new pit must be selected. Due to the very drastic effects which a change in sand grain size and distribution can have in coreroom practice, the selection of a new sand pit is a task requiring care and judgment. In such a case it is the custom for the sand supplier to submit a series of sand samples from prospective pits.

Grain distribution curves for these samples are determined by the laboratory and the most suitable one selected. It is the usual practice then for the sand supervisor and coreroom supervisor to visit the prospective pit, taking a new set of samples at various locations and depths.

Sieve analyses of each sample and also a composite sample are made by the sand laboratory. These are reviewed carefully. If they are suitable, and the sand is satisfactory from other standpoints, the pit is approved for use.

Control of Incoming Sand. Sand is hauled to the plant by truck in loads of from 10 to 35 tons. Figure 2 shows a plan view of the sand storage building. Trucks are able to dump the raw bank sand through an opening in the sidewall of the building into a pit which is called the truck dump.

From this dump the sand may be picked up by the 5-ton traveling crane bucket and placed in either the wet sand storage bin or, more often, directly into the 30 ton wet sand hoppers feeding the sand driers. Each of these wet sand hoppers is equipped with a heavy 4-in. mesh screen which prevents excessively large roots and other foreign matter from getting into the feeders.

In a foundry producing large volumes of castings, the econmical production of suitable cores constitutes a major problem. In the plant with which the author is associated, this fact has been recognized and a department with ample floor space, equipment, and staff has been developed to do the job. Likewise, it has been recognized that exacting control is necessary to permit the making of a large volume of cores having the proper physical properties.

Control of the incoming sand is exercised in two ways. First, a visual check is made of truckloads for excessively coarse sand, unusual amounts of lime, and presence of topsoil. Second, a grain distribution curve is plotted from a daily composite sample of sand taken at the sand drier feeders.

By means of these checks, any changes in the sand with respect to grain size, distribution, and purity may be ascertained at an early stage. Figure 3 shows a typical grain distribution curve for the bank sand. For a closer analysis of the distribution, it is the practice to use the additional sieves shown.

Sand Drying. Incoming sand from the pits varies in moisture content from 3 to 9 per cent. Accurate moisture control of core mixtures can be exercised only if the sand is dry, and cool. Raw bank sand is therefore dried and cooled before using.

Sand drying equipment consists of two parallel counterflow type driers which reduce the moisture of the sand to approximately 0.2 per cent. Each drier will dry and cool 8 tons of sand (7 per cent moisture) per hour. Increased tonnages can be run through the driers if moisture content of raw sand is lower than 7 per cent, and vice-versa.

Figure 4 shows a diagrammatic sketch of a drier and controls. Air for combustion is furnished by a small blower. The volume of air-gas mixture supplied to the firebox is

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regulated by a dual proportioning type valve which responds to the firebox temperature controller.

Surrounding the firebox is a space through which a large volume of air may be blown by the secondary blower. This space is baffled in such a manner that the secondary air must travel spirally around the box. A portion of the secondary air is allowed to enter the box through adjustable louvres at the forward end.

At the far end of the firebox the combustion products mix with the preheated secondary air and pass into the mandrel; they then double back between the mandrel and the outer drum where they are in contact with the sand.

Sand is fed into the outer drum of the drier by means of an electrical vibrating feeder. As the drum rotates, the sand is picked up by blades which lift it to the top of the drum, from which point it drops in a curtain to the bottom of the drum. Thus the sand, as it proceeds toward the mid-point of the drum, is exposed to the heated gases.

The sand and hot gases move in opposite directions (counterflow) in the heating section. As the gases proceed toward the feeding end, where they encounter wetter and colder sand, the moisture content of the gases increases and their temperature falls until they are saturated. The gases discharge from the stack at a temperature of less than 212° F. Such conditions make for high efficiency of the drier.

When the sand reaches the midsection of the drier, it is picked up and dropped into special chutes through the baffle plate into the cooling section. At this point the moisture content of the sand is about 1½ per cent. This amount of moisture is needed in order to take advantage of the cooling action arising from the evaporation of the moisture from the sand.

Cold air is blown through the mandrel in the cooling section by means of a separate blower which has been set to run at a desired constant volume. The air stream moves between mandrel and outer shell parallel to the sand flow.

Fines Removal

By adjusting the velocity of the air, fines may be carried out of the sand in varying amounts as desired. Fines, so removed, are carried by duct into a wet collector unit. Sand is discharged from the cooler at temperatures between 95° and 110° F., directly into an elevator which carries it to second floor for screening.

Operating control of the drier may be briefly described as follows: As the sand passes through the chutes from the heating to the cooling section of the drier, it drops onto a thermocouple. As shown in the sketch (Fig. 4), this thermocouple is connected to a control instrument which operates a butterfly valve on the secondary air blower intake.

Control instruments are set at a specific temperature, usually between values of 140 and 160° F. If the temperature, as measured by the thermocouple, falls, the damper on the secondary air blower opens more and thus increases the volume of cold air blown around the firebox. In turn, the firebox control instrument increases the supply of combustion gas and air due to the lowering of the firebox temperature, which normally is held at 1400° F. The increase in volume of heated air supplied to the heating section brings conditions at the midpoint of the drier back to normal. If the sand temperature at the midpoint rises, the reverse of the above actions take place.

To assist in control adjustments, temperatures of sand samples taken at the discharge end of the drier are recorded hourly. It is important to keep the sand discharge temperature between 95 and 110° F.

Sand at temperatures over 110° F. will be too hot for use, even after storage. Hot sand not only causes sticking of core mixtures in core boxes, but defeats the very purpose of drying sand by causing loss of moisture control of the mixtures through increased evaporation at the higher temperatures. Sand coming from the drier at temperatures less than 95° F. will not be thoroughly dried.

Screening. As the sand is discharged from the drier, it is taken by elevator to the second floor of the storage building where it is discharged on 20-mesh vibrating screens set at an angle of 33°. A mechanical spreader device distributes the sand evenly to permit effective screening.

Large sand grains, roots and other undesirable materials pass over the screens into the tailing chute. A periodic check of grain distribution of the tailing sand indicates the efficiency of the screens. Figure 5 is a representative grain distribution curve of the tailings, showing the presence of few sand grains finer than 40.

The usable portion of the sand falls through the screens onto a belt

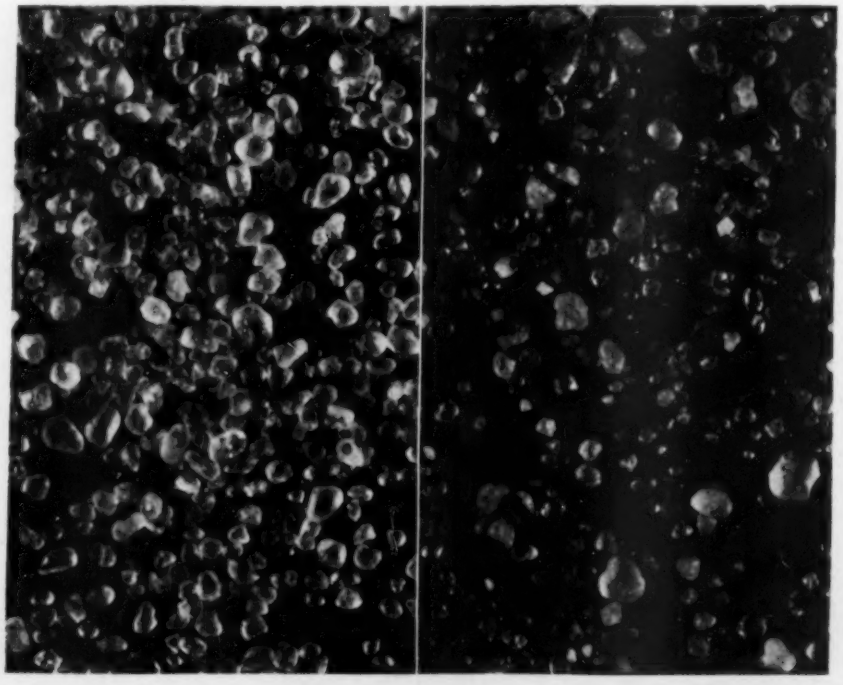


Fig. 1 (left)—Ottawa silica sand (washed and dried). (right)—Bank sand. 10X.

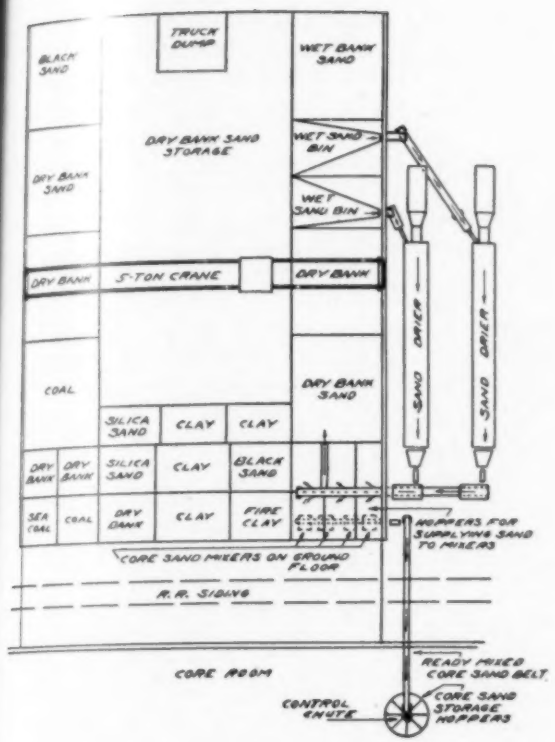


Fig. 2—Plan view of sand storage and preparation units.

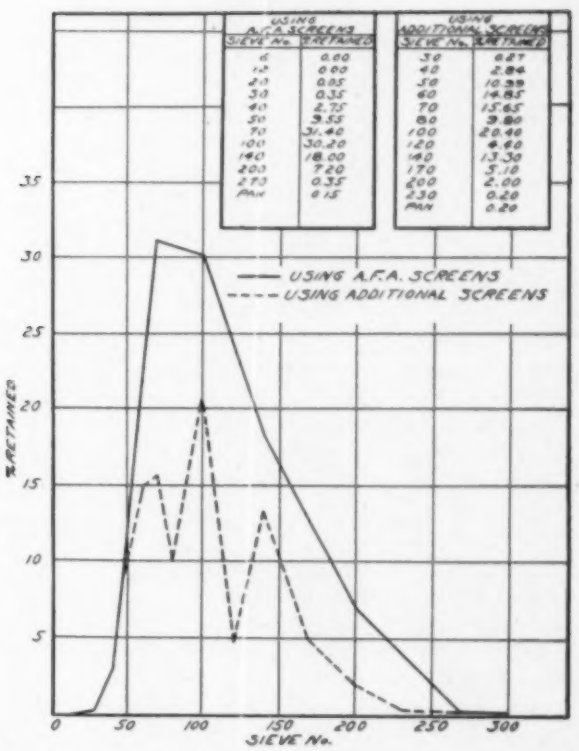


Fig. 3—Typical grain distribution curve for bank sand.

conveyor which discharges it into a 1000-ton dry bank sand bin. From this bin it is picked up by the clam bucket and transferred to a 3000-ton dry sand storage bin.

The purpose of the transferal is to blend the sand with the existing pile, thus smoothing out possible variations in grain distribution as the sand comes from the drier, and producing a more uniform sand for use. Sand from the large bin is finally transferred, as needed, to the steel hoppers above the mixers.

Selection, Storage, and Control of Other Raw Materials

Cereal Binder. Corn flour is used in the core sand mixtures to produce the necessary green compressive strength. In selecting a brand of flour, particular attention is paid to two characteristics of the flour. First, it is desirable to have a flour which develops the best green compressive strength, by test, for the least cost. Second, the characteristics of the flour must be such that core mixtures made with it have excellent "blowability" in a blow machine.

The latter characteristic is of special importance because so many of the production cores are made on blow machines. It might also be pointed out at this time that when all such factors affecting blowability are closely controlled, more and more hand-rammed jobs, and jobs of an intricate nature, can be converted to a blowing operation which, in a majority of cases, will increase production and lower costs.

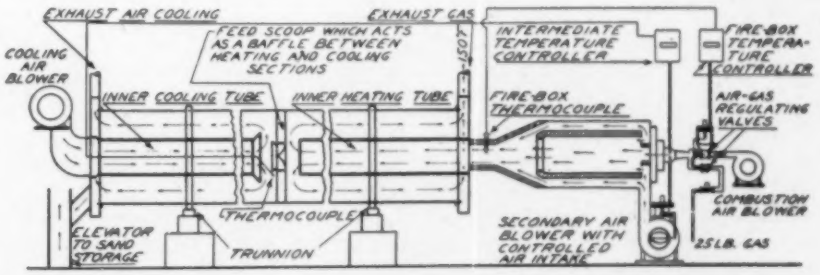


Fig. 4 (above)—Diagrammatic sketch of combination parallel and counterflow sand drier and cooler.

It has been the experience that when the lighter flours (10 to 13 lb. per cu. ft.) are used, the core mixture will blow much better than when the heavier flours (14 to 20 lb. per cu. ft.) are used. On this basis a maximum of 13 lb. per cu. ft. has been set up as an arbitrary figure for the acceptable density of flour.

Incoming cars of flour are checked for density, and for green compressive strength developed in an emperical mixture with standard sand.

Variations in flour from shipment to shipment can cause no end of trouble in the core room if precautions are not taken in segregating

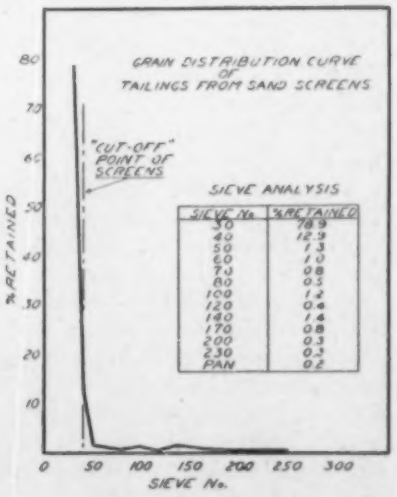


Fig. 5—Representative grain distribution curve of tailing sand.

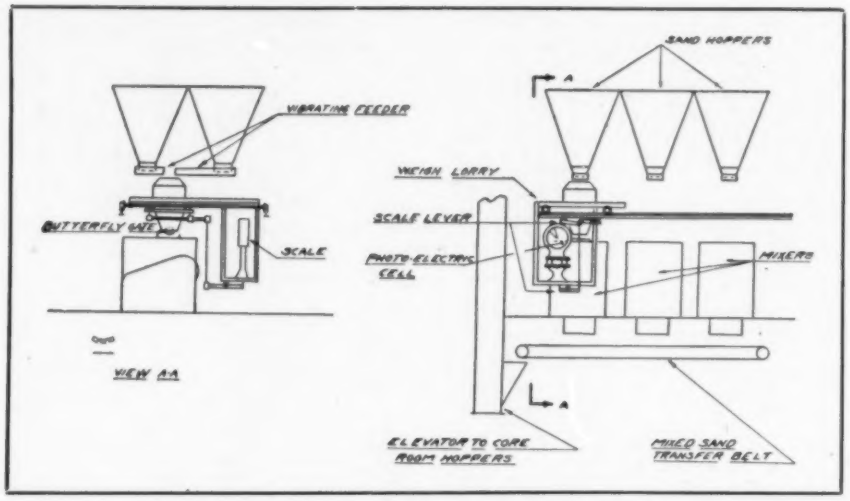


Fig. 6—Views of core sand weighing and mixing equipment.

the shipments. Moreover, if flour stands for long periods of time, or becomes damp, the green compressive strength developed by it is lower.

A system has, therefore, been established wherein the oldest flour is used first; incoming batches are piled in rotation. Flour adjustments in the core mixtures may then be made as needed, and consistent results obtained.

Core Oil. Liquid core binder, purchased in tank car lots, is used in all of the core sand mixtures. It is stored in an outside tank, from which it is pumped to the core sand mixing room in the sand storage building. The tank is equipped with steam coils to keep the viscosity of the oil at a normal level in the winter months.

In the selection of a core oil, a number of characteristics of the oil are considered:

1. Economy—Oil should be economical on a price-green compressive strength basis.
2. Uniformity—Shipment to shipment.
3. Baking — Oil should develop and maintain its maximum properties on the cycles and temperatures used in the production ovens.
4. Viscosity — Low viscosity oils are preferred due to their more thorough dispersion through the sand in a minimum mixing time.
5. Stickiness—Oil should produce a core sand mixture which permits drawing of core boxes without sticking.
6. Gas evolution—Oil should be one which develops a minimum of gas when metal is poured around the core in the mold.
7. Odor—Oils with offensive odors should be avoided.
8. Effect on skin — Oil with a strong tendency to cause dermatitis should not be used.

Routine tests are made by the laboratory on samples of each car of oil. The sand laboratory conducts two sets of tests. One of these consists of baked tensile tests made on specimens taken from a laboratory mix using standard sand and baked in the laboratory oven. The other is identical, except that the baking is done in production ovens. The latter test is a check on baking characteristics of the oil on the production cycles being used.

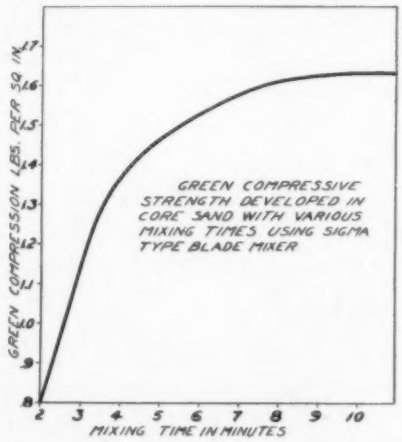


Fig. 7—Curve showing relationship of green compressive strength to mixing time (sigma blade type mixer).

Bentonite. This material is used in a few core mixtures where high green compressive strength is needed, and where the added hot strength is not detrimental. It replaces a relatively large quantity of flour in such mixes, but requires additional oil to produce a given baked tensile strength. The principal precaution in storing bentonite is to keep it dry to prevent lumping.

Iron Oxide. Iron oxide is used in several core mixtures to prevent "veining" in cored holes and, in general, produce cleaner cored surfaces in castings.

Core Sand Mixing Equipment, Mixtures and Mixture Control

Equipment. The core sand mixing equipment is located in the sand storage building directly below the series of steel hoppers containing the dried sand. These hoppers hold from 20 to 30 tons of sand each, and are refilled from the 3000-ton dry sand storage bin by means of the clam bucket.

Each of the hoppers is equipped with a vibrating electric feeder which is used to feed sand from the hopper into either of two weigh lorries (Fig. 6). Each weigh lorry is equipped with a scale and an electric eye device which will automatically stop the electric feeder on the sand hopper when the desired weight of sand is in the lorry. The lorry may then be moved along an overhead track to a position above

Table 1

TYPICAL CORE SAND MIXTURES

Mixture No.	Sand, lb.	Type Sand	Flour, lb.	Oil, qt.	Water, qt.	Bentonite, lb.	Iron Oxide, lb.	Moisture Standard, per cent	Green Compression Standard, psi.	Baked Tensile Standard, psi.
1	1800	Bank	19	11	19	—	—	1.7 -1.9	1.4 -1.55	140-160
2	1800	Bank	16	12½	14	—	—	1.25-1.4	1.3 -1.45	130-145
3	1800	Bank	14	34	14	—	—	1.25-1.4	1.3 -1.45	292+
4	1800	Bank	12	12	12	—	—	1.15-1.3	1.1 -1.3	100-125
5	1800	Bank	7	31	10	—	—	1.0 -1.5	0.95-1.1	292+
6	1800	Bank	23	13	21	3	—	2.0 -2.15	2.0 -2.2	120-140
7	1800	Bank	20	9	21	3	18	2.0 -2.15	2.0 -2.2	85-100
8	1800	Bank	14½	22	14½	1	11	1.3 -1.45	1.3 -1.45	250-275
9	1050	Bank	7	7½	7	—	11	1.2 -1.4	1.3 -1.5	110-130

any one of the four core sand mixers. Sand may be discharged from the lorry through a butterfly valve into any mixer.

The sand mixers are of the kneading type having two sigma-shaped blades. Each mixer will mix 1800 lb. of sand per batch. After a batch of sand has been mixed, it is discharged onto a conveyor belt below the mixers, taken up an elevator and then carried by belt conveyor to a cluster of storage hoppers in the core room.

An electrically operated chute above the cluster of hoppers is used to divert a particular type of mixture into the proper bin. The position of the chute is remotely controlled by push buttons in the mixing room. Core sand mixtures are transported from the storage hoppers to core machines by means of tote boxes on electric trucks.

Core Sand Mixing. Once the raw materials have been properly prepared, the duplication of a given core sand mixture, batch after batch, is largely dependent upon the accuracy of weighing and measuring the ingredients, and upon the mixing practice. In practice, all dry materials are weighed and all wet materials are measured by volume.

Quantities of the ingredients to be used for each of the various types of core sand mixtures are posted in the mixing room on a large blackboard. Consequently, no excuse exists for misunderstandings between operators, coreroom foremen, or sand laboratory supervisor with regard to the ingredients of a mixture.

Mixing prodecure is set up to:

1. Feed sand from the weigh lorry into the mixer, adding the weighed amount of flour to the stream.
2. Add any other dry ingredients such as iron oxide, or bentonite.
3. Mix dry for one min.
4. Add measured amounts of water and oil in that order.
5. Mix additional 7 min. (total 8 min.), using timeclock for obtaining exact time.
6. Discharge the batch.

Adequate mixing time is of prime importance. The curve (Fig. 7) shows the relationship between mixing time and green compressive strength developed in a core mixture when using a mixer equipped with sigma blades. A number of tests were made in developing this curve, to arrive at a mixing time which would be acceptable from three standpoints:

1. Uniformity of the batch.
2. Control of properties.
3. Economy in raw materials, particularly flour.

From the curve (Fig. 7) it can be seen that only a slight increase in green compressive strength may be had beyond the 8-min. interval, but that in a shorter time much less strength is developed. Also, mixtures made with a shorter time interval have less uniformity of green compressive strength.

Obviously, by increasing the flour, high strengths could be developed quickly. However, such a practice has the following drawbacks:

1. Non-uniformity of the batch.
2. Increased cost of flour.
3. Increased flour will demand more moisture and will:
   (a) Decrease blowability.
   (b) Increase baking time.
   (c) Evolve more gas in the core when metal is poured about it.

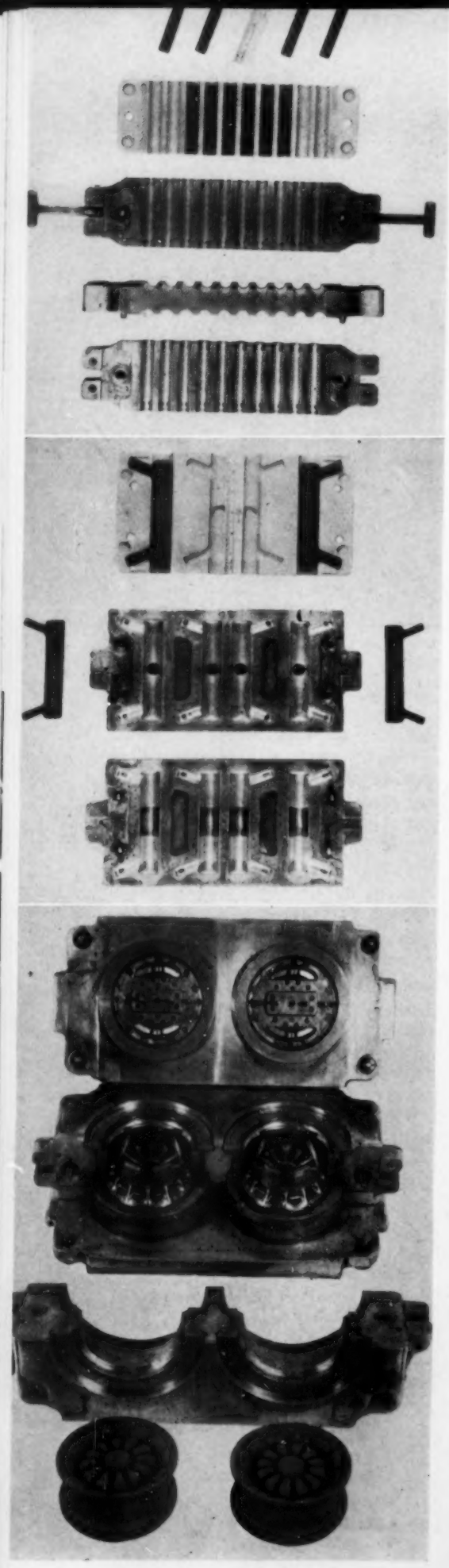
Through the design of the equipment, selection and control of raw materials, core sand mixture control has intentionally been taken as much as possible out of the operator's hands. It has long been the feeling that control should never be based on the judgment, personal whims, or



Fig. 9—Making a differential case core on a special machine equipped with air-squeeze cylinder.



Fig. 8—Making a differential core on a special machine adjacent to core oven.



sense of "feel" of an operator, but should, on the contrary, be based on scientific procedures.

A number of standard core sand mixtures have been developed to take care of the major core production requirements. For these mixtures, standards have been set up for moisture, green compressive strength and baked tensile strength. Representative mixtures and the standards to which they are made are shown in Table 1.

On each shift samples of sand from each type of batch made are tested for moisture, green compressive strength and baked tensile strength. The baked tensile tests are made on briquettes baked in the standard laboratory oven at 420° F. for 1½ hours.

Briquettes are also put through the production ovens at frequent intervals as a check on production baking. On the basis of laboratory tests, adjustments are made by the sand supervisor in the various amounts of ingredients in the core mixture to maintain standards established.

An alteration in a standard for a mixture is seldom made, and then only after a thorough investigation conducted by the sand supervisor and coreroom supervisor has indicated the need for the change.

Coremaking Equipment

Efforts to control core sand mixtures as a means of speeding production of cores are nullified if core room equipment and methods are antiquated. About ten years ago our core room equipment and methods underwent a severe revamping. Further improvements are being made.

Coremaking Machines. Production coremaking machines used in the coreroom may be placed in two general classes:

1. Core blowing machines.
2. Special core machines designed in the plant, including roll-overs, roll-arounds, etc.

Core blowing machines are of a standard make, but have been altered to suit particular needs. Designs of the various special machines, on which many of the cores are made, are based on motion-time analysis studies of individual core jobs. Some of the features incorporated in such designs are:

1. Elevation of the sand to a position above the box so that it may be scraped into the box for filling rather than scooped up and lifted in.
2. Use of grating for removal of excess sand from bench.
3. Counterweighting and pivoting of core boxes and equipment for ease in rolling over or lifting.
4. Convenient and definite placement of auxiliary materials used in making the core (rods, chills, etc.).
5. Convenient and definite placement of all working tools such as rammers, hammers, vent wires, air hoses, etc.

Such features of design, together with the instruction of the operator in the prescribed motion paths, have been instrumental in securing high production and uniformity of cores with minimum fatigue of the operator. Figures 8 and 9 show two types of special machines being used at the present time.

Core Boxes, Driers, and Plates. The major portion of the core boxes used are made of cast aluminum with brass parting strips. Steel plates and inserts are employed in positions on the boxes where wear or erosion is encountered. Some of the boxes used in production of large quantities of small cores, and which may be suitably counterweighted in the machine, are made of gray iron.

Core driers are made of cast aluminum. Two kinds of core plates, steel and asbestos fibre boards, are in general use. Figures 10 to 15, inclusive, show a series of core boxes, driers, and cores representative of those in use at this plant.

Reading (left) top to bottom:
Fig. 10—Three-section core box for blowing small pin cores. Aluminum drier.
Fig. 11—Core box, drier, and cores used in making an oil pump screen cover support.
Fig. 12—Four-section core box for blowing 4½ in. diesel piston head cores.

Core Baking

Ovens designed for the greatest baking efficiency and speed are needed to produce large volumes of cores at low cost. Heating equipment for the ovens must be designed to envelop every core with a rapidly moving stream of hot air which is at a temperature sufficient to bake the core rapidly and yet not so high that

it will bu[illegible] the outside of the core before the inside is baked out.

Figure 16 shows a diagrammatic sketch of a batch oven equipped with a heater of the recirculation type showing the path of the hot gases. The heating equipment used on the batch type and also the conveyor type ovens in the coreroom employ this system.

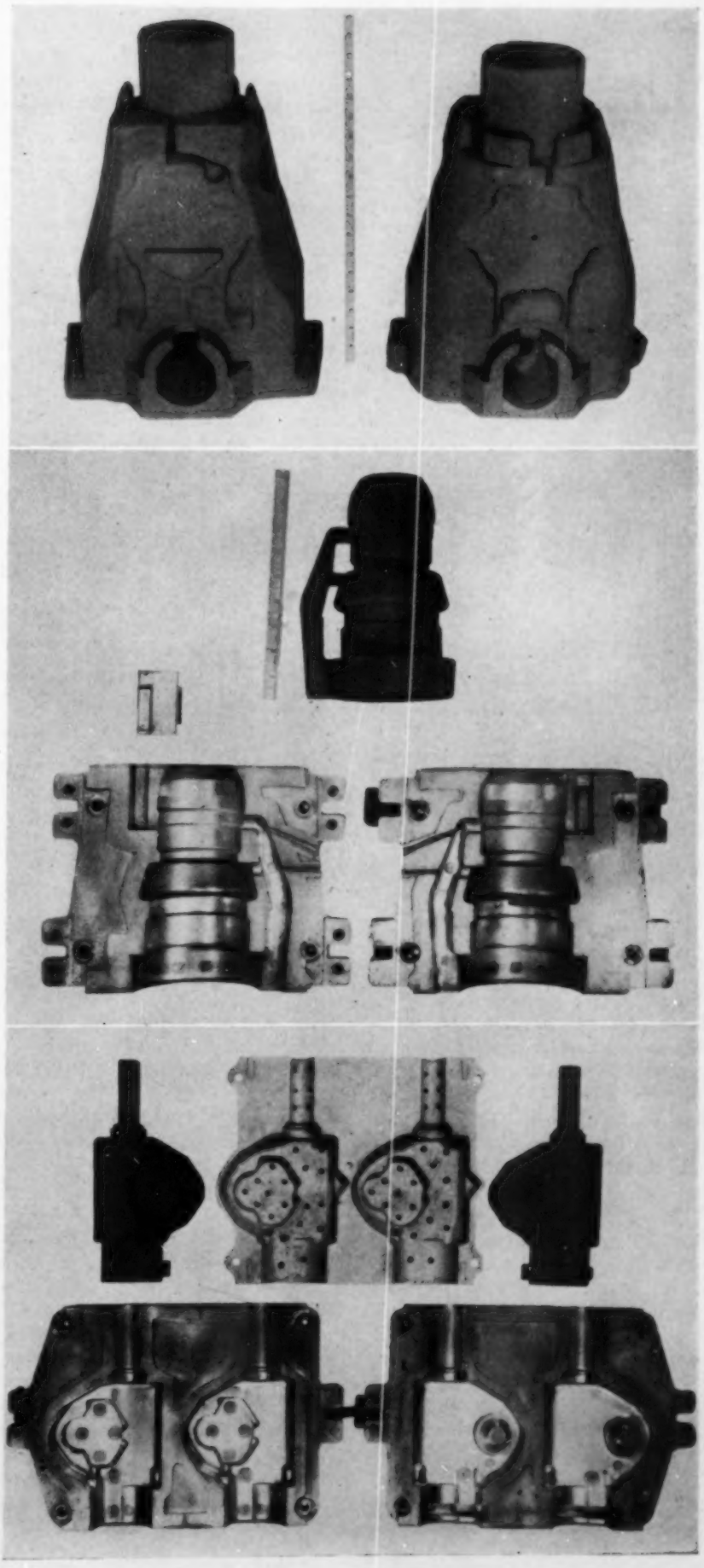
The key to fast baking in a system of this sort lies in having plenty of capacity in the recirculating fans to produce rapid change of gases in the oven in order to transfer the heat to the cores at low delivery temperatures (low temperature head).

Selection of the type of oven to be used is influenced by a number of factors such as the type, size, and also number of cores to be made, the methods of making and handling cores, and the particular layout of the coreroom with respect to floor space. In this particular foundry all three of the common types of ovens (horizontal, vertical, and batch) can be used to good advantage.

Horizontal Oven. A new horizontal oven, similar in design but of greater capacity than the old one, has been installed recently. Inasmuch as a great many coremaking stations may be placed next to the conveyor, core handling and breakage costs are reduced to a minimum with this type of oven.

Its principal use in this shop is for small cores which will bake out on a fast cycle. Blow machines and other production machines are arranged along two sections of the conveyor, as shown in Figs. 17 and 18.

The oven proper is approximately 96 ft. long, by 13 ft. wide, and is elevated to allow the use of floor space below it. A steel shroud extending beyond the exit door of the oven has been provided to permit exhausting of much of the objectionable smoke arising from freshly



Reading (right) top to bottom:

Fig. 13—Large differential carrier cores, representative of largest cores made in this foundry.

Fig. 14—Core box, and core for making pinion bearing cage casting.

Fig. 15—Steering gear housing core box, drier, and cores.

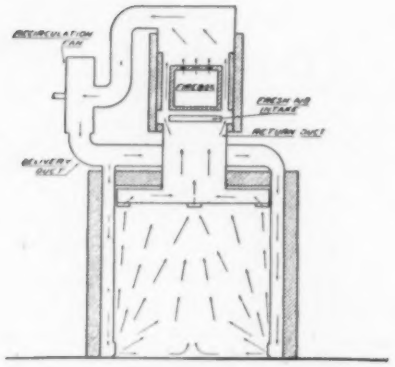


Fig. 16—Batch type core oven equipped with recirculation type heater.

part of the baking section governs the firing of these heaters. The other three heaters are similarly grouped on one manifold and are fired in response to a control thermocouple located in the midpoint of the baking section.

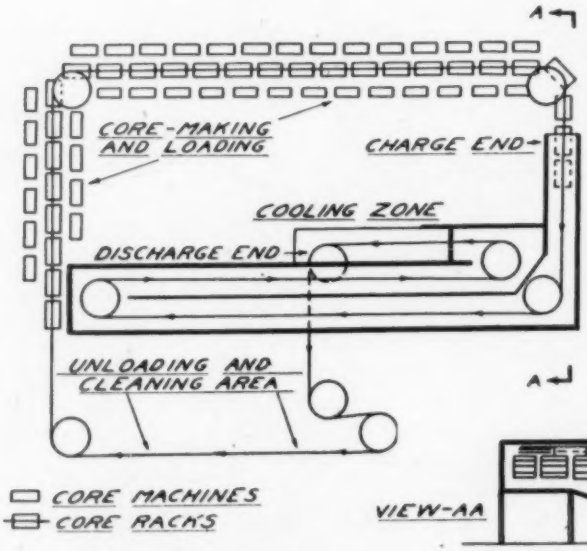
Use of two control points provides a close temperature control within the oven, and also is particularly valuable during stopping, starting, breakdowns and lunch periods; these interruptions in the cycle cause serious upsetting of temperatures in certain portions of the oven which, if not so adequately controlled, would result in underbaked or overbaked

proximately 70 ft. in over-all height and occupies a floor space 12x15 ft.

Of the 26 racks on the conveyor, 20 are in the oven (15 in baking zone, 5 in cooling zone); of the remainder, four are in the pit, one in the loading position, and one in the unloading position. Each rack will carry a total of approximately 1,600 lb. of cores, plates, and driers.

Conveyor travel is intermittent and is regulated by a timer and limit switch in the same manner as the horizontal oven.

In the case of the vertical ovens, the heater is built-in rather than being external. However, the method



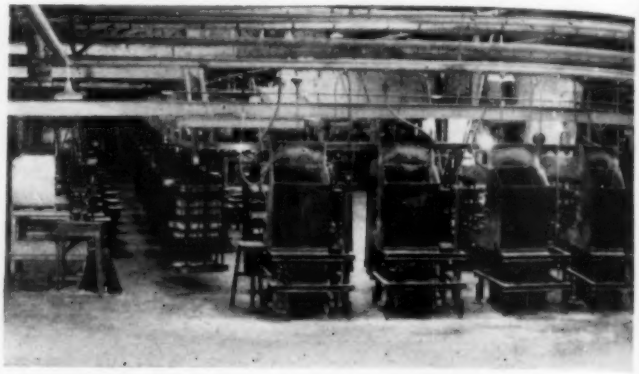


Fig. 17 (left)—Plan view of horizontal core oven.

Fig. 18 (above)—View of coremaking station and conveyor for horizontal core oven.

Fig. 19 (below)—Unloading baked cores from conveyor racks of horizontal core oven.



baked cores, and to speed core cooling to permit handling.

Of the 89 racks on the conveyor, 37 are in the baking section of the oven, and the remainder in the cooling section and at floor level in various stages of loading and unloading. Each rack is designed to carry an average load of 500 lb. of cores in addition to the weight of plates and driers. Figure 19 shows a close-up view of the racks.

Conveyor travel is intermittent, the distance which the racks move each time being equal to the distance between rack hangers. Each movement is started by an electric repeating interval timer, and is stopped by a limit switch actuated by the passage of another rack.

The oven is equipped with five external heaters of the recirculation type previously mentioned. The first two heaters are supplied from a common gas manifold. A control thermocouple located in the forward

cores. Normal temperature settings of these controllers are 450 and 470° F.

Baking time in the oven is governed by the setting of the repeating interval timer, which, as previously mentioned, determines the frequency of movement of the racks. The normal timer setting of 2 min. means about 1¼ hours $\left(\frac{37 \times 2}{60} = 1\frac{1}{4} \text{ hr.}\right)$ baking time in the oven.

Vertical Core Ovens. Three vertical core ovens of almost identical construction are in use (Fig. 20). These ovens have been especially economical in the production of medium and large size cores such as differential carrier cores; in this case the tonnage production per coremaking machine is sufficiently high to permit full loading of the oven directly from a few machines located in the limited space at the entrance.

Each of the vertical ovens is ap-

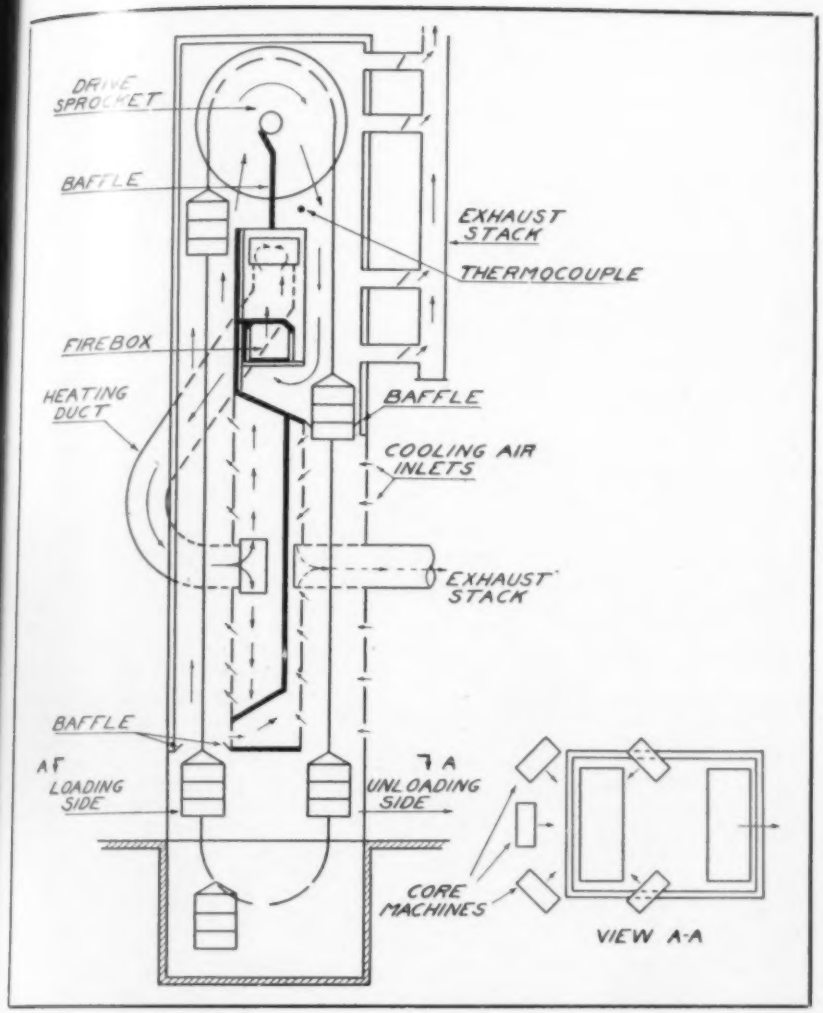


Fig. 20—Cross sectional views of vertical core oven.

of recirculation of gases is similar to that of the other ovens. Firing of the twin burners is regulated automatically by a single oven thermocouple and temperature control instrument. Three-quarters of the space in the oven is devoted to heating and one-quarter to cooling. Air for cooling is pulled into the oven by means of a large fan.

Normal oven control temperature settings (when making the differential carrier and similar size cores) range from 410 to 420° F. Normal cycle time is 4 hrs. (3 hr. heating, 1 hr. cooling). The setting of the timer for this cycle is 6 min. (the conveyor moves twice for each rack, $\frac{2 \times 20 \times 6}{60} = 4$ hr.)

Batch Ovens.—Six batch ovens equipped with recirculation type heaters are in service (Fig. 17). Automatic temperature controls regulate the oven temperature and keep the delivery temperature within safe limits. The principal use of these ovens at present is the baking of extra large cores, or odd lots of cores, not suited to the cycles being run on the horizontal and vertical ovens.

Conclusion

The author has intentionally devoted a major portion of this article to description of this foundry's equipment and to basic controls. It is his personal opinion that in many cases too little consideration is given to the design and control features of coreroom installations. In such instances, it is next to impossible to exercise the basic controls essential to the process, despite use of the most elaborate testing equipment that money can buy. On the other hand, when the installation is good, and it is well operated, a comparatively few well chosen control procedures are sufficient to maintain a high standard of control.

DISCUSSION

Chairman: R. J. ANDERSON, Belle City Malleable Iron Co., Racine, Wis.

Co-Chairman: D. I. DOBSON, General Malleable Corp., Waukesha, Wis.

F. L. HARRIS[1]: I understand the sand mixes you referred to were practically all bank sand mixes.

MR. CLARK: Yes, that is correct.

MR. HARRIS: Have you any idea what hot strengths you get with the different sand mixes?

MR. CLARK: No, I do not.

MR. HARRIS: I notice you use a lot of oil in some of your mixes. In some you use bentonite which increases the hot strength. I am wondering what we are going to do when corn flour becomes unavailable.

MR. CLARK: Frankly, I do not know.

MR. HARRIS: Do you know from experience that the tensile strengths of those mixes are the results you are looking for? Do you not have excessive hot tears?

MR. CLARK: That is right. We gage the mixture by the castings we are making in the foundry. We try to carry as high a tensile strength as we can without causing cracking and hot tearing, because it does allow us to handle the cores with a minimum of scrap loss due to handling.

MR. HARRIS: That is what most of us have done, but now we are confronted with the possibility of running out of corn flour. Recently we started some tests to see if there was a substitute material that would result in the same or nearly the same properties as effected by corn flour.

We knew only from experience that the cores were all right, but we did not know what the core hot strengths actually were. We have since found out what the hot strengths were. Substitute material is available for the necessary green strengths, the majority of them, however, result in high collapsibility and hot strength. Substitute materials do not equal the green strength and collapsibility properties of corn flour. It is going to be a serious problem for those who have high collapsibility in the cores and produce castings with heavy and light sections combined.

MR. CLARK: That is right. Many foundries get much of the baked tensile strength from the corn flour, which also promotes collapsibility.

MR. HARRIS: You mentioned that in your horizontal oven the cores go through a 1¼-hr. cycle. At what temperature?

MR. CLARK: That is 450° F. on one thermocouple and about 470° F. on the other. Those are standard temperatures in the oven.

MR. HARRIS: Is that on the smaller core?

MR. CLARK: That is on small cores.

MR. HARRIS: On the larger cores I think you mentioned a 2 or 3-hr. cycle.

MR. CLARK: The larger cores are baked from 2 to 4 hr. at 420° F.

MR. HARRIS: Would that include carrier cores?

[1]Belle City Malleable Iron Co., Racine, Wis.

MR. CLARK: The carrier cores would run about 4 hr.

MEMBER: Do you make a daily grain size check on this composite sample? What would you do should the distribution deviate too far from the desired point?

MR. CLARK: We would have to examine the curve and determine how far it shifted out of line and whether the shift came on the grain sizes or at a point in the curve where it was going to give us trouble. In plotting the distribution curve we particularly watch the shape of it. If we had the day's run of sand in the bins there would be little we could do about it, but use it and try to make adjustments accordingly in the core sand mixture. If it shifted very much we would immediately complain to the sand supplier.

MR. HARRIS: Was the mix which had around 31 or 35 quarts of oil used for your skim cores?

MR. CLARK: Yes.

MR. HARRIS: Do you bake that core in your horizontal oven or vertical oven?

MR. CLARK: In the horizontal oven.

MR. HARRIS: Do you make it with that much oil?

MR. CLARK: Yes. It is a very light core, less than ½-in. thick and perforated.

D. F. SAWTELLE[2]: I made a quick calculation on that core mix, and I think the percentage of oil is a little over 3 per cent by weight. Statements have been made by people investigating core oil mixes that cores with over 2½ per cent oil will not bake thoroughly. If that high amount of oil is used for strength, could not that same end be arrived at by using less oil, but perhaps a little more clay? Then you would have a core which would thoroughly bake and still give you the needed hot strength?

MR. CLARK: I do not know whether we have ever tried that. I do think that the skim cores are thoroughly baked. They are a perforated core with a very light section.

Do you know whether we have ever tried using clay or bentonite in that skim mix, Mr. Morrison?

CHARLES MORRISON[3]: Yes, we have, but there is no point in it. The clay consumes more oil. The tensile strength drops, and you have to add more oil.

MR. CLARK: You would have to add more oil to take care of the clay.

MR. MORRISON: It makes a more difficult mixture to use in the core box, too.

MR. CLARK: That is made in a gang roll-over core box.

MR. SAWTELLE: You would be interested in strengthening a core like that at high temperatures. Do you think the mix, using the clay, would give you a higher strength at high temperatures, than the mix using oil and sand?

MR. CLARK: That particular mix is the only one that we have for skims. If we keep it up to the proper strength, we do not have disintegration. That core is an important core, because if it disintegrates during the pou[illegible]g, dirty castings result.

It has been very satisf[illegible]tory. At one time we used silica sand [illegible]stead of Ottawa sand in that particular mix, but we have gotten by very well with it, when we keep the strength up, using the bank sand and preponderance of oil.

MR. MORRISON: The reason we do not use clay in that core is because it is a 32-gang core box having small cavities. If we put clay in the mixture, we have to add additional water to bond the clay, which results in a mixture that will neither ram nor jolt properly in the core box. As a matter of fact, we have made this particular core out of a sand that has no flour or water at all. The last thing we want to use in it, even though we know it would give good hot strength, is clay, because it creates a difficult production problem in core making.

MR. CLARK: The green strength on that production job is down to a pound or less and the mix is quite dry.

MR. SAWTELLE: In steel foundry practice many times more oil is added to produce a stronger core. Oftentimes they go to such extremes that they are defeating the purpose. The core oil content becomes so high that the cores do not bake satisfactorily. That is not the recommended way to get hot strength.

MR. CLARK: That is true. We do not want to use any more oil than is necessary on cores merely from the standpoint of gas formation, particularly in steel foundry practice. A number of steel foundries use nearly all bentonite because it keeps the gas content down.

[2]Malleable Iron Fittings Co., Branford, Conn.

[3]Saginaw Malleable Division, General Motors Corp., Saginaw, Mich.



Fig. 5—Macrostructures of cast bronzes. Etched with 50 per cent HNO_3 solution. (A) Gun metal cast at 1150° C. (B) Gun metal cast at 50° C. above liquidus. (C) Nickel-tin (3 per cent) bronze cast at 1150° C. (D) Nickel-tin (3 per cent) bronze cast at 50° C. above liquidus.

MECHANICAL PROPERTY DISTRIBUTION IN SAND CAST BRONZES

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R. G. Hardy, CSp(X),
and B. M. Loring,
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Anacostia Station, Washington, D. C.



IT IS generally recognized that the mechanical properties of large castings are not necessarily equal to the mechanical properties of the separately cast coupon, and may vary from section to section in the same casting. This is especially true of castings of the tin bronzes which have pronounced coring and which also possess an inherent tendency for dispersed microshrinkage.

The composition of the bronze, foundry practice and the design of the casting appear to have important influences on the distribution of the microporosity, which in turn affects the mechanical properties. Little information is available regarding the correlation of these variables and the properties of tin-bronze castings.

The purpose of the present investigation was to study the effect of composition on the variations in mechanical properties in a heavy bronze casting. Five different bronzes were cast in heavy blocks and in separately cast coupons. The blocks were then sectioned so that corresponding locations could be studied with respect to microstructure, grain size, specific gravity, chemical composition and mechanical properties.

Previous investigations have been limited to single compositions of bronze because of the great amount of time and labor involved. Rowe[1], in his investigation of an alloy of 90 per cent copper—10 per cent tin, has shown that the effects of microshrinkage on properties may be minimized by a high rate of solidification. Rapid freezing gave finer grain size, greater hardness and higher density. Rowe favored high rates of solidification along with low casting temperatures.

Rahm[2], in his investigation of hydraulic bronze, reported that grain size had little effect on the mechanical properties and that the density was a more important factor. The investigation was made on castings

This paper was presented and discussed at a Brass and Bronze Session of the Fiftieth Annual Convention, American Foundrymen's Association, in Cleveland, May 10, 1946.

• ***An investigation of variations in the mechanical properties of various bronze alloys as affected by chemical composition. The opinions expressed herein are those of the authors and do not necessarily reflect the views of the Navy Department.***

2¾ in. square and weighing 50 lb. each. The best properties occurred at the corners of the casting and were below the tensile properties of the test coupon as separately cast.

Pearson and Baker[3] made a careful investigation of tin bronze (90:10) and phosphor bronze. The maximum density of the tin bronze was obtained by melting and casting in nitrogen and in a vacuum. Unsoundness was associated with the long freezing range of the bronze and was said to exist in the form of inter-dendritic fissures which increased in magnitude in sand cast (slowly cooled) bronzes. The variations in tensile properties in an ingot of tin bronze were related to concentrations of porosity.

Experimental Methods

(a) The Coupon and Casting

The separately cast coupon, shown in Fig. 1, is of the type required for gun metal, valve bronze and hydraulic bronze. It was gated at one end and two bars per mold were poured from a common runner.

The casting (Fig. 2) selected for the study of mechanical properties was a solid block, 8x6x3 in., horn gated as shown. The mold was titled at an angle of 10 degrees during pouring, with the riser end lowest. The mold was then tilted in the opposite direction at an angle of 10 degrees to the horizontal during the period of solidification.

There was a tendency for microshrinkage at the center of the block for certain compositions of bronze, showing that the test was selective. The nominal compositions of bronze studied were as follows:

	Composition, per cent				
	Cu	*Sn*	*Zn*	*Pb*	*Ni*
Gun Metal	88	9	3	....	....
Valve Bronze	89	6.5	3	1.5	....
Hydraulic Bronze	85	5	5	5	....
Nickel-Tin Bronze	88	6	3	....	3
Nickel-Tin Bronze	88	3	3	....	6

(b) Foundry Practice

The following metals were used in making the alloys: Grade A copper, electrolytic nickel, electrolytic slab zinc, Straits tin, and lead of high purity. The furnace charges consisted of the virgin metals, copper-nickel master alloy and scrap of known composition from previous heats of virgin metal. A lift coil induction furnace with a clay-graphite crucible was used for melting.

The surface of the molten bronze was exposed to the atmosphere, but the melting was rapid in order to



Fig. 1—Separately cast coupons of type required for gun metal, valve bronze and hydraulic bronze.

Table 1

CHEMICAL COMPOSITION AND MELTING PRACTICE—SAND CAST BRONZES

Heat No.	*Location of Specimen*	*Composition, per cent*					*Pouring Temperature, °C.*	*Deoxidation Addition, P-Cu, oz.*
		Cu	*Sn*	*Zn**	*Ni*	*Pb*		
1G	Coupon	88.23	8.87	2.90			1150	1
2G	A	88.60	8.86	2.54			1100	2
	B	88.76	8.37	2.87				
	C	88.33	9.22	2.45				
3G	A	88.16	9.74	2.10			1050	3
	B	88.51	9.39	2.10				
	C	88.31	9.86	1.83				
4G	A	88.12	9.52	2.36			1150	3
	B	88.41	9.22	2.37				
	C	88.22	9.54	2.24				
5M	Coupon	87.61	7.09	3.82		1.48	1150	1
6M	A	87.56	7.04	3.80		1.60	1150	1½
	B	87.90	6.74	3.90		1.46		
	C	87.82	7.05	3.59		1.54		
7H	A	84.88	5.35	4.52		5.25	1150	1½
	B	84.96	5.27	4.52		5.25		
	C	84.96	5.28	4.58		5.18		
8G-3 Ni	Coupon	87.61	6.08	3.26	2.83	0.22	1120	2
9G-3 Ni	A	88.91	6.43	2.34	2.32		1100	5
	B	89.05	6.06	2.55	2.34			
	C	88.66	6.56	2.47	2.31			
10G-3 Ni	A	88.21	5.25	3.60	2.72	0.22	1150	5
	B	88.28	5.10	3.69	2.73	0.20		
	C	88.24	5.27	3.54	2.73	0.22		
11G-6 Ni	A	89.06	3.26	2.54	5.14		1150	5
		89.05	3.12	2.65	5.18			

*Zinc by difference.

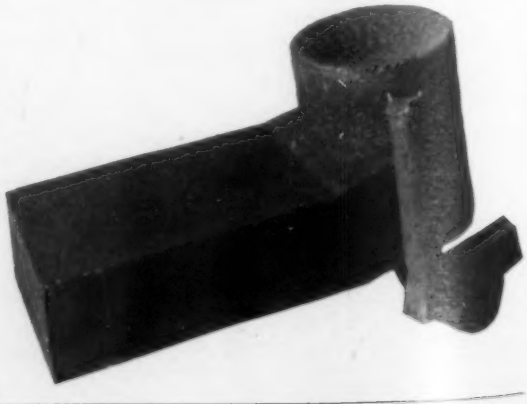


Fig. 2—Solid block casting, 8x6x3 in., horn gated, for study of mechanical properties of various bronzes.

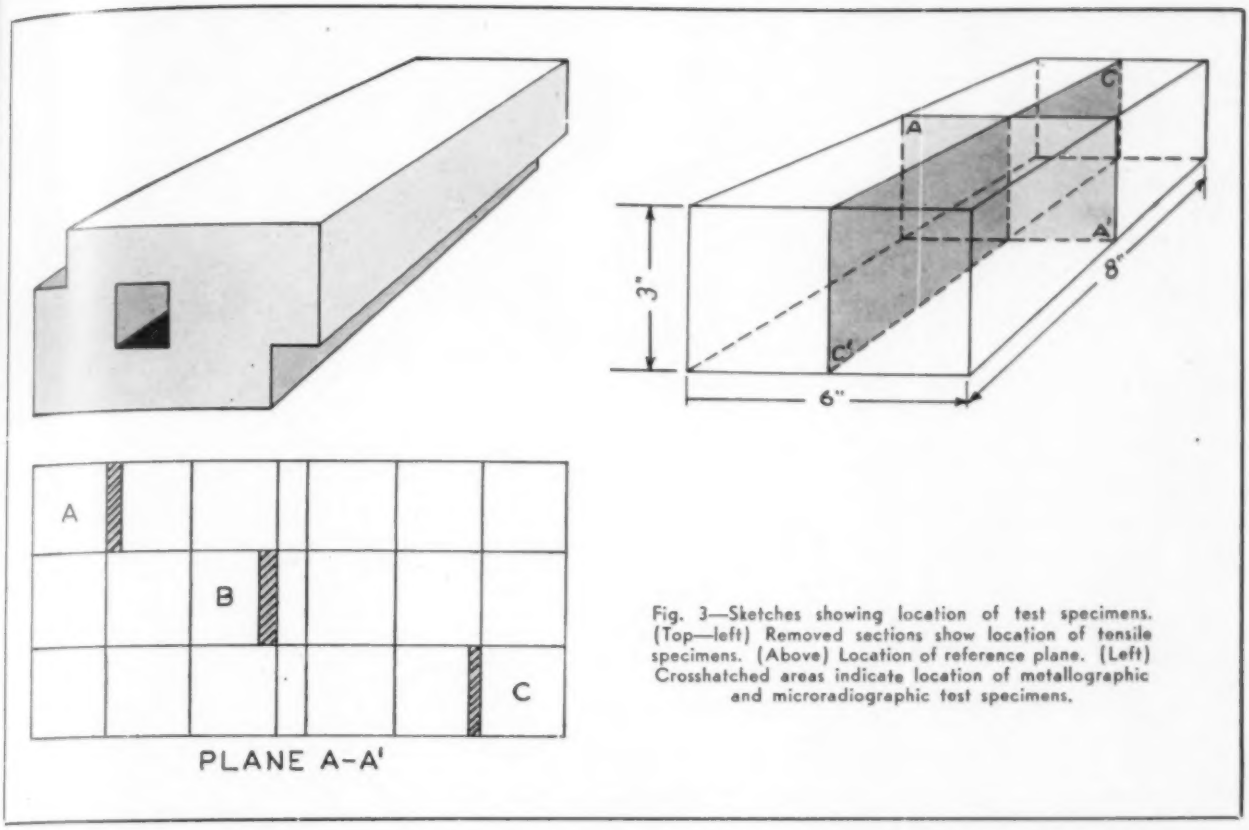


Fig. 3—Sketches showing location of test specimens. (Top—left) Removed sections show location of tensile specimens. (Above) Location of reference plane. (Left) Crosshatched areas indicate location of metallographic and microradiographic test specimens.

prevent excessive oxidation. At 1200° C. (2190° F.) an addition of phosphor-copper (10 per cent phosphorus) was made. Then the alloying elements were added to balance the composition, together with a second addition of phosphor copper (Table 1).

The bronzes were cast at 1150° C. (2100° F.), and several were cast at approximately 50° C. (90° F.) above the experimentally determined liquidus. The test coupon was cast first and the block immediately thereafter. The sprue was kept full so that the rate of metal flow into the mold cavity would be constant for each casting. The riser was covered with an anti-piping compound.

The molds were prepared from Albany green sand (Grade 1) with about 6 per cent moisture. The green compressive strength was 4 to 6 psi. and the A.F.A. permeability was 22. The molds were air dried for 24 hr. before casting.

(c) Investigation of the Block

Only radiographically sound blocks (penetrameter 1½ per cent) were sectioned for the mechanical tests. The material taken for the tensile specimens, as shown in Fig. 3, also served for Brinell hardness and specific gravity measurements. Samples from adjacent positions were used for the grain size study, chemical analysis, hardness survey and metallography.

An attempt was made to take all of the measurements as closely as possible to the gauge lengths of the respective tensile test specimens in order to insure good correlation. The Brinell hardness was taken on cross sections of the threaded ends of the machined tensile test specimens themselves.

The samples for chemical analysis came from the 1/16 in. thick layer

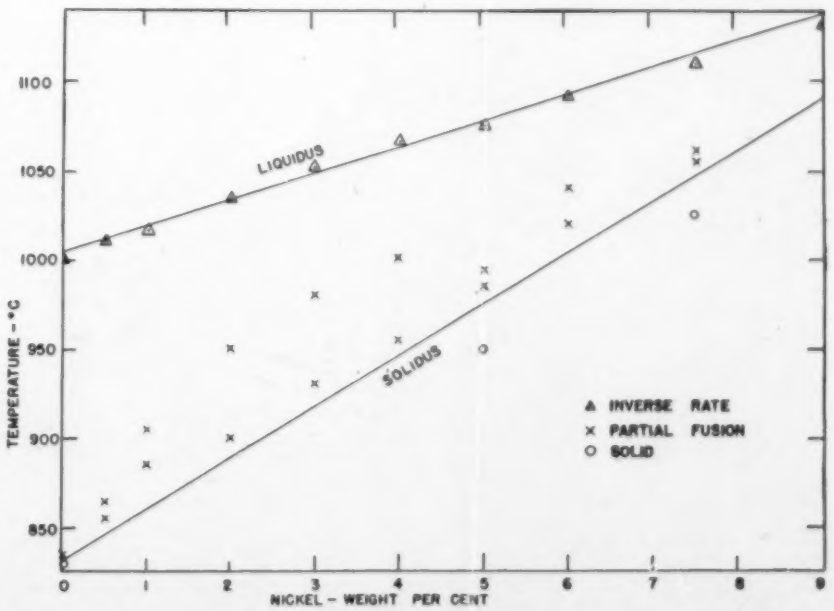


Fig. 4—Liquidus and solidus of Cu-Ni-Sn-Zn alloys of 88 per cent Cu, 3 per cent Zn, 9 per cent (Ni + Sn).

machined from around the periphery of the gauge length, and the specific gravity was determined from the machined tensile test specimens.

The specimens for microscopic examination were taken from a point not more than 3/16 in. from the center of the gauge length of the tensile test bars. After the photomicrographs had been taken, the etching on the specimens was removed by light polishing and the polished surace was then mounted face down on a steel holding plate.

The back of the specimen was then removed by milling and grinding until a thin slice 0.0025 in. thick remained. The thin slice was microradiographed with iron radiation and fine-grained spectroscopic film. By this means, the photomicrographs and microradiographs utilized were

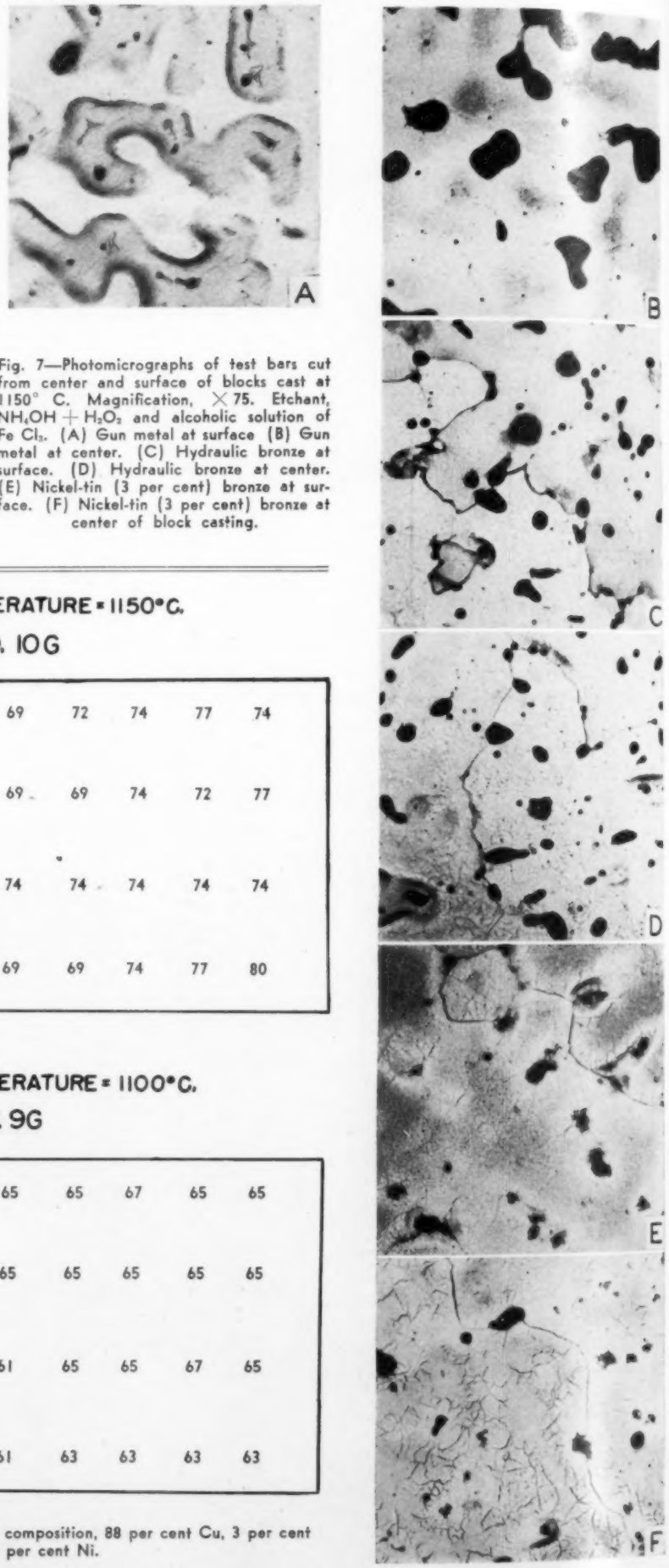


Fig. 7—Photomicrographs of test bars cut from center and surface of blocks cast at 1150° C. Magnification, × 75. Etchant, $NH_4OH + H_2O_2$ and alcoholic solution of $FeCl_3$. (A) Gun metal at surface (B) Gun metal at center. (C) Hydraulic bronze at surface. (D) Hydraulic bronze at center. (E) Nickel-tin (3 per cent) bronze at surface. (F) Nickel-tin (3 per cent) bronze at center of block casting.

POURING TEMPERATURE = 1150°C.

HEAT NO. 10G

	COPE									
RISER END	69	69	67	74	69	69	72	74	77	74
	69	67	67	74	69	69	69	74	72	77
	69	72	69	72	69	74	74	74	74	74
	72	69	72	72	74	69	69	74	77	80
	DRAG									

POURING TEMPERATURE = 1100°C.

HEAT NO. 9G

	COPE									
RISER END	67	65	65	67	65	65	65	67	65	65
	63	67	67	67	65	65	65	65	65	65
	63	63	63	61	61	61	65	65	67	65
	63	63	63	63	63	61	63	63	63	63
	DRAG									

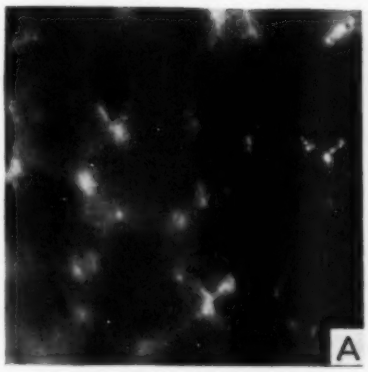
Fig. 6—Hardness survey of cast block. Nominal composition, 88 per cent Cu, 3 per cent Zn, 6 per cent Sn, 3 per cent Ni.

made of the identical bronze structures.

(d) Liquidus-Solidus Measurements

Since it had been reported that the long freezing range of tin bronzes is a cause of their unsoundness, the freezing ranges of various bronzes in which nickel was substituted for tin were determined. The nickel and tin totaled 9 per cent in this series. Two-lb. heats were cooled in graphite crucibles from a temperature 50° C. above the liquidus at a rate of 2° C. per minute.

The cooling curve for the liquidus temperature was determined to within 1° C. by the inverse rate method. However, the cooling curve for the solidus showed no clearly defined thermal arrest by this method. The solidus temperature



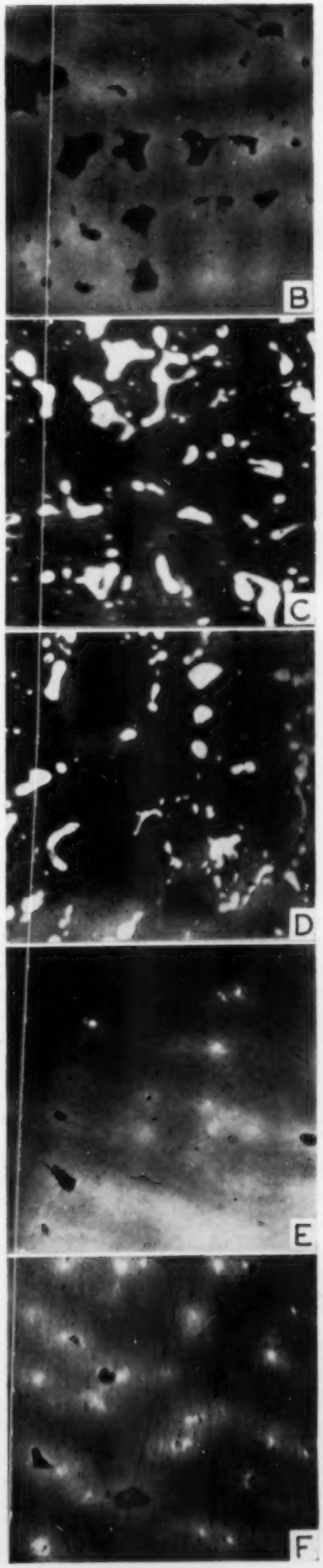


Fig. 8—Microradiographs of test bars cut from center and surface of blocks cast at 1150° C. Magnification, × 75. Iron radiation — 35 kv. (A) Gun metal at surface. (B) Gun metal at center. (C) Hydraulic bronze at surface. (D) Hydraulic bronze at center. (E) Nickel-tin (3 per cent) bronze at surface. (F) Nickel-tin (3 per cent) bronze at center of block.

Table 2

PROPERTIES OF SAND CAST BRONZE TEST BARS

Heat No.	Location of Test Bar	Pouring Temperature, °C.	Mechanical Properties: Tensile Strength, psi.	Elongation, per cent in 2 in.	BHN 500 kg. load 10 mm. ball 30 sec.	Specific Gravity at 20 °C.
1G	Coupon	1150	41,800	25.5	69	8.578
	A		42,400	55.0	78	8.679
	B		30,800	24.0	57	8.410
	C		43,900	40.0	69	8.664
2G	Coupon	1100	42,000	23.0	72	8.726
	A		45,200	60.0	74	8.789
	B		32,300	26.0	65	8.558
	C		44,100	35.0	72	8.736
3G	Coupon	1050	49,700	33.0	72	8.842
	A		41,700	23.0	77	8.787
	B		40,150	24.0	74	8.814
	C		53,700	68.0	74	8.824
4G	Coupon	1150	49,700	46.0	74	8.771
	A		49,700	63.0	74	8.691
	B		38,800	52.0	69	8.507
	C		45,800	54.5	63	8.706
5M	Coupon	1150	39,900	38.0	63	8.547
	A		38,800	36.0	63	8.613
	B		28,100	35.5	50	8.377
	C		39,100	50.0	80	8.710
6M	Coupon	1150	42,400	36.0	65	8.752
	A		39,800	66.5	61	8.796
	B		31,700	28.0	57	8.687
	C		41,300	45.5	59	8.804
7H	Coupon	1150	38,700	36.5	58	8.861
	A		35,700	36.5	57	8.855
	B		27,100	25.0	52	8.780
	C		31,100	26.0	56	8.852
8G-3 Ni	Coupon	1120	46,000	38.0	68	8.845
	A		43,400	50.0	65	8.851
	B		40,000	34.0	65	8.841
	C		48,400	43.5	69	8.883
9G-3 Ni	Coupon	1100	42,900	27.5	74	8.832
	A		49,900	29.5	72	8.883
	B		48,800	25.0	69	8.828
	C		49,050	57.5	77	8.878
10G-3 Ni	Coupon	1150	47,700	50.0	69	8.787
	A		43,400	50.0	69	8.718
	B		41,300	46.0	63	8.742
	C		43,800	47.5	69	8.750
11G-6 Ni	Coupon	1150	42,800	34.0	69	8.821
	A		43,750	38.0	72	8.807
	B		39,900	36.0	69	8.806
	C		41,600	38.0	72	8.822

was determined by microscopic examination for incipient melting in samples that had been cold worked and held for 2½ hr. at various temperatures before quenching in cold water (Fig. 4).

Some representative data showing the effect of pouring temperature on the grain size and the distribution of hardness in gun metal and nickel-tin bronze are shown in Figs. 5 and 6. Low pouring temperature is responsible for fine grain size and more uniform hardness throughout the block.

Discussion of Data

In Fig 5, the centerline at the junction of large grains disappears at low pouring temperature. In Fig. 6, the hardness is shown to vary not more than 6 Brinell numbers when the nickel-tin bronze is poured at a temperature of approximately 50° C. above the liquidus.

The influence of composition on the distribution of porosity is illustrated in Figs. 7 and 8. The photomicrographs were taken at positions *A* and *B* (Fig. 3) of each block. The microradiographs in Fig. 8 and the photomicrographs in Fig 7 were made of the same samples.

Porosity in both figures appears as black angular shapes. The microporosity which is evident in all three bronze compositions occurs adjacent to the low melting constituents in the interstices of the dendrites and grain boundaries. The microradiographs show further that lead in hydraulic bronze tends to fill these cavities. This is perhaps the reason for the high density of the hydraulic bronze at the center of the block.

Lead Distribution

The distribution of the lead which occurs as a network between dendrites in this location causes nearly as much loss in mechanical strength, however, as do the voids in the center of the heavy section of the gun metal. The nickel-tin bronze has the most uniform distribution of microporosity with no tendency for the intercommunicating type shown by gun metal and hydraulic bronze. The narrow solidification range of the nickel-tin bronze is a beneficial factor.

The differences in mechanical properties between the separately cast coupon and the casting largely depend on the composition and pouring temperature of the bronze. Gun metal, valve bronze, and hydraulic bronze, cast at 1150° C. (2100° F.) had a greater variation in mechanical properties than the nickel-tin bronzes with 3 to 6 per cent of nickel.

Lower casting temperature improved the soundness of gun metal and the nickel-tin bronzes, but the effect of differences in composition was still apparent. The tensile strength and elongation of selected heats (Table 2) cast at 1150° C. (2100° F.) may be summarized as shown in Table 3.

The nickel-tin bronze containing 3 and 6 per cent nickel has less variation in properties between the separately cast test coupon and casting and in different parts of the same casting. In casting heavy sections of this bronze, good foundry practice as well as proper deoxidation methods are important if adequate mechanical properties are to be obtained.

The pouring temperature should be low without causing cold shuts or low fluidity. A special attempt should be made to secure directional solidification in the heavy section and feed the section with adequate risers.

Acknowledgments

The authors wish to acknowledge the assistance of the Metallography Group and Radiographic Sections of the Division of Physical Metallurgy of the Naval Research Laboratory.

Table 3

TENSILE STRENGTH AND ELONGATION—SAND CAST BRONZES

Alloy	Heat No.	Separately Cast Bar Tensile Strength, psi.	Separately Cast Bar Elongation, per cent in 2 in.	Center of Block Tensile Strength, psi.	Center of Block Elongation, per cent in 2 in.	Top Corner of Block Tensile Strength, psi.	Top Corner of Block Elongation, per cent, in 2 in.
Composition G	1G	41,800	25.5	30,800	24.0	42,400	55.0
Composition M	6M	42,400	36.0	31,700	28.0	39,800	66.5
Red Brass	7H	38,700	36.5	27,100	25.0	35,700	36.5
Ni-Sn Bronze	10G	47,700	50.0	41,300	46.0	43,400	50.0
Ni-Sn Bronze	11G	42,800	34.0	39,900	36.0	43,750	38.0

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2. Rahm, A. M., "*A Study of the Distribution of Structural and Physical Characteristics Throughout Castings of Red Brass*," A.I.M.E. Technical Publication No. 1117, METALS TECHNOLOGY (1939).
3. Pearson, R. F., and Baker, W. A., "*Causes of Porosity in Tin-Bronze Castings*," JOURNAL, Institute of Metals, vol. 67, p. 231 (1941).
4. Scheuer, E., METALLWIRTSCHAFT, vol. 14, p. 337 (1935).
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DISCUSSION

Chairman: G. P. HALLIWELL, H. Kramer & Co., Chicago.

Co-Chairman: A. H. HESSE: R. Lavin & Sons, Inc., Chicago.

G. M. THRASHER[1]: There is a question in my mind about the type of riser the author has on the large block in Fig. 2. It seems inadequate for the size of the block. In a casting of that size, knowing the feeding conditions necessary for it and lack of chilling in a thin section, it seems to me that the riser should be heavier or higher to take care of the microshrinkage in the center of the casting.

MR. BROUK: That is perhaps true. In planning the program for this investigation, several techniques of risering and gating were proposed. However, each block was radiographed and only those blocks that were radiographically sound were used for subsequent tests. All unsound castings were rejected.

MR. THRASHER: Would you show the actual composition of the different bars as cast?

MR. BROUK: The composition of the various bars is shown in Table 1. It is apparent that inverse segregation has occurred within the block. The lower melting constituents composed of the eutectoid and free lead have segregated, leaving deficiencies of the constituents in the center of the block. This would account for the lower mechanical properties in the center; however, in red brass, the low properties at the center cannot be attributed to the microsegregations but to porosity.

MR. THRASHER: In most cases, it is evident that the copper is higher and the other elements somewhat lower at the center of the bar (Coupon "B").

MR. BROUK: That is the effect of inverse segregation.

MR. THRASHER: I have found that to be the case in separately cast test bars. For instance, there is one case where the chemist attempted to take the outside surface of the test bar after cleaning up, (taking the last shavings), as sample for analysis. The copper appeared too low for the specification, and higher in tin, and when he cut the cross section of the same bar, it came within the specifications.

MR. BROUK: I am assuming the section of the coupon was ⅝ in. round. There is often inverse segregation taking place in the bar itself, so a better sample for chemical analysis would be the bar itself.

MR. THRASHER: You mean the cross-section of the bar.

G. E. DALBEY[2]: In regard to the inverse segregation, is it not always associated with gas in the metal?

MR. BROUK: I would hesitate to conclude that segregation is always associated with gas. Tests have been made of chilled casting of both thin and thick sections, and investigated for porosity. Many were quite sound, yet by chemical analysis, inverse segregation was evident. In other words, the tin was higher at the surface of the bars.

I believe the presence of gas augments inverse segregation. It must, because the pressure of the gas coming out of solution forces the low melting alloys that are still in a liquid state to the surface of the casting. This takes place within the solidification range.

MR. THRASHER: I have seen cases where 85-5-5-5, running, as it sometimes does, as low as 4.30 per cent tin, with approximately the 5 per cent or a little higher in lead and normally 5 per cent zinc, when cast in some small castings, as proved when fractured, shows a picture frame structure and tears at the center, showing a coppery color. While if given enough zinc to balance off the low tin, it comes out almost a uniform fracture with gray all the way through and without inverse segregation.

MR. BROUK: We have experienced the same observation.

MR. THRASHER: The vapor pressure of the zinc seems to drive out the gas which may be absorbed there. The vapor pressure of the zinc is pretty high. Especially when casting at a temperature of 2200° F., the vapor pressure is approximately 100 psi.

CHAIRMAN HALLIWELL: This question of gas porosity is one that is very interesting and baffling. As better methods of melting and founding are developed, improvement in mechanical properties generally result. However, fractures which appear sound may show considerable pin hole porosity by means of the microradiograph.

If gas is assumed to be a cause of inverse segregation, an apparently dense fracture would be no criterion for the lack of inverse segregation, since the presence of microporosity as shown by the microradiograph indicates that some of the factors which may influence inverse segregation are still present.

MR. BROUK: That is true. The unsound, rejected castings were examined by fracture and in each case, the broken surface was multicolored. However, the sound castings showed a fibrous fracture.

MR. THRASHER: In the last instance I mentioned with samples having a coppery color center, I made the density determination myself. The one which I treated with zinc in the next heat with the same original composition, I made density determinations to the fourth decimal place and found in the one instance the density was only approximately 8.6, whereas it was 8.8 in the treated bar, which had about 7 per cent zinc instead of 5 per cent.

E. PORTMAN[3]: This paper points out quite conclusively that the nickel bronzes are better suited for pressure casting than gun metal, and I have found that to be true in actual casting practice. I did notice in Table 1 that the deoxidation practice for nickel-bronze requires about three times as much phosphor-copper as does gun metal and that, I believe, is the secret of good results.

MR. BROUK: Caution should be taken because nickel-tin bronzes tend to rim as seriously as the Monels, so care should be practiced in degassifying with phosphor-copper, as we have done, or in some cases with magnesium. If the operator becomes experienced in this technique, he can produce a pressure tight casting with these nickel-tin bronzes.

E. J. BOYLE[4]: Would you not expect these large castings to have inferior properties as compared to the separately cast coupons purely on a grain size basis, that is, much coarser grain size in the heavy castings because of the slower solidification?

MR. BROUK: I do not believe so. We have had gun metal with tensile strengths as high as 48,000 psi. and 75 per cent elongation at the surface of the casting where the grain size was very large and columnar. The separately cast bar had a 44,000 psi. tensile strength with columnar grains but smaller in size. The center of the block was but 37,000 psi. The only conclusion made was that the distribution of mechanical properties is more uniform in castings of equi-axed grains than in columnar grains. There is much to be done to evaluate properly the effect of grain size on the properties in sand castings.

MR. BOYLE: I think you could get inverse segregation without gassing the metal, merely because of the contraction of the metal, the contraction forcing the lower melting constituents to the surface.

MR. BROUK: Yes, that is microshrinkage. When the lower melting constituents start to solidify at the surface of the casting, the only available feed metal is that metal in the center of the casting. These small reservoirs are tapped by the contraction demand of the solidifying alloy. It thus leaves areas void of metal, particularly when the low melting constituent is entrapped by massive dendrites which obstruct the flow of feed metal. That is microshrinkage without gas porosity.

MEMBER: During the last war, we poured two coupons in the same mold through one gate, using Navy "G" Metal. What causes one to pull 3,000 or 4,000 psi. more in tensile strength than the other bar?

MR. BROUK: I wish I could answer that. We have experienced the same thing, but I do not think 3,000 or 4,000 psi. difference in tensile strength is serious. It is rather difficult to explain differences greater than 5,000 psi. There are just too many unaccounted factors that are rather difficult to isolate when casting test bars.

MEMBER: We just took the average.

MR. BROUK: There could be many causes in a test coupon, such as inclusions, cold shuts, mis-runs, moisture, sand properties and a host of others.

MEMBER: We poured these all in the core. We poured both of them together, using the same gate. The only dividing line was the levelness of the core.

MR. BROUK: I do hope that the test bar design is changed so we can have at least an account for many of these discrepancies.

CHAIRMAN HALLIWELL: I think if you have only 3,000 psi. difference in tensile strength between bars you are fortunate. I wish we could all do that. In the work reported upon by Dr. Clamer in the Fourth Annual Foundation Lecture the inconsistent tensiles was one of our main problems and we have not yet been able to determine why one casting out of the same mold, poured right alongside another is different. There are some complex relationships there and we have not begun to isolate them.

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[1] R. Lavin & Sons, Elmira, N. Y.
[2] U. S. Navy, Mare Island Yard, Vallejo, Calif.
[3] Nassau Smelting and Refining Co., Inc., New York.
[4] The New Jersey Zinc Co., Palmerton, Pa.

EFFECT OF MOLD MATERIALS ON MAGNESIUM CASTINGS QUALITY

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and
J. G. Mezoff
The Dow Chemical Co.
Midland, Mich.

IN RECENT YEARS aluminum and magnesium have been extensively used in aircraft components where soundness and quality have been of paramount importance. To expedite the production of sound castings chills were liberally employed by many foundries, although it has been conclusively demonstrated that soundness can be attained by other means.

Many types of chill materials are used by different magnesium foundries. Gray cast iron is the most commonly used chill material because of its low cost, ease of casting into intricate shapes to conform to the contour of the casting, durability, satisfactory cooling capacity, ie., thermal conductivity and specific heat, and ready separation from the molding and core sand by magnetic means.

Frequently, a more drastic chilling medium than gray cast iron is desired, and copper is used to meet this requirement. When greater toughness and durability are required aluminum bronze or steel may be employed. To meet certain conditions, various other materials such as graphite, magnesium "M" alloy, etc., have been tried.

Sand cores generally are considered to have the same relative effect on the metal in the casting, with respect to directional solidification and cooling rate, as the surrounding green sand mold. Therefore, chills are used in cores as in green sand to promote solidification.

Inconvenience and troubles connected with placing chills in cores have led to the development and use of core sand mixtures[3, 4, 5] having increased chilling capacity. These mixtures have been used to form either the whole or parts of a core. The use of these special core sand mixtures involves considerable additional expense because of the increased cost of the raw materials and extra labor in handling.

Fundamental Data. The purpose of this work is to evaluate the following items and thereby establish fundamental data on which future research can be advantageously based:

1. Quantitative evaluation as to chilling characteristics of various mold and chill materials.

2. Effect of cooling rates and thermal gradients upon the soundness and mechanical properties of magnesium castings.

Review of the Literature. In view of the antiquity of the art of casting and of the magnitude of the foundry industry today, it is disappointing to find so little information in the literature dealing with the effect of the various mold materials on cooling rates, thermal gradients, and mechanical properties of the resultant castings.

Certain mold variables such as moisture content, permeability, bonding material, washes, strength, collapsibility, etc., have been studied extensively. On the other hand, the evaluation of the basic raw materials used for molds and chills in reference to their capacity to influence the way in which the metal cools and solidifies has been neglected.

A review of the available literature indicates that molding materials for casting gray iron and steel have been investigated briefly for their effect on solidification rates and thermal gradients.

Womochel and Sigerfoos[6] have concluded that (1) solidification rates of gray iron castings in green sand molds and in dry sand molds are approximately the same, (2) a variation of moisture content from 4.9 to 7.0 per cent had but little effect upon the cooling rate of the casting, and (3) castings made in bentonite-bonded molds cooled somewhat more rapidly than

Production of sound sand castings has always been the goal of the foundry industry. Recent foundry investigations[1, 2] have clearly demonstrated that soundness in castings is a function of the proper application of controlled directional solidification. Factors lending themselves to the control of directional solidification are numerous. Among those factors which can be readily controlled and which have found prominent application are pouring temperature, pouring rate, riser size and location, gating method, means for keeping the risers hot, and chills.

Presented at an Aluminum and Magnesium Session of the Fiftieth Annual Meeting, American Foundrymen's Association, at Cleveland, May 6, 1946.

the same castings made in non-bentonite bonded molds.

Briggs and Gezelius[7] in their work on steel castings have confirmed, in part, the foregoing conclusions by stating, "The data tend to show that the rate of solidification is practically the same in either of the common types of green or dry sand. Variations in the ordinarily used molding materials result in little more than a perceptible change in the velocity of solidification."

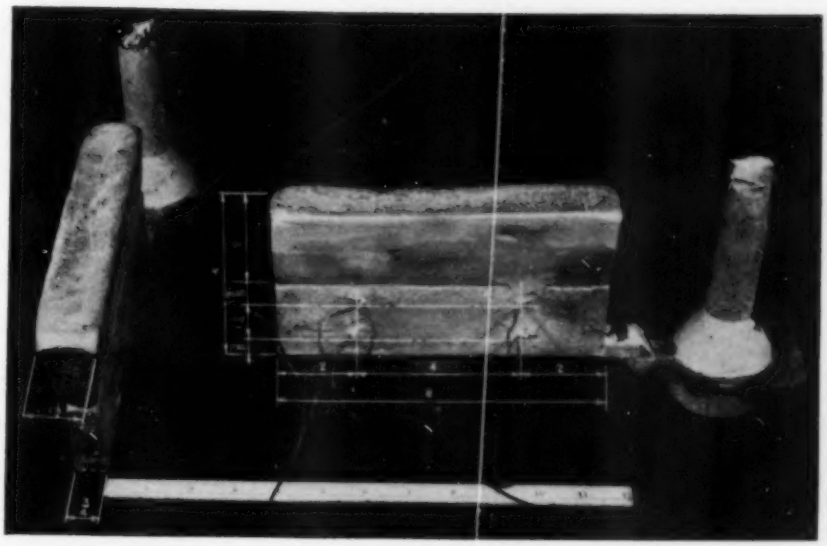
Baker[8] in his extensive work on porosity and unsoundness in magnesium castings has ascertained that, "Of the many factors which govern the solidification process in the casting, only a limited number can be conveniently varied in practice. From the point of view of freedom from porosity the overriding consideration is to produce and maintain steep temperature gradients from the feeders to the remote parts of the castings, and the most important factors which can be conveniently varied to achieve this object are (*a*) the method of pouring, (*b*) the pouring speed, and (*c*) the pouring temperature and mold temperature."

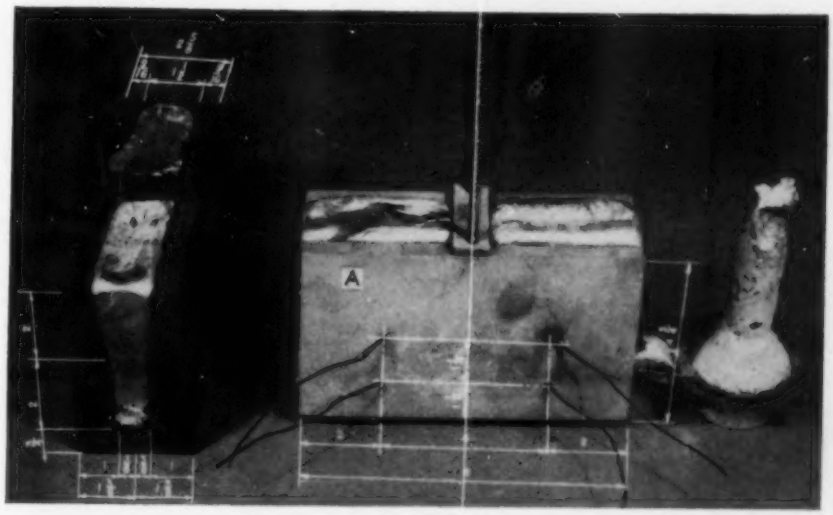
Experimental Procedure

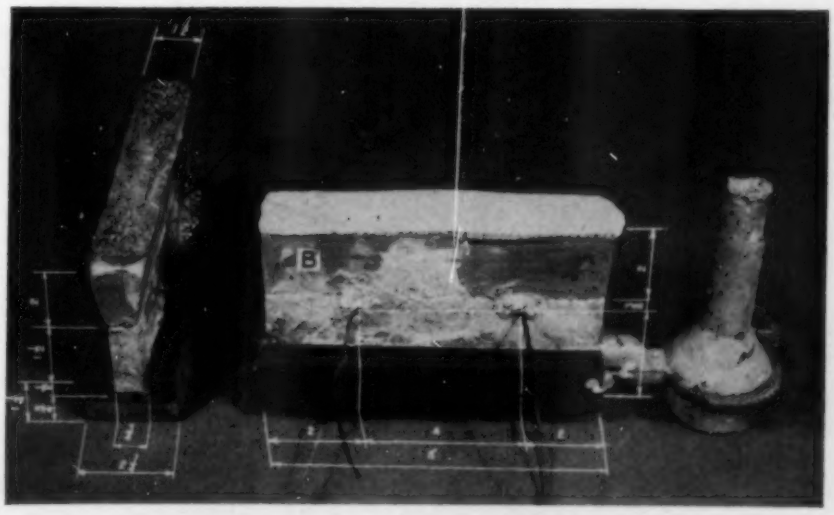
As previously stated, the primary purpose of this study was to demonstrate the influence of mold and core materials on two factors of importance in the casting of magnesium alloys, namely:

1. Relative effect of the various materials on cooling rates and mechanical properties of the resultant castings.

2. Magnitude of the thermal gradients established within the castings by means of combinations of the various materials, and the effect of thermal gradients upon soundness and mechanical properties.

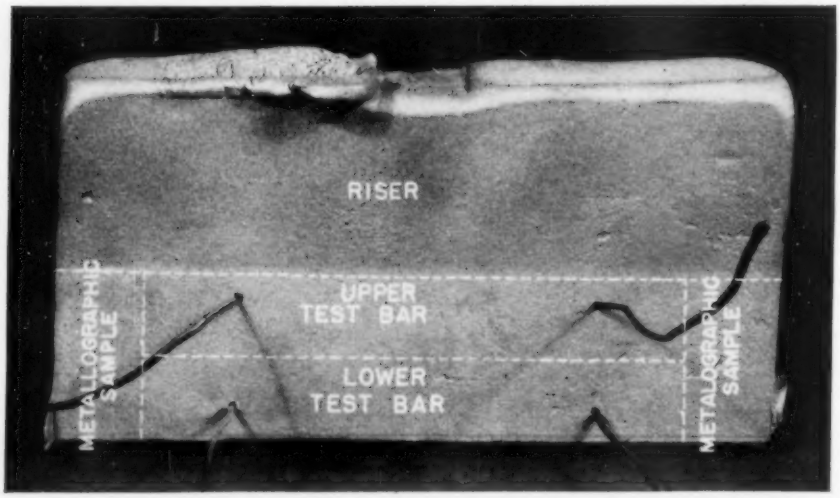


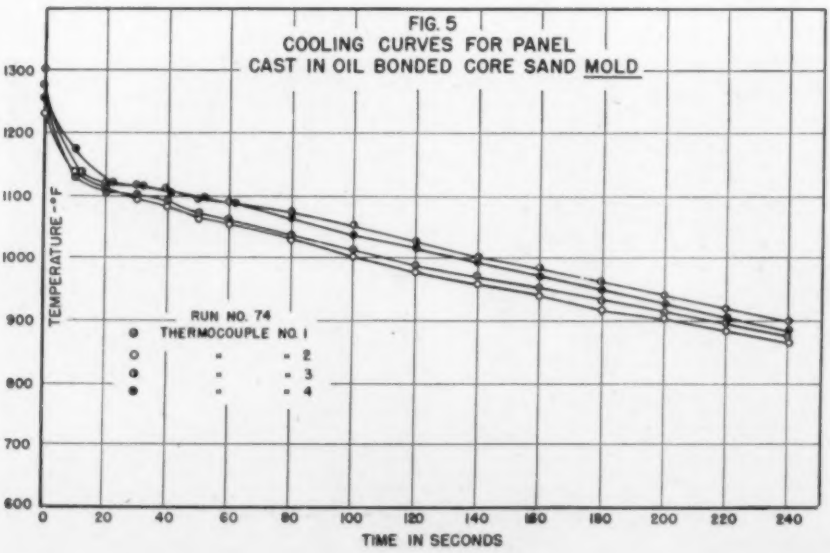




Reading (right) top to bottom:
Fig. 1—Panel used for study of effect of mold materials upon cooling rates, thermal gradients and properties.
Fig. 2—Locations of thermocouples and dimensions of (A) molds and (B) chills used for cooling-rate studies of mold materials.







Casting Type Selected. A panel casting ¾-in. thick, 2-in. high and 8-inch long was arbitrarily chosen for these tests. A riser 2-in. high, 8-in. long and tapering from a ¾-in. width at the base to 1½-in. at the top was selected so as to produce a radiographically unsound casting when made in green sand.

A radiographically unsound casting was desirable in that a better qualitative evaluation of the effect of cooling rates and thermal gradients upon the soundness and mechanical properties could be obtained than with a casting which was basically sound. Figure 1 is a photograph of a panel casting with gating, risering, dimensions, and thermocouple locations shown.

Description of Chills and Molds. For studying the relative cooling rates of various mold materials, a mold such as is shown in Fig. 2A was used. In this mold, 80 per cent of the surface area of the test panel and riser are in contact with mold material which is varied from run to run. The remaining 20 per cent of the panel surface is in contact with molding sand ends (8 per cent) or atmosphere (12 per cent).

In studying thermal gradients and their effect upon mechanical properties, an entirely different problem is encountered and a different sort of mold must be used. In this case, a chill (Fig. 2B) whose composition is varied from run to run is placed along the bottom of the mold.

This chill contacts 15 per cent of the surface area of the panel and riser, and provides maximum directional solidification and thermal gradients. The size of this chill is larger than normal foundry practice would employ; thus a range of thermal gradients much broader than found in usual founding operations was studied.

As employed in this paper, the

Reading (left) top to bottom: Fig. 3—Cutaway mold showing position in mold of cast panel and chill.

Fig. 4—Location of test bars and original metallographic samples taken from panels cast in study of mold and chill materials.

terms "mold" and "chill" do not necessarily conform to customary foundry nomenclature. The term "mold," as a matter of convenience and distinction, is applied to the shapes as illustrated in Fig. 2A, or to green sand molds, which are employed in the determination of cooling rates, irrespective of the materials used. On the other hand, the term "chill" is applied to shapes as illustrated in Fig. 2B, which were used essentially for determination of thermal gradients. In some cases these "chills" do not actually act as chills.

Description of Chill and Mold Materials. A few brief comments on the source and some features of the mold and chill materials tested are in order. Composition and properties, with accompanying screen analyses, are shown in Tables 1 and 2.

Gray Iron. The mold was machined from cast 5-in. diameter round stock. It contained coarse graphite, was quite open, and had a density of 6.53. The chill, on the other hand, was machined from a cast 3-in. diameter round of much better quality (density 7.43).

Using "M" Alloy

Magnesium "M" Alloy (1.5 per cent Mn, remainder Mg). Of several magnesium alloys which could be used for mold and chill materials, "M" alloy was chosen for these tests because of its good corrosion resistance and high thermal conductivity, and because it is already in occasional use in production foundries. Both mold and chill were machined from cast "M" alloy blocks of good quality.

Graphite. The mold and chill were machined from 6-in. and 3-in. diameter electrode stock, respectively. The 6-in. diameter stock had more porosity than the 3-in.

Carbon-Bonded Graphite. A carbon-graphite product having a transverse strength of 10,000 psi., 75 scleroscope hardness, thermal conductivity of 12.95×10^{-3} cal/cm/cm²/°C/sec., and 8 per cent porosity was used as a chill.

Graphite Core. It has been conjectured that graphite additions to core sand mixtures would provide a highly conductive film over the sand grains; and a molding mixture of graphite powder would be still more conductive. Therefore, various mixtures of graphite powder and binders were tried as mold and chill materials.

Of the various mixtures tried, the only one which could be used for this purpose consisted of 100 parts graphite, 10 parts urea-formaldehyde binder, and 10 parts water. This formed a spongy mass which has considerable tendency to spring back on ramming. The molds and chills used had a poor surface and were too weak to consider for use in a production foundry.

Core Sands. Three typical core sands were used in these tests. The urea formaldehyde bonded core mixture conforms to the practices employed in many magnesium foundries. The other two, steel shot and zircon mixtures, have been used under the assumption that they had chill characteristics.

Table 1

COMPOSITION AND PROPERTIES OF BAKED CORES

	Core Mixtures (parts based on weight of dry sand)					
MATERIALS*	Urea Formaldehyde Bonded Core	Steel Shot	Zircon	Urea Formaldehyde Bonded Graphite	Molding Sand	Plaster
Gratiot Bank Sand (60-65 Fineness No.)	100	30			—	—
Sulphur (325 mesh)	0.78	1	0.78	0.75	—	—
Boric Acid (powder)	0.50	0.7	0.50	0.50	—	—
Cereal Binder	0.625		0.47	0.625	—	—
Core Oil	—	1.8	0.6	—	—	—
Water	4%	1.2%	0.75%	10%	4%	100
Urea Formaldehyde	0.625			10	—	—
Vassar Bank Sand (90-95 Fineness No.)	—			—	92	—
Bentonite	—	0.75		—	4	—
Inhibiting Agent	—			—	4	—
Graphite	—			100	—	—
Chill Shot	—	70		—	—	—
Zircon	—		100	—	—	—
Stucco	—					80
Talc						20
PROPERTIES						
Green Permeability	127	52	28	21	80	0
Green Compression, psi.	1.5	4.8	1.35	9.8	7.0	—
Deformation, in./in.	—	—	—	0.050	0.020	—
Dry Permeability	152	65	40	52	—	7
Dry Compression, psi.	830	—	—	280	160	75
Tensile Strength, psi.	178.5	225	310	5.8	—	—
Hardness	86	95	93	85	—	—

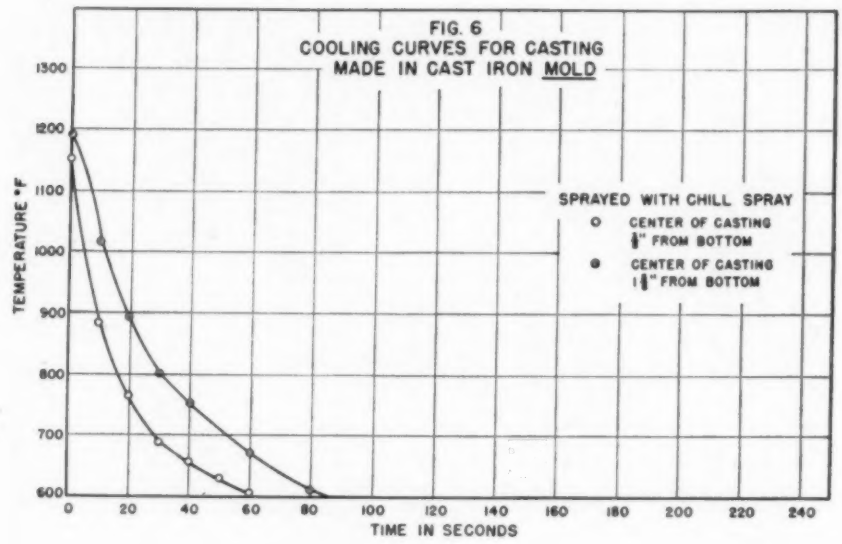
SCREEN ANALYSIS

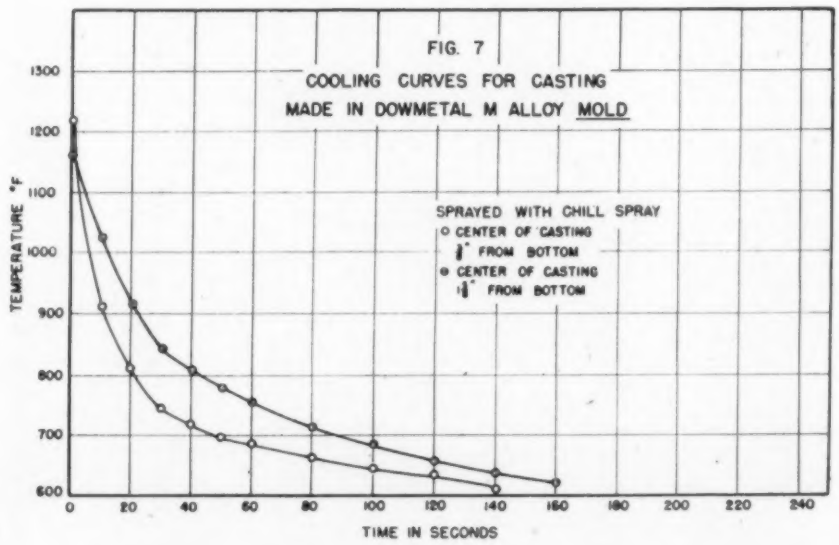
	Retained on Screen, per cent—					
	Bank Sand		Steel Shot			
Sieve No.	Gratiot	Vassar	28X	No. 40	Zircon	Graphite
12	0.0	0.0			—	35.0
20	0.2	0.0			—	31.2
30	0.4	0.0	4.02	2.63	—	20.2
40	1.8	0.2	93.26	95.28	0.06	7.2
50	6.9	0.3	2.72	2.04	0.02	1.4
70	22.7	5.0		0.05	0.22	1.2
100	62.9	45.0			17.03	1.0
140	4.7	35.2			67.61	1.0
200	0.2	12.6			14.92	0.4
270	0.0	0.6			0.12	0.4
Through 270	0.2	1.1			0.02	1.0
Fineness No.	64	91.2	29.8	28.9	100.8	
Distribution No.	50	83.2				

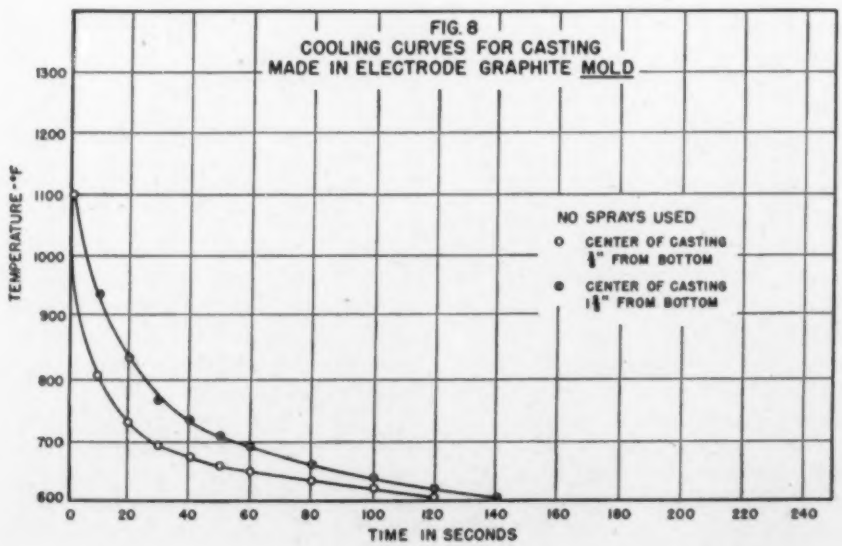
*Trade names of products used may be obtained from the authors.

Protective Spray

Both the steel shot and zircon cores, employing an oil binder, were given a protective spray before use because of loss of agent during baking at temperatures of 425-475° F. Urea formaldehyde-bonded cores, on the other hand, were not sprayed, since they were baked at lower temperatures of 300-350° F. with consequent greater retention







of protective agent. Table 3 shows the composition of core sprays.

Molding Sand. The molding sand employed in these tests is a fluoride inhibited type commonly used in magnesium foundries. Its composition and properties are shown in Table 1. Green sand molds were neither skin dried nor sprayed.

Plaster. The plaster molds and chills were calcined for periods of 2 hr. at 250° F. The temperature was then increased for a 6-hr. period to 800° F., followed by cooling slowly to 200° F., and maintained at this temperature until used. The plaster surface was given a protective spray just before insertion into the green sand mold.

Casting Procedure. A brief description of the procedure employed for each test follows. The metal molds or chills were sandblasted, sprayed with a wash, and thermocouples anchored in place. The assembly was heated before insertion into a sand mold which had, in the meantime, been rammed to a uniform hardness. The heating was so adjusted that sufficient residual heat remained in the mold or chill for it to be at a temperature of approximately 80° F. at the time of pouring.

Cores

Molds and chills made of either graphite or sand cores were treated in a similar manner except, of course, that they were not sandblasted.

Following insertion of the mold or chill, the green sand mold was closed, and the projecting thermocouples connected to conveniently arranged portable potentiometers.

Simultaneously, 10 lb. of magnesium "H" alloy (6 per cent Al, 3 per cent Zn, 0.15 per cent Mn, remainder Mg) were melted in a steel crucible under No. 310 flux, superheated for 15 min. at 1650-1700° F., cooled to 1450° F., and poured. The metal temperature was determined by means of a 20-gauge chromel-alumel thermocouple enclosed in a ¼-in. seamless steel protection tube and connected to a temperature indicator.

Metal was poured rapidly into the mold, filling it just to the top of the riser. At the moment the mold was filled, the interval timer was started and the four metal temperature readings obtained sim-

ultaneously at 10-sec. intervals for the first 5 min., then at one-min. intervals for the next 5 min., then at 5- and 10-min. intervals until the casting temperature had dropped below 300° F.

Data for the cooling curves of the panel were obtained manually by means of portable potentiometers connected to four 24-guage glass-insulated iron-constantan thermocouples located at definite positions within the casting, as shown in Figs. 1 and 2. The thermocouples were held in position either by anchoring in holes drilled through the mold or chill or by embedding in the molding sand. The thermocouple positions chosen are shown in Table 4.

All thermocouple hot junctions were located at equal distances from the sides of the casting. Location of the thermocouples at 2-in. from each end of the casting was for the purpose of mitigating any end effect from dissimilar materials. Figure 3 is a cutaway view of a mold showing the position of the panel and chill, and general arrangement of equipment for these tests.

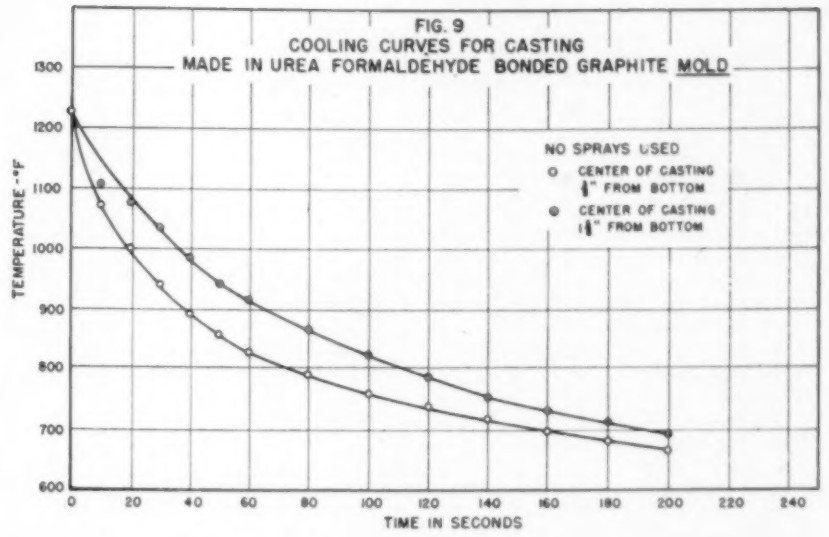
Casting Design

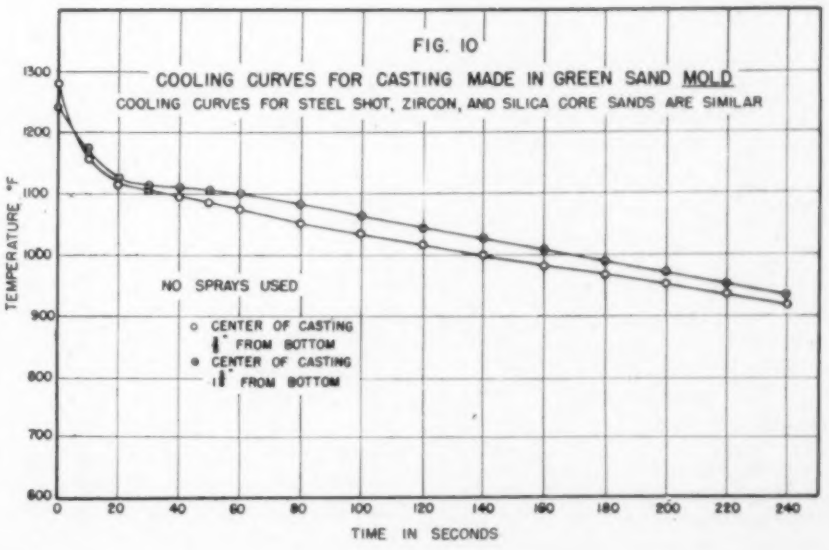
A few additional tests were conducted in green sand molds in which the width of the panel and riser size were varied as shown in Table 5. This was done to learn the influence of casting design upon the variables under study.

Duplicate runs were made for some conditions in which chlorinated[9] metal was used to determine its effect upon the quality of the casting. Chlorination is used in many foundries for degassing and refining the metal. A chlorination treatment was employed in which about 2 per cent chlorine was passed through the metal at 1350 to 1400° F.

Panel Testing. After the cooling curves were obtained each panel was radiographed, then given a heat treatment in an atmosphere of 1 per cent SO_2. The heating cycle consisted of 2 hr. at temperatures from 500 to 730° F., plus 12 hr. at 730° F., followed by cooling to room temperature in air. These heat-treated panels were then sectioned, as shown in Fig. 4, for metallographic and tensile samples.

Radiographs were arbitrarily grouped into five uniform classes





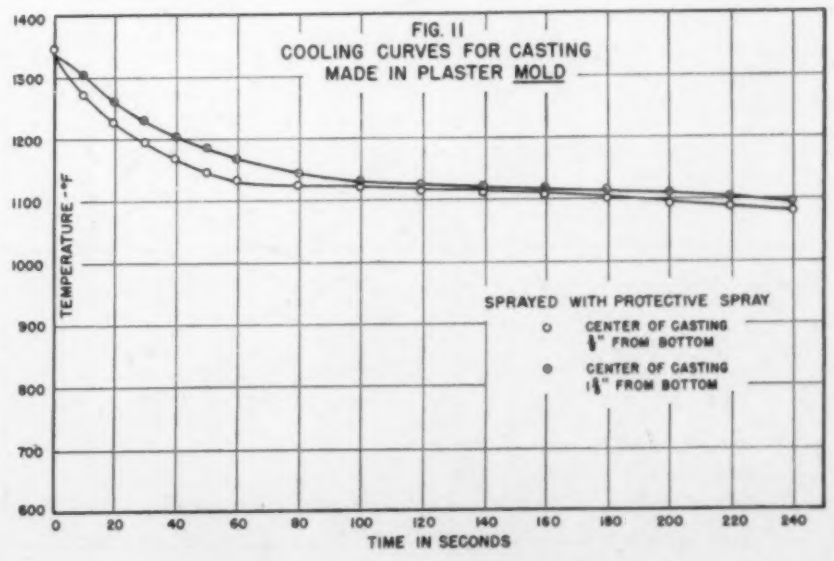


Table 2

PHYSICAL PROPERTIES OF MOLD MATERIALS

Material	Density* gm/cm³	Density* lb/in³	Thermal Conductivity, cal/cm/cm²/°C/sec	Specific Heat	Heat Capacity cal/cm³/°C
Cast Iron					
—Mold	6.53	0.236	0.110(12)†	0.132(12)†	0.863
—Chill	7.43	0.268	0.110	0.132	0.980
Magnesium "M" Alloy					
—Mold	1.76	0.064	0.310(13)	0.249(13)	0.438
—Chill	1.75	0.063	0.310	0.249	0.436
Electrode Graphite					
—Mold	1.55	0.056	0.347(14)	0.195(14)	0.302
—Chill	1.635	0.059	0.347	0.195	0.319
Carbon-Bonded Graphite					
—Chill	1.70	0.0615	12.95×10^{-3}(15)	0.251(15)	0.427
Urea Formaldehyde-Bonded Graphite Cores	—	—	—	—	—
Green Molding Sand	1.76	0.0636	—	—	—
Core Sand					
—Steel Shot	3.22	0.116	10.7×10^{-4}(16)	0.216(16)	0.695
—Zircon	2.27	0.082	7.91×10^{-4}(16)	0.186(16)	0.422
Urea Formaldehyde-Bonded Core	1.60	0.058	10.0×10^{-4}	0.254	0.406
Plaster					
—Mold	0.62	0.0224	—	—	—
—Chill	0.64	0.0231	—	—	—

*Density of solid materials determined by immersion; cores by calculation from weight and volume.

†Figures in parentheses refer to bibliography at end of paper.

Table 3

SPRAY COMPOSITIONS

Chill Wash	
Denatured Alcohol, gal.	4.5
Mazein, qt.	0.75
Talc (white), lb.	5⅝
Protective Spray	
Methanol, gal.	4
Boric Acid, lb.	3¼
NH_4BF_4 (No. 100 inhibiting agent), lb.	1⅝
Talc (white), lb.	5⅝

(Fig. 23), ranging from radiographically sound panels with a rating of one to those with a rating of 5 which exhibited the maximum degree of porosity and unsoundness observed in this study.

Heat treated metallographic samples were examined for average grain diameter, massive compound rating, and porosity. It was observed that a marked gradation of porosity from sprue to far end existed in slowly cooled panels. Therefore, porosity ratings were also determined on the fractured ends of the tensile bars, and it is these values that are presented in Table 6.

Tensile samples were machined to 0.5-in. diameter bars, polished, and tested in tension for elongation, yield strength (0.2 per cent offset), and ultimate strength. Sections from ⅜-in. panels were machined to flat tension bars, polished, and likewise tested in tension. Table 6 presents a summary of the data.

Graphs and Calculation of Data. The data obtained are too voluminous to present as tables and, therefore, are shown entirely in graphical form.

Cooling Curves

Figure 5 shows the cooling curves for four locations, as previously discussed, in a panel cast in a core sand mold. It is observed that the curves for locations 3 and 4 (upper) follow parallel paths and nearly coincide. The same is true for locations 1 and 2.

This concurrence is even more apparent in the panels cooled more rapidly, whether cast in molds or chills. Therefore, the remaining graphs are plotted from the average values of the upper two thermocouple locations (3 and 4) and also from the average of the lower two thermocouple locations (1 and 2).

Cooling Curves. Figures 6 to 12 are time-temperature graphs of panels cast in the various mold and chill materials in which the data for the curves represent the average of at least two, and usually three or more tests. Although readings were taken until the panel temperature had dropped below 300° F., the graphs below 600° F. were practically straight-line functions which merged. Therefore, the graphs are stopped off at 600° F., which is below the eutectic temperature, in order to expand the scale for the promotion of greater clarity.

Cooling Rate Curves. The cooling rates (Figs. 13-21) of the panels cast in the various molds and chills were obtained from the cooling curve data by plotting the average temperature drop (° F./sec.) obtained from two consecutive readings against the average temperature in this interval. These rate

Table 4

THERMOCOUPLE POSITIONS

Thermocouple No.	Position in Casting: From Bottom, in.	From Sprue End, in.
1	⅜	2
2	⅜	6
3	1⅝	2
4	1⅝	6

Table 5

PANEL WIDTH AND RISER SIZE

	Panel Width, in.	Riser, in. Height	Riser, in. Width	Riser, in. Length	Volume Ratio Riser/Panel
Original Panel	¾	2	¾x1½	8	1.5
Thin Panel	⅜	2	⅜x¾	8	1.5
Large Riser	¾	3	¾x2½	8	3.83

curves serve to accentuate any temperature retardation that may not be apparent in the cooling curves.

Cooling Rate at 1000° F. These values, the average of the lower thermocouple readings 1 and 2, are taken directly from the cooling-rate curves.

Thermal Gradients. These values, in °F. per in., are obtained from the cooling curves by obtaining the average temperature differential between the upper and lower thermocouple locations over the temperature range of 1100-700° F. for the lower two thermocouples, and dividing by 5/4 (distance between upper and lower thermocouples is 1¼ in.)

Those thermal gradients established between the surfaces and the center of the panel, although appreciable, were not measured or given consideration in the calculations. Figure 22 shows representative thermal gradients established in panels under four different conditions of cooling. For purposes of comparison, Figs. 24 and 25 show respectively the effect of thermal gradient established in the temperature range of 1100-700° F., and of cooling rate at 1000° F., on the ultimate tensile strength of panels.

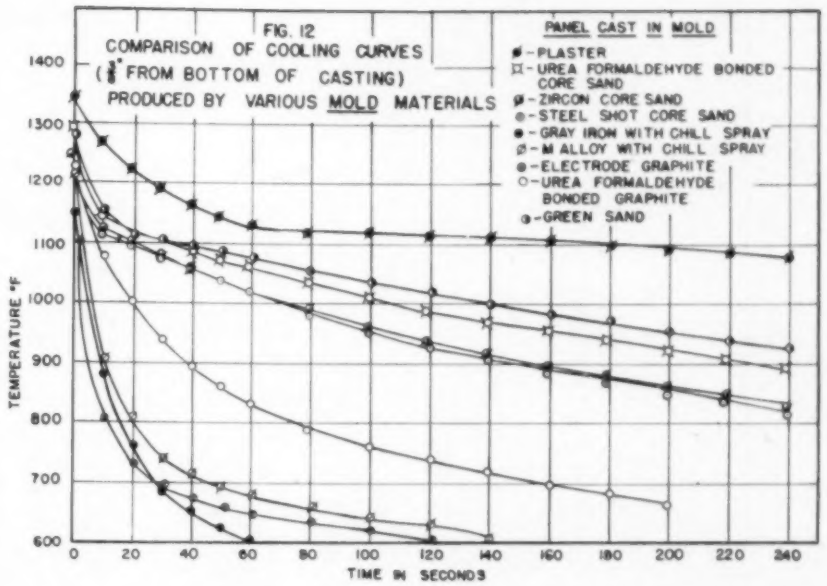
Average Time to Reach 700° F. These values were obtained from the data for the lower thermocouples.

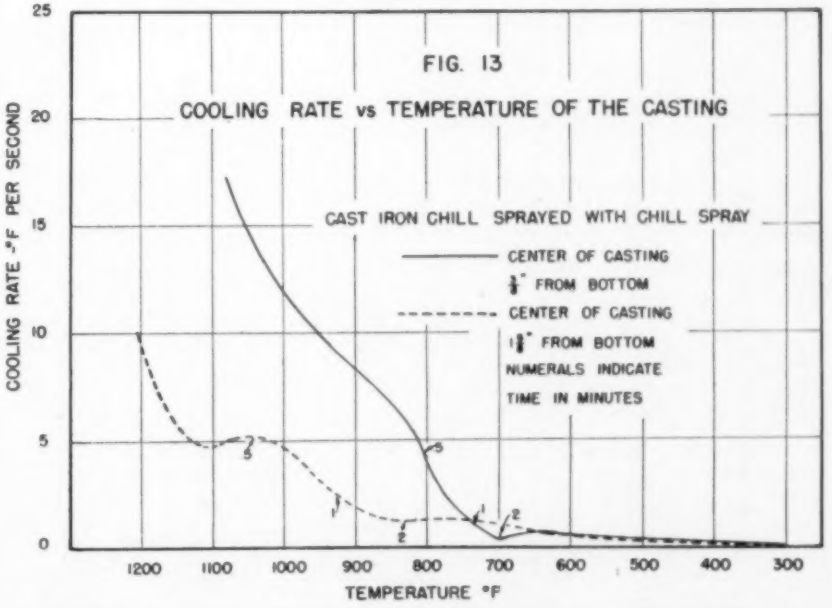
Experimental Discussion

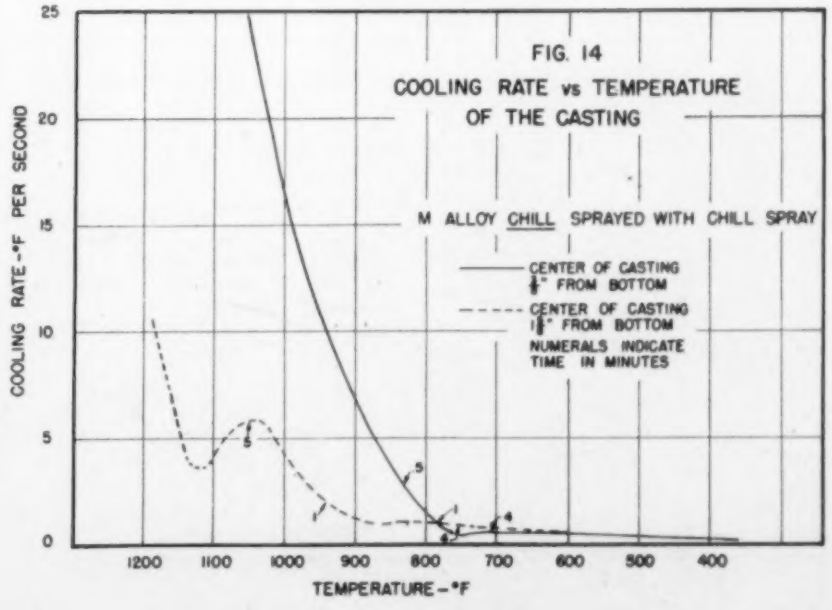
One of the important characteristics of a mold material for casting metals is its "chilling capacity" or the rate at which it is able to extract heat from the metal in the casting cavity, for most of the heat contained in the liquid casting must be removed through the mold itself.

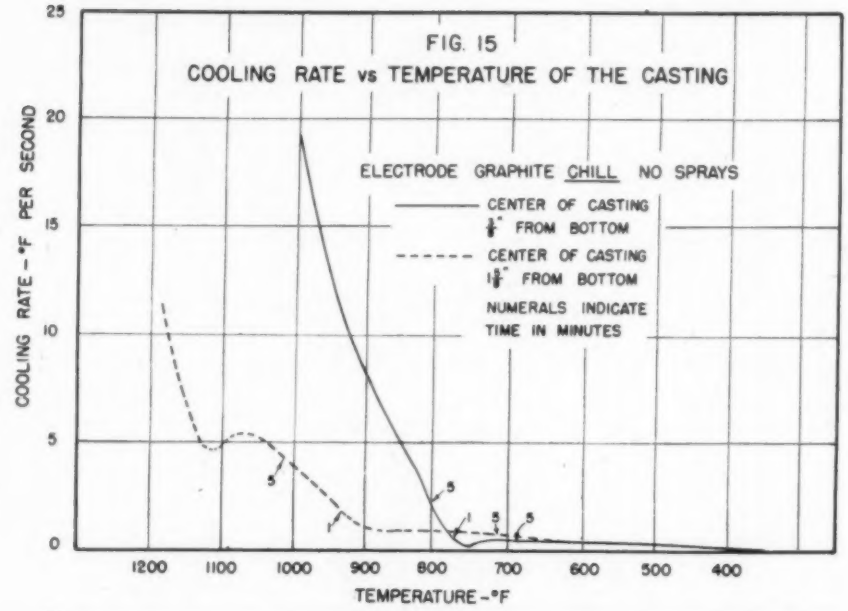
Therefore, the chilling capacity of the mold material determines to a considerable extent the cooling rate of the casting. Of the factors influencing the solidification of castings, the judicious selection and use of mold materials is perhaps one of the most important.

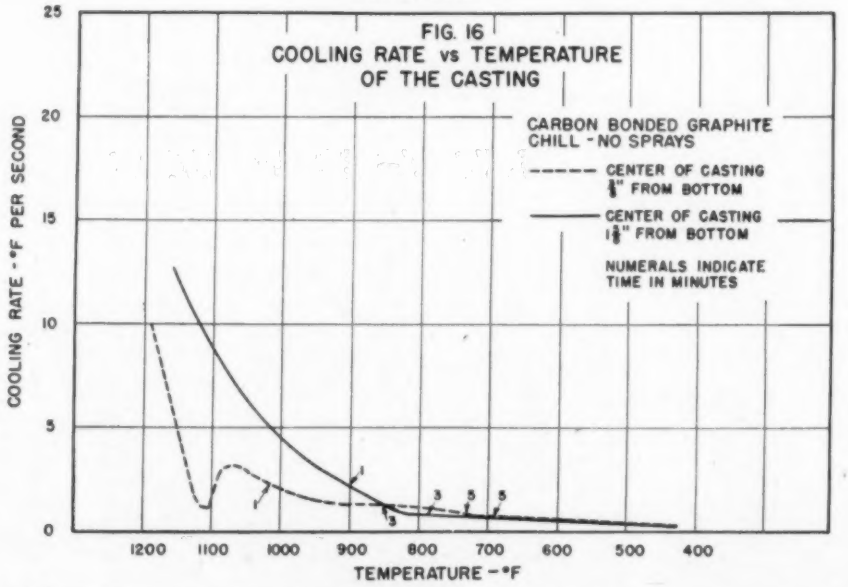
Chilling capacity of mold materials to be used in casting magnesium-base alloys is of particular interest in the temperature range 1100-700° F., which is the approximate solidification range of the commercial Mg-Al-Zn type ternary

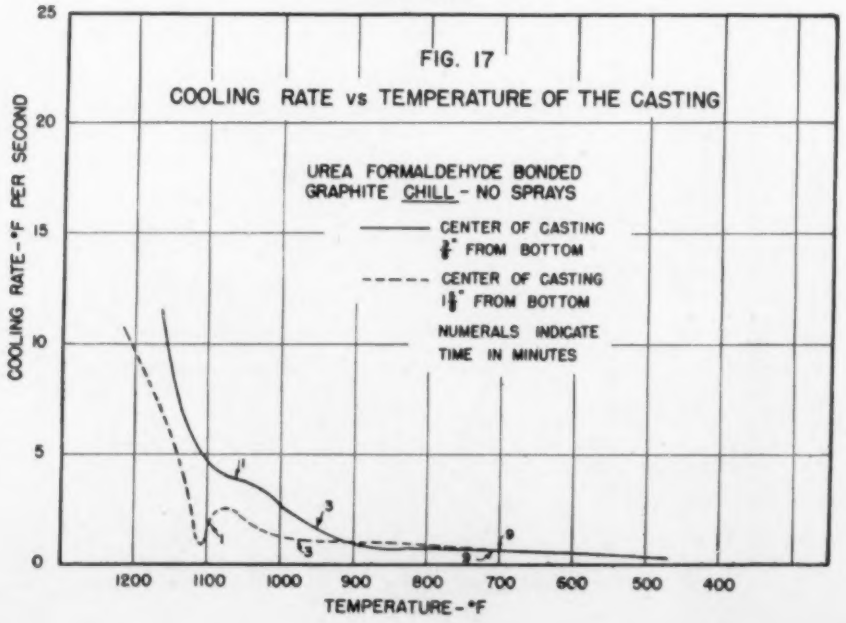












alloys used in this country. Perhaps some small part of that range is most critical during solidification and feeding.

At present, it must be assumed that the whole range is important. The chilling capacity ratings of the mold materials discussed in this paper have been based upon their effectiveness throughout the solidification range of the ternary commercial magnesium-base alloys.

Chilling Capacity of Mold Materials. A comparison of the chilling capacity of the various mold and chill materials on the inadequately risered ¾-in. panel is provided by Table 7. Method *A* is based upon the average cooling rates of panels produced by the molds in °F./sec. during the interval the lower thermocouples were at 1000° F. Method *B* is based upon a comparison of the relative length of time for the panels cast in molds to cool to a temperature of 700° F. as measured at the lower thermocouples, the basis of comparison being green molding sand.

Chilling Capacities

Method *C*, on the other hand, is based upon the magnitude of thermal gradients produced by the various chill materials as the panels cooled from 1100 to 700° F., assuming green molding sand to be unity. The close agreement among these various arbitrary methods of evaluating chilling capacity indicates that these mold materials have chilling capacities in the order listed.

In any discussion of the factors influencing the chilling capacity of a material, at least three properties should be considered: (1) thermal conductivity, (2) specific heat, and (3) density. The product of the latter two physical properties forms a third value which, for the purpose of convenience in the remainder of this report, will be called "heat capacity."

An examination of the data summarized in Table 6 reveals that all materials with high chilling capacity have also high thermal conductivity, while the heat capacity, specific heat, and density vary over a relatively wide range. Therefore, if a material is to have a high chilling capacity it must have primarily a high thermal conductivity, and secondarily, a high heat capacity.

Foundry cores generally have a high heat capacity, but their thermal conductivity is low. The thermal conductivity of a core is a function of both the properties of the basic core material and its manner of agglomeration. Thus, steel shot, for example, has a high thermal conductivity in itself, but the method of joining the shot and sand* together through a point-to-point contact, which itself is insulated by the binder being used, reduces the thermal conductivity of the steel shot core to a low value.

When metal is cast against such a core the layer of sand grains and steel shot at the surface removes heat from the casting at a relatively high rate. However, as soon as the surface grains are in temperature equilibrium with the casting, which probably occurs almost instantaneously, the core can then absorb heat no faster than heat can be removed from the hot metal-core interface through the core.

Radiation

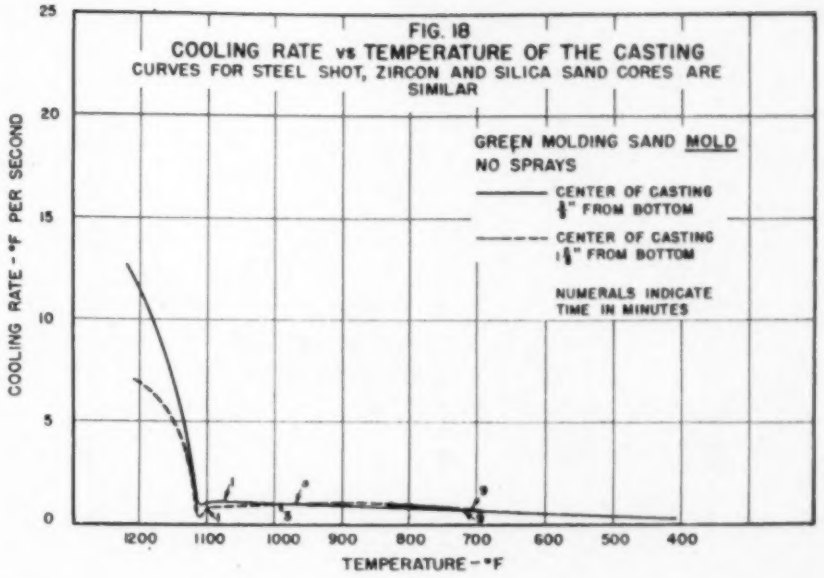
In the case of such cores heat transmission by conduction from grain to grain is slow, and possibly the greater part of the heat removal may occur by radiation[10] and by convection through the gases in the interstices surrounding the individual grains of which the core is composed. This accounts for the relatively low chilling capacity of steel shot and zircon core sands.

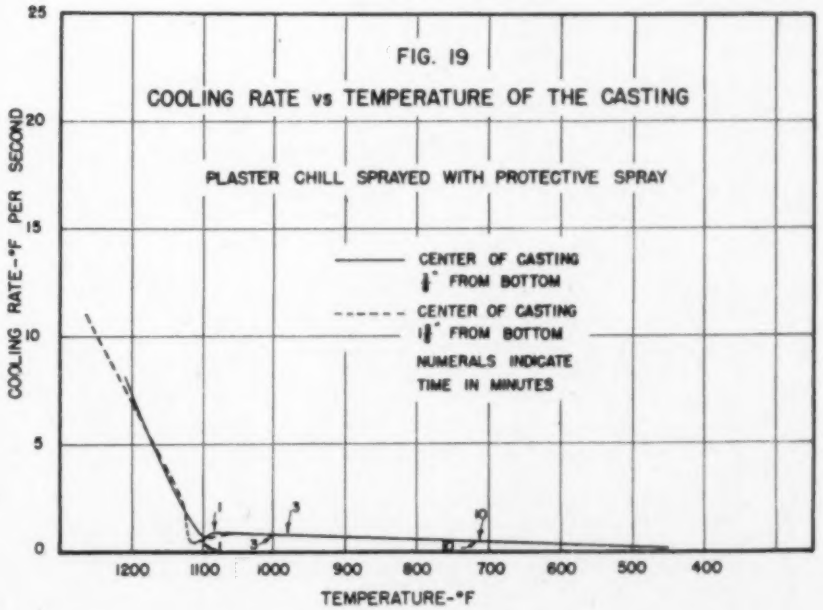
Graphite-base cores had a higher chilling capacity than did the sand-base cores, but the ratio of the chilling capacities was only of the order of 5 to 1. However, the ratio of the thermal conductivities of graphite and silica is of the order of 145 to 1.

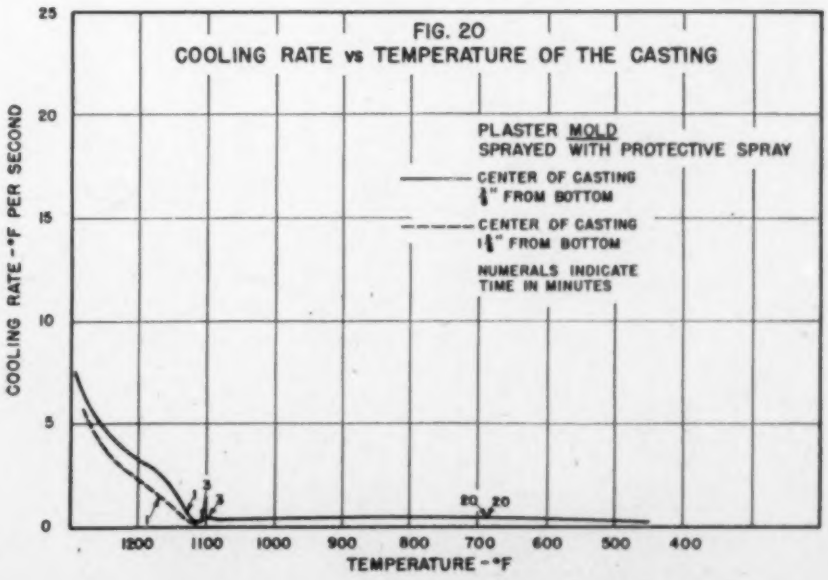
It is thus demonstrated that, although the increased thermal conductivity of the base material results in a small increase in chilling capacity, the main factors controlling thermal conduction in an agglomerate are the resistance to heat transfer offered by the binder and poor contact between the particles.

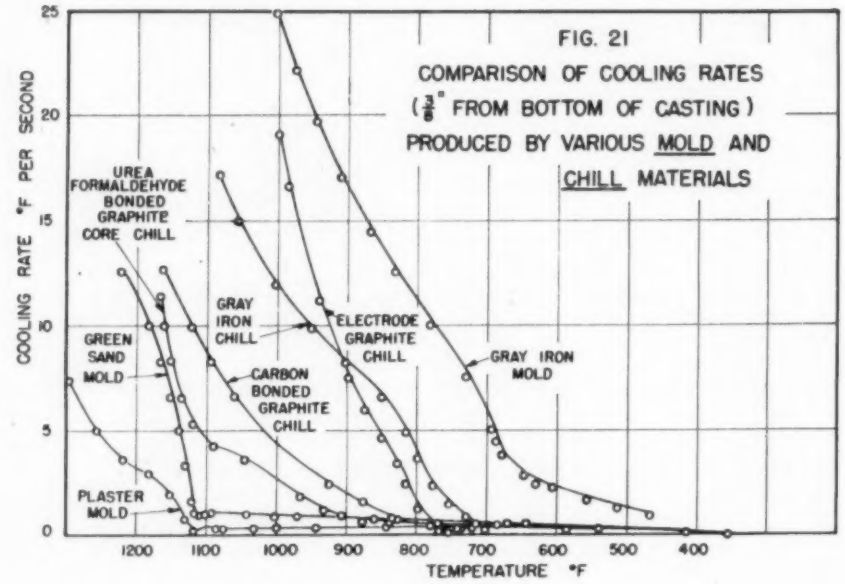
Use of Chill Spray. Castings poured against bare metal chills and molds which were heated and placed in the mold just prior to casting frequently exhibited blows.

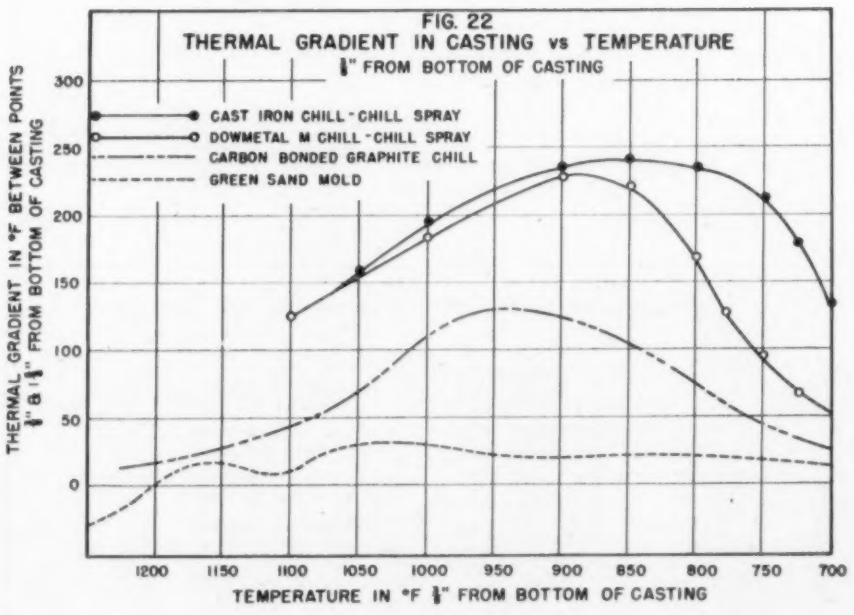
*Thermal conductivity of a piece of silica is about 0.0025 cal/cm/cm²/°C/sec.











These blows locally insulated the casting from the mold or chill, thus reducing the chilling capacity rating of the uncoated chill and mold materials.

When chill and mold materials were coated with a chill spray the tendency toward formation of blows was eliminated. As may be seen from the summary of data, Fig. 4, chill spray had a slight insulating effect upon heat transfer to the chills and molds.

Effects of Thermal Gradients upon Properties. In addition to the effects of the mold and chill materials upon cooling rates and thermal gradients, it is desirable to relate the magnitude of those thermal conditions with the mechanical properties of the resulting castings, and also to demonstrate the importance of thermal gradients in securing sound castings.

Cooling Rates

In the mold tests a wide range of cooling rates was produced, while in the chill tests a wide range of thermal gradients was obtained. Castings made by both methods were sectioned for test bars, and subsequently broken in tension in the solution heat treated condition. By plotting the thermal gradient values of all castings against the ultimate tensile strength of the lower test bars, as in Fig. 24, a curve is produced which demonstrates the effectiveness of increasing thermal gradients on the properties and soundness of the panel.

Similarly, plotting the values for cooling rate at 1000° F. against the ultimate tensile strength of the lower test bars, as in Fig. 25, indicates a lack of correlation. For example, a cooling rate of 1° F./sec. may produce tensile properties ranging from 30,000 to 43,000 psi., and from an unsound to a sound casting. High cooling rates contribute to high tensile properties and casting soundness primarily through the promotion of adequate thermal gradients.

When the thermal gradient was greater than 98° F./in., both bars had better than typical properties (Table 8). When the thermal gradient was above 21° F./in., the properties of the casting were just above the specified minimum.

It must not be assumed that these values hold for all sections

Table 6

RELATION OF MOLD MATERIALS TO COOLING RATES AND PROPERTIES OF CASTINGS

Material	Density, gr/cc	Thermal Conductivity, cal/cm/cm²/° C/sec	Heat Capacity cal/cm³/°C	Spray	Cooling Rate at 1000°F., °F/sec	Thermal Gradient, °F/in.	Average Time to Reach 700°F., sec.	Mechanical Properties — Upper Test Bar — Elongation, % in 2 in.	Yield Strength, psi.	Tensile Strength, psi.	Lower Test Bar — Elongation, % in 2 in.	Yield Strength, psi.	Tensile Strength, psi.	Metallographic Properties H-HT — Massive Compound†	Porosity†	Grain Size, in.	Radiographic Rating*
Molds																	
Cast Iron	6.53	0.110	0.863	None	19.1	73	27	8.0	14,700	33,700	18.0	15,800	44,200	0	0.5	0.002	2
				Chill	13.5	92	28	13.7	15,200	39,200	16.9	15,600	42,800	1	1	0.002	1+
Mg "M" Alloy	1.76	0.310	0.438	None	13	55	50	9.5	15,700	36,900	14.8	15,500	42,700	1.5	3	0.002	1.5
				Chill	12	62	47	14.1	15,400	41,400	17.2	15,300	43,300	1	3	0.002	1.5
Electrode Graphite	1.55	0.347	0.302	None	17.5	72	29	12.9	15,100	39,700	17.6	15,400	43,000	1	2	0.002	2
Urea Formaldehyde Bonded Graphite Core				None	5	51	158	9.8	14,000	37,300	17.0	14,700	43,100	0	0	0.002	2
Zircon Core	2.27	7.91×10^{-4}	0.422	Protective	1	21	452	8.5	13,700	33,800	11.1	14,700	38,600	2	3.5	0.003	4
Steel Shot Core	3.22	10.7×10^{-4}	0.695	Protective	1.2	31	402	9.0	14,300	35,700	12.5	14,900	40,800	1	2.5	0.0025	3
Molding Sand	1.76			None	1	15	551	8.4	12,700	28,900	7.9	12,800	32,700	0	3.5	0.005	4
Silica Core	1.60	10.0×10^{-4}	0.406	Protective	1	19	492	7.4	12,900	30,000	10.2	14,500	36,500	1.5	1	0.0035	4
Urea Formaldehyde Bonded Core	1.60		0.406	None	1	18	558	6.8	12,800	27,600	10.2	14,000	34,800	1.5	2	0.003	4
Plaster	0.62			Protective	0.5	6	1162	2.8	9,200	15,000	3.0	12,400	22,200	3	3.5	0.005	5
Chills																	
Cast Iron	7.43	0.110	0.980	None	7	171	72	15.9	14,500	41,700	17.9	15,700	43,600	0	2.5	0.0025	1
				Chill	4.7	144	116	16.0	14,400	41,500	16.4	15,100	42,200	1	2	0.002	1+
Mg "M" Alloy	1.76	0.310	0.438	None	4.7	120	225	15.3	13,900	41,500	15.1	15,400	42,100	0	1	0.002	1+
				Chill	4.4	117	253	15.1	14,200	40,800	16.4	14,900	42,600	0	2	0.002	1+
Electrode Graphite	1.635	0.347	0.319	None	4	129	286	14.7	14,600	41,500	14.8	15,700	42,200	0	2	0.002	1.5
Carbon-Bonded Graphite	1.70	12.95×10^{-3}	0.427	None	2	58	300	11.6	13,100	38,200	16.1	14,300	42,600	0.5	1	0.002	2—
Urea Formaldehyde Bonded Graphite Core				None	1.3	42	398	14.3	13,300	40,700	15.7	14,800	43,400	0	0.5	0.002	2
Steel Shot Core	3.22	10.7×10^{-4}	0.695	Protective	1	19	518	7.3	13,500	31,300	9.6	14,000	36,300	2.5	2.5	0.0035	3.5
Zircon Core	2.27	7.91×10^{-4}	0.422	Protective	1	16	530	7.5	12,500	27,800	9.9	14,200	35,100	1.5	0.5	0.0025	4
Silica Core	1.60	10.0×10^{-4}	0.406	Protective	1	15	574	6.9	13,400	29,900	8.0	14,200	33,200	2.5	3	0.003	3.5
Urea Formaldehyde Bonded Core	1.60	10.0×10^{-4}	0.406	None	1	12	544	7.5	12,500	30,500	9.5	13,300	34,600	1.5	3	0.0035	4+
Plaster	0.64			Protective	0.8	10	632	5.7	12,400	29,800	5.5	13,300	30,400	3	4	0.003	4

Mold Material	Panel Size, in.	Riser Size	Degassed	Pouring Temp., °F.											
	¾	Small	No	1450	1	15.0	551	8.4	12,700	28,900	7.9	12,800	32,700	0.0	3.5	0.005	4.0
	¾	Small	Yes	1450	0.7	15.2	570	11.3	12,100	34,200	13.5	13,200	39,100	1.0	2.5	0.007	2.2
	¾	Large	Yes	1400	0.6	24.7	758	13.2	13,000	38,400	12.8	13,500	38,600	1.5	1.0	0.006	1.2
	¾	Large	Yes	1450	0.5	24.7	739	12.5	12,300	37,500	14.0	12,900	38,800	1.5	2.0	0.006	1.2
	¾	Large	Yes	1500	0.5	24.0	940	12.8	13,100	38,300	15.0	13,500	40,500	1.0	0.6	0.007	1.5
Molding Sand	¾	Large	No	1450	0.6	22.3	800	7.0	11,800	29,300	13.5	13,000	37,600	1.5	1.0	0.036	2.2
	⅜	Small	Yes	1400	2.0	32.0	215	12.7	13,100	38,700	14.8	13,900	40,500	0.5	1.5	0.004	2.7
	⅜	Small	Yes	1450	2.0	31.1	226	14.7	13,800	40,700	14.2	14,100	41,100	0.0	2.0	0.004	4.0
	⅜	Small	Yes	1500	1.6	32.8	321	11.8	13,600	38,300	16.0	14,000	41,300	0.5	2.5	0.004	3.0
	⅜		No	1450	2.5	30.4	240	11.5	13,500	37,300	14.5	13,900	41,300	0.0	3.0	0.004	4.0

* An arbitrary numerical rating in which the value of 1 is assigned to a radiographically sound panel and a value of 5 to the panels with minimum soundness.
† See reference No. 17 in bibliography.



ALLOY NO. 26390 CAST IRON MOLD-CHILL SPRAY
RADIOGRAPHIC RATING = 1



ALLOY NO. 26084 ELECTRODE GRAPHITE MOLD - NO SPRAY
RADIOGRAPHIC RATING = 2



ALLOY NO. 26219 STEEL SHOT CORE MOLD - PROTECTIVE SPRAY
RADIOGRAPHIC RATING = 3



ALLOY NO. 29159 GREEN MOLDING SAND MOLD - NO SPRAY
RADIOGRAPHIC RATING = 4



ALLOY NO. 29471 PLASTER MOLD - PROTECTIVE SPRAY
RADIOGRAPHIC RATING = 5

Fig. 23—Representative radiographs of magnesium "H" alloy panels cast in molds of various types. Demonstrating the rating system.

of all castings, for they relate only to the casting studied in this work or similar sections acted upon by the same thermal and feeding conditions. For each set of thermal and feeding conditions there is undoubtedly a minimum thermal gradient which will produce a sound section.

Effect of Degassing the Metal. All tests discussed thus far were performed using metal prepared according to normal commercial practice except that no final degassing treatment was applied. Additional tests were made with the green sand mold using metal that was degassed with chlorine. The ultimate tensile strength of the casting made in this manner was improved 20 per cent over the unchlorinated condition, which raised it well above the specified minimum value without improving the thermal gradient.

Riser Inadequate

It must be remembered that in the green sand mold the riser was designed to be inadequate. Assuming that gas is present in the molten metal in quantities greater than the solubility value at the freezing point, we may conclude from the data presented in Table 6 that the effectiveness of dissolved gas in causing gas porosity is greatest under those conditions where the riser is inadequate.

Continuing our views on gas porosity, it is presumed that in the process of metal solidification, shrinkage of the solidifying casting demands either the formation of a void, a collapse of the casting wall, or the feeding of liquid metal. As the temperature of the riser decreases, liquid metal from the riser feeds at a diminishing rate to the casting, but demands of the solidifying casting for "feed" continue undiminished.

In the case of an inadequate riser a stage is reached in which the "feeding pressure" on the shrinking mass, which is the summation of all pressures contributing to feeding, is reduced below the vapor pressure of the dissolved gas. To satisfy its vapor pressure the gas precipitates out of liquid solution, causing porosity.

In a casting solidifying similarly, but with more pronounced thermal gradients, the riser becomes more effective as a feeder. Adequate feeding pressure is maintained for a

longer period of time so that the stage of gas interference with feeding described in the foregoing is delayed, thus reducing or even eliminating its effect.

Use of a Larger Riser for More Adequate Feeding. The reduced significance of dissolved gas on microporosity when the feeding is more adequate is shown in a series of tests performed using the same experimental conditions as before except for an enlarged riser. Castings were made from both chlorinated and unchlorinated metal.

For castings with enlarged risers made from unchlorinated metal the thermal gradient was increased by 65 per cent over that of the original panel, while the ultimate tensile strength was increased by 15 per cent. This demonstrates the normal result of improving the feeding conditions.

Degassing Beneficial

On the other hand, degassing the metal increased the ultimate tensile strength by 4 per cent over that of metal not degassed. Evidently, improved thermal conditions and a more adequate supply of feed metal reduced the deleterious effect of the dissolved gas.

Thin Panel Casting. In order to obtain some idea of the comparative thermal conditions and casting quality attainable by use of a casting and riser of different size, both the panel and riser thicknesses were reduced by 50 per cent, the other dimensions remaining the same. This change did not alter the riser: casting size ratio as used in the original test.

Again the casting was made in a green sand mold. The reduction in casting and riser thicknesses resulted in a doubling of the thermal gradient. The ultimate tensile strength of the thin casting was 26 per cent greater than that of the thick casting.

In additional tests with the thin casting, degassing the metal by chlorination produced no further improvement in the ultimate tensile strength.

Data for these feeding studies may be summarized as shown in Table 9.

Effect of Pouring Temperatures. For the special panels, namely, that with the enlarged riser and the thin panel, castings were poured at tem-

Table 7

COMPARISON OF CHILLING CAPACITIES OF VARIOUS MOLD AND CHILL MATERIALS

Mold Material	Chilling Capacity — Molds — Method A[1] (°F/sec)	Molds — Method B[2]	Chills Method C[3]
Cast iron	19	20	20
Electrode graphite	18	19	16
Magnesium "M" alloy	13	11	14
Carbon-bonded graphite*	—	—	7
Urea formaldehyde-bonded graphite	5	4	5
Oil-bonded steel shot core sand	1.2	1.4	1.5
Oil-bonded zircon sand	1.0	1.2	1.0
Fluoride type molding sand	1.0	1.0	1.0
Urea formaldehyde-bonded silica sand	1.0	1.0	0.7
Plaster	0.5	0.5	0.5

*This material was not available in sizes large enough to be used as a mold.

[1]Based on cooling rate at 1000° F.

[2]Based on reciprocal of time for casting to cool to 700° F., assuming the value of green molding sand to be 1.0.

[3]Based upon the comparable thermal gradients in ° F./in., again assuming the value of green molding sand to be 1.0.

Table 8

TENSION TEST BAR PROPERTIES

(Solution heat treated Mg "H" alloy)

	Ultimate Strength, psi.	Yield Strength, psi.	Elongation, % in 2 in.
Typical Properties	40,000	14,000	12
Specified Minimum	34,000	10,000	7

Table 9

INFLUENCE OF IMPROVED FEEDING UPON PROPERTIES

	Panel Size, in.	Thermal Gradient, °F./in.	Improvement in Ultimate Tensile Strength over Basic Test, %
Basic test (panel cast in green sand mold; metal not degassed)	¾	15.0	0
Metal degassed with chlorine	¾	13.8	20
Riser enlarged; metal not degassed	¾	20.3	15
Riser enlarged; metal degassed with chlorine	¾	22.5	19
Same casting; riser size ratio as in basic test; metal not degassed	⅜	27.6	26
Metal degassed with chlorine	⅜	28.3	26

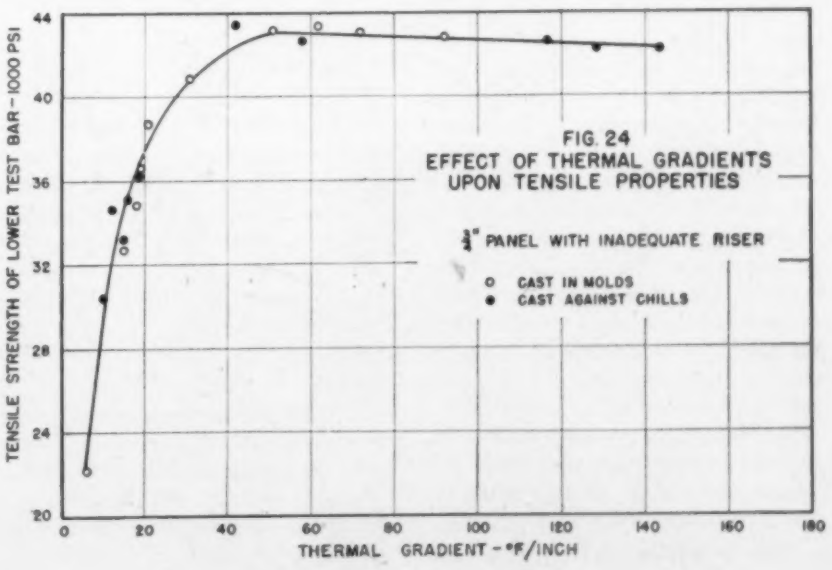
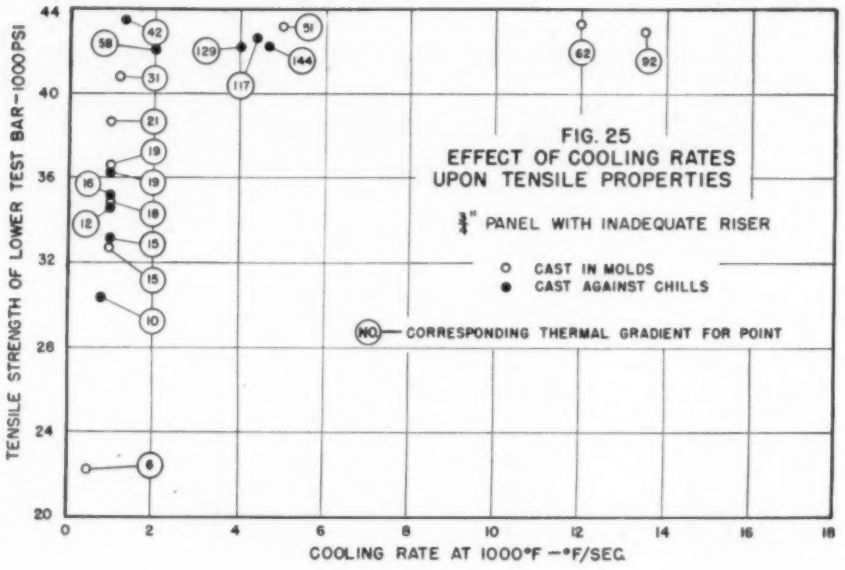


FIG. 24
EFFECT OF THERMAL GRADIENTS UPON TENSILE PROPERTIES



peratures of 1400° F., 1450° F., and 1500° F., using degassed metal. The ultimate tensile strength was improved slightly by increasing the pouring temperature. There was no noticeable effect upon the thermal gradient; however, the time required for the casting to cool to 700° F. correlated with the casting temperature. Increases in pouring temperature produced longer cooling times.

Cooling and Cooling-Rate Curves. Both the cooling and cooling-rate curves display some interesting phenomena. For the slowly cooled panels the cooling curves show a retardation at about 1100° F. The corresponding cooling-rate curves show arrests in the cooling rate at the same temperature.

These retardations and arrests indicate sudden decreases in the cooling rates of the panels at 1100° F. Starting about this temperature (the liquidus temperature) the latent heat of fusion of the metal is released as the casting solidifies. This release of heat continues until the casting is entirely solid, so that the cooling rate of the casting is retarded in the entire solidification range. Curves for the slowly cooled panels do not show any marked change in cooling rate around 700° F. (the solidus temperature). Undoubtedly some change in cooling rate occurs, but the temperature readings in this range were taken at too great intervals to show the effect.

For swiftly cooled panels the cooling-rate curves show definite retardations or arrests in the range between 1100 and 700° F. The arrests at 1100° F. are considered to be similar to those previously described for the slowly cooled panels.

Arrests observed at lower temperatures are thought to be caused primarily by (*a*) release of heat of fusion in the riser as it begins to solidify, and to minor extents by (*b*) release of heat of fusion of low melting point eutectic mixtures formed in large amounts by a shifting of the areas of the equilibrium diagram to the left during rapid solidification, and (*c*) undercooling. It is to be noted that the more rapid the cooling rate the lower the temperature at which the arrests occur.

Conclusions

1. Chilling capacity of a mold material depends directly upon its thermal conductivity and its heat capacity. Since the thermal conductivity of mold materials may range from 0.001 to 0.350 cal/cm/cm²/°C/sec. (350 to 1) whereas heat capacity varies from 0.300 to 0.980 cal/cm³/°C (3 to 1), it follows that changes in heat conductivity will exert normally a greater influence on chill capacity than changes in heat capacity.

2. Chilling capacity of a heterogeneous mold is particularly sensitive to the intergranular area of contact and to the thermal conductivity of the film separating the grains.

3. Use of steel shot core sand, or of zircon core sand, provides slightly greater chilling capacity than does silica sand, but this increase is deemed insufficient to justify their use for the production of magnesium castings.

4. High cooling rates contribute to high tensile properties and casting soundness primarily through the promotion of adequate thermal gradients.

5. Dissolved gas is detrimental to casting soundness when feed metal or feeding pressure is inadequate.

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ENGINEER OF HUMAN RELATIONS

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In the eyes of the average worker, the supervisor is the company. Management policies of human relations, which must be carried out by the supervisor, are only as sound as the understanding and practice of basic supervision principles.

THOSE OF US actively engaged today have seen the era when industry turned the corner from mass production into the age of human relations. Always wide-awake to progress, industry has concentrated on a program of indoctrinating its supervisors with the basic principles underlying sound supervision. Basically, this training is, in a broad sense, a thorough course in psychology or personality development.

Every supervisor recognizes that in the final analysis his men make or break him. Naturally it is recognized that the supervisor represents management to the men, but to the average employee the supervisor is the company. Thus proper or improper supervision can make or break the company.

In industry the training programs take many courses. Some plants advocate the conference method, in which questions and answers serve as the means of providing instructions. Many other means are used, however, to provide new material—lectures, laboratory or experimental sessions, motion pictures and slides. Perhaps a combination of all these means provides the most complete treatment for all classes of supervision.

However, above all, it is particularly stressed and impressed upon the supervisor that he, or she, should learn to develop a genuine interest in people. It cannot be stressed too strongly that by acquiring the habit of success in dealing with employees, the supervisor strengthens his interest in them and with it his power to work with them.

Because of our modern industrialization, it is brought home to us emphatically that man must live with himself and with others. To live means to function, and behavior is the material of psychology. The unit of study is the individual. Psychology is one of the most useful of all of our sciences. Today industry recognizes that successful supervisors must be versatile. They must be skillful and intelligent and—they must have personality.

In one subsurvey studies were made of 3,607 men and women who had lost their jobs. It turned out that 77 per cent had been fired for tactlessness, unfairness, irritability, bad manners, etc. Therefore, specifically, well-liked supervisors must be cooperative, loyal, polite, tactful, friendly, patient, alert, daring, confident and cheerful. These qualities can be attained only by a planned foreman training program.

Not a Short Cut

First, let us stress most strongly that foreman training is not a short cut to success in business, or in life. Do not believe, as some people do, that to train supervisors is to find an open sesame to success. It will not make a mentally strong and powerful individual out of a weakling; but it can be of great assistance in bringing out the latent possibilities of people and in enabling them to adjust themselves to each other.

Industry up to the present generation frequently assumed that the minor supervisors and workers were passive instruments. In other words, the management looked upon the factory as upon a machine. No effort was made, or encouraged, to give employees an opportunity for self-expression. No effort was made to enlist and utilize their personal desires and aspirations. The personalities of the supervisor and workman were completely neglected.

Any show of undesirable individuality (and almost all individuality was considered undesirable) was respressed by discipline. The individual supervisor, or workman, did not count. Men assumed that even such complex organizations as the factory, or the mind, could be best controlled and brought to attainment by mere force.

Enlisting Support

But today, both industry and psychology are turning attention to the component units. Industrialists and psychologists are beginning to realize that the price of control by force is high—they realize the dangers of repressing and the advantages of enlisting the support of the units which compose the industrial and the mental organization.

Inasmuch as a far-thinking person can readily see how industry cannot help but benefit from the policy of training its supervision, let us approach the effect that such training can have on the supervisor personally. The human body may be likened to a plant, the personality being the flower, or bloom. Many plants survive the span of their existence without justifying their energy with its full meaning—that of pro-

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ducing a flower.

Cultivation and a degree of care are necessary to obtain the replete accomplishment. The supervisor, when trained to take a genuine interest in people, cannot help but develop poise and personality. Personal charm arises mainly from a sincere and deep interest in other people and a genuine liking for them.

He learns self-control. He builds up, by practice, the habit of censoring anger at its inception, that is, to develop strong standards of poise and self-control in all dealings. He learns by careful watching to recognize the symptoms of incipient ill temper.

He can learn to recognize that when he feels hot about the eyes, or tingling with indignation, there is trouble ahead unless prompt remedial action is taken. He can learn at the first sign of such a condition to "unhook" from the grip of the situation, to look at the matter as he judges he will a month hence, and to try to understand the point of view of the other side. He can learn to strive, by one means or another, to regain his calm.

By such measures he usually can check anger at the source. If in any case he fails to do so, he can at least get away from the situation—postpone decision, take a recess, make arrangements to go into the matter tomorrow—anything to free himself from the present tenseness and give himself time to cool off. This is a hard thing to do. It demands great self-control, for to do so he must keep in mind that he is "unfit for business."

By repeated effort, even a person with a "hair-trigger" temper can build up a sensitive censoring habit which will warn him of danger at the earliest signs of approaching anger before the excitement has taken a strong hold.

It has been said that "you cannot change human nature" in the sense that the deeper desires of man are for all practical purposes unchangeable, but we certainly can change the habits which give that "human nature" expression. By proper means and training, the supervisor can greatly develop in intellectual interest, ability, and even in character.

We can look to industrial leadership to transform through habit the unfavorable forms of expression of the common human desires into forms conducive to industrial efficiency, well-being, and peace.

Certainly, more than we ordinarily care to think, we are made skillful or inefficient, pleasant or disagreeable, noble or base, by our habits. In all of these phases of ability, or character, the habits we have acquired by learning are of tremendous importance.

Sound leadership, like other phases of supervision, does not come easily to some men, but only after long study and painstaking practice. Thus, it is and should be the purpose of foreman conferences to strengthen and develop the supervisor's leadership and, incidentally, his personality, by changing and al-

The conference method of instructing supervisors.



tering his habits.

A properly organized training program should enable the supervisor to: (1) recognize the nature and extent of individual differences; (2) recognize his influence on the actions of his men; (3) learn to observe the increasing effectiveness resulting from taking advantage of individual differences in getting work done; (4) understand the nature and method of the functioning of group attitudes and morale; and (5) understand the influence of the group attitude and morale upon individual members of the group.

The supervisor of today must be an engineer in human relations. He must manage, direct, control, guide, lead and inspire those with whom he works. Industry today is teaching its supervision "how to think." Industry today is training its supervision not only in the fundamentals of supervision, but also how these principles must and can be made a part of the supervisor's personality.

Industry no longer has a place for the old-fashioned hit-or-miss methods of supervision. Men trained in the old methods must be retrained and brought up to date. Just as industry keeps its machinery up to date, so must supervision be kept up to date.

Training Methods

Primarily, the conference method of instructing supervisors in human engineering is one of indoctrination, making them management conscious, training them in the methods of formulating sound judgment, teaching them the fundamentals and principles inherent in good supervision, guiding them in the application of these principles to their own problems and following up to see that they practice what they are taught.

They are being taught new "habits" and how to apply them with good results. The most successful conclusion to any conference course is when the principles taught to the supervisor are made a part of that supervisor's own personality. One supervisor may apply them differently from another. The principles followed must be the same, but the methods of using them will vary from supervisor to supervisor.

These principles must become an integral part of the supervisor's personality so that they are the basis of his every action. Management must follow up such programs, constantly hammering home the sound principles underlying supervision until they become second nature, or part of his personality, to every supervisor. Indoctrination of supervisors with the principles of sound supervision cannot be overemphasized.

If we are to progress as successfully in the human relations era as we have in the mass production era, it will be necessary for industry to concentrate on a program of indoctrinating its supervisors with the basic principles underlying sound supervision, to see that its supervisors make these principles a part of themselves, and to follow up its supervisors to see that they practice supervision as management in its policies wants its human relations handled.

DISCUSSION

Chairman: F. E. WARTGOW, American Foundrymen's Association, Chicago.

W. F. GRADEN[1]: I am quite interested in this subject of changing a man's mental attitude and creating in him a genuine interest in people. I can not see how we can do it by the discussion method. Is there a part of your program for which you would use the lecture method?

S. G. GARRY: You can use it in the discussion method by grouping the supervisors and presenting the problem to them. Suppose you have a specific problem in mind and everyone gives his opinion on the problem. In that way you can get results from the discussion method because you may find some way of dealing with that problem that the other man did not think about and one that *you* think will work.

Just to give an illustration, I was confronted with a problem that involved an employee who would never change his work shirt. It was a rather sensitive subject to discuss with the employee. How would you approach that man? You know, to begin with, that he is a very sensitive individual. I tried for a week to think of a way to approach the employee but I did not have the solution. In our supervisory discussion I brought it up and we finally agreed on the approach. We decided to wait and approach him at the opportune moment, not to discuss it while he was in his work area.

After approaching the employee and passing a few casual remarks the discussion turned to work shirts and how hard they were to get. Right along with that point it was brought out how much longer shirts would wear if they were washed periodically. Do you see what you are doing? You are getting that individual to do the thing you want him to do by putting it in such a way that he will want to do it knowing that he will benefit by it. Whenever you tell a man that he is going to save money by doing a certain thing, he is interested.

To answer the second part of your question, you can not put it over too well in lectures. An individual can absorb only 7 sec. of instructions at one time. I think the discussion method, in which questions and answers serve as a means of getting a message to supervisors, is more beneficial than either slides or lectures because you get so many different viewpoints. There are many things that you can get in a discussion that are impossible to get in a lecture.

MEMBER: How can you get more work out of the union committee, the "invisible foreman?"

MR. GARRY: Your foreman conferences will teach you how to get along with those men. I do not mean you have to give them everything they want because there are ways of telling people "no." Put it to them in a way that they will accept.

MEMBER: In our work they set the standard of how much work they think is a fair day's work.

MR. GARRY: They set the standards? Wherever you have a group that feels they have the authority to change standards it becomes the foreman's job to change that idea if he is not in agreement with it. Our part of the program is to try to put that information over to them in the best way we know how. We are not always going to be successful.

MEMBER: This training problem is a serious one. From listening to your remarks, I think that your efforts have been directed to trying to combat the unionization of foremen. Nevertheless, there is another job to be done. I have participated in foreman conferences since World War I. I do not think techniques or procedures offer the answer.

You have stressed in your paper the application of the conference technique. What other procedures do you use with your foremen to try to inculcate into the newly appointed foreman that he is a part of management? Do you use such things as special sick leave, special insurance, compensation, vacation with pay, and other privileges and prerogatives that appeal to some of the higher levels of management?

MR. GARRY: In this discussion we talked about not being able to take any program and make it suitable for all the people that you have. We have our programs for our new supervisors in which material will be discussed that pertains to the new supervisors. As he progresses, so the subject material progresses. In other words, we start him at the bottom as far as foreman conferences go. We do not start him at the top with conferences in economics or effective speech. The only way that I can answer that is by saying that we do have the material to satisfy the needs of the individual supervisor.

I do not think that your foreman training program will cover all of your problems. Your foreman program is made up so that we can help the supervisor with some of his difficulties and relations are the biggest part of his troubles. The foreman training program is not going to stop strikes because if the people want an increase of 20 cents on the hour and management does not feel that they can pay it, the supervisor is not in a position to make that decision. What we are trying to do in the foreman conferences is

to try to help the supervisors with relations problems.

CHAIRMAN WARTGOW: That was a management question. Of course we consider the foreman a part of management. Nevertheless, he can go only so far. My suggestion would be to start your foreman training with top management.

MEMBER: With the particular association I am with, we do not have anything that is unique. It has been tried. We went through a strike. In the last 25 years I have been looking for something to solve the problem.

I know a good many companies in these United States who have been blessed by not having, first of all an organized shop and, secondly, have not lost an hour of production and who have a very loyal foreman group. I know of instances where they have done no foreman training and still have the conditions that I would like to see in our organization and a lot of other people would like to see in theirs. What have they got that we do not have and that we can not see? I would like to know the answer to that.

MR. GARRY: This will be only a partial answer. You speak of organized foreman groups that do not participate in foreman training. You may have a very high quality of supervisors in that particular department that are so far above the average foreman that they already know how to get along with people. Not all of us have to be taught how to get along with people. To some it comes naturally. They have pleasing personalities and everybody gravitates to them. That is the reason I say that the foreman is the "king pin" in the department; as he goes, so go the people.

C. W. JINNETTE[2]: Has there not been a ceiling on wages and salaries paid to foremen? Have the foremen not become disgruntled and has their loyalty not been disturbed?

MR. GARRY: I could cite you many instances when they were disgruntled and management could do nothing about it because of the ceiling on wages. We did have conditions existing where the man on the machine made more money than the supervisor. I think the foremen of World War II are the martyrs of this war. They have had to put up with more than any other group. Any supervisor who had 40, 50 or 100 men under him had that many individual problems. And believe me, they were individual problems.

T. E. MALPASS[3]: We are a little foundry. We have been in operation since 1883 and we have never had a strike. I want to add, though, that we do not think it is a disgrace in this day and age to have a strike. We take our hats off to a foreman who says, "Either give us a day's work or get out." That man is giving the country production. I take my hat off to that foreman because he is fighting a difficult problem.

Did you honestly get that man to change his shirt?

MR. GARRY: Yes.

MR. MALPASS: By just talking to him?

MR. GARRY: Yes. Later he told me that when he got home that night he found out why I discussed shirts, which from our standpoint was to be expected. All we tried to do was to keep from embarrassing the employee.

MR. MALPASS: You mentioned that if you had a man on a job snarling up the details, you should find out what is wrong at home; why he is not doing his job right. I wonder how long you would leave him on that job because there are some things at home that you nor I can change. You might be doing an injustice to ten men in that team by leaving him there. I do feel that during the war we took men into the shops who had never had a job in their life. The sooner that we can weed them out, the less trouble we will have. I think you have a big problem if his wife is buying $150 worth of books from every salesman that comes to the door. Can you change that?

MR. GARRY: In some cases yes, in other cases you would not be able to change it.

In answering your question as to how long I would leave him on the job, I think that should be the job of the foreman who knows the individual. These cases are not going to work out perfectly but the supervisor is always striving to help an employee. If he puts forth a little effort then he is going to show a little progress.

MEMBER: In selecting a supervisor, is there any general rule to promote a man and spend a lot of time training him for that job or to get a man from outside?

MR. GARRY: When you are looking for a supervisor you are looking for a leader. You can teach a man the practical side of the business, but if he already has leadership, the rest of his work is made easier.

Now as to whether I would take him out of the ranks or go outside, I would rather take him from the ranks, bring him up in the work he is familiar with.

MR. GRADEN: I think it is absolutely one of the worst things that management can do to get a man from the outside. We have had some personal experience along that line and if there is anything that will destroy the morale of the men in the department it is going outside to bring a man in.

N. KITCHEN[4]: In this day and age there are very few men who want to come in and do the dirty foundry work. Ninety-nine times out of 100, in our own experience, it is a man who has been thrown out of every other industry in town. We try to follow that program of dragging a man from the ranks, but in many cases the department fails to materialize or develop a man with that ability.

We have instituted a program whereby we hire men who we feel have the necessary background and we tell them when we hire them that we are going to keep in personal contact with them in the hopes that they will be the foreman we are looking for. In other words, we are trying to institute a program to better our personnel. In many cases they are your secret foremen if you are fortunate enough to find them out. But we find mostly that it is an individual problem rather than a collective problem.

We had a matter of goggles at pouring time. The men would not wear them in spite of the fact there is a state law that they must be worn. Eight months ago we approached the problem in a different manner. We bought every type of goggle made. If we asked a man why he would not wear them he might say that he can not see out of the side so we give him the type of goggles that give him perfect vision all over. Or he might say that he liked the kind used in the grinding room, so we give him a pair of those. We have eight different types of goggles. It was strictly an individual problem. If you are small enough so that you can make individual contacts, I think you get far better results.

CHAIRMAN WARTGOW: I think we have a tremendous job to get young people to go into a foundry and work. You have got to build the job up for them.

MEMBER: The way that we tried to overcome that in our foundry was in our new union contract we tried to put in all the previous problems we had with them before. So far it has worked out very well. It is about two months old. There have been several times when they have tried to do something that they did before but now it is in the contract.

The story about the shirt interested me very much. I approached it from a different angle. I go up and bet a man a shirt that he could not do a certain trick with the provision that he has to wear the shirt. He catches on and he comes over to me and apologizes for being that dirty. He did not realize it and did not think it made any difference.

Talking about these invisible foremen, we just had them to put on a new shake-out program. We had a young strike on our shake-out group. They had a colored boy and a white boy who did most of the talking. When it came time to choose the foreman, I chose the white boy and I think he is going to be a success. I think he is going to be just what I want, because, in the first place, I knew he was capable. In the second place, I tried to impress upon him that I did not want him to work for nothing but I did expect a good day's work for a good day's pay. If I was overworking the men, I had to be sold on the idea that I was expecting too much. Until he could sell me, I was right. I think I will have success.

MEMBER: We are talking about foreman training. I am sure Mr. Garry will join with me in saying that foreman training alone is not the answer. We may have to do even a little more than on the job training, but it is an overall job. One of the solutions of getting good bona fide men is to have an adequate supply of apprentices.

It was my pleasure to meet a good many of our former apprentices over the past 25 years and there is not one of them who I met who is not today a foreman. I think that is one of the sources of good foremen so when we talk about training, we should think of the overall training program.

[2]Simonds Abrasive Co., Philadelphia.
[3]Norton Co., Worcester, Mass.
[4]East Jordan Iron Works, East Jordan, Mich.
[5]Sweet's Foundry, Johnson City, N. Y.

WET TYPE DUST COLLECTORS

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Dust particles, in wet type dust collectors, are entrapped in a liquid, usually water, and conveyed by the liquid to a settling tank. Four general types of wet type dust collectors, namely, spray or water curtain type, dynamic precipitator type, impingement type, and the combination types of collectors are discussed.

WET type dust collectors have a common basic operating principle. Dust particles are entrapped in a liquid, usually water, and conveyed by the liquid to a settling tank or disposal area.

General Types

The method of entrapping the dust particles varies with the design of the collector. The desired water patterns for collection are obtained by using a fresh water supply, recirculating pump, rotor, or the air stream itself. The types of wet collectors are as follows:

1—Spray or water curtain type.
2—Dynamic precipitator type.
3—Impingement type.
4—Combination of the above types.

Spray or water curtain type collector. In the spray or water curtain type dust laden air is drawn through one or more stages. The water pattern is produced by nozzles, high air velocity, or a rotor. A considerable amount of water is carried upward in the air stream to the moisture eliminators where it is separated from the air and drains downward flushing the surfaces.

Usually the liquid is recirculated after a settling period in an integral tank. A small amount of make-up is required to compensate for evaporation and sludge handling.

Dynamic precipitator type collector. In the wet type dynamic precipitator (Fig. 1) the collecting unit also serves as the exhaust fan for the system. Sprays are located in a precleaner section and in the impeller unit.

The entrained liquid and dust are thrown against the impeller blades and the dust and water precipitated through a by-pass into a sludge hopper or tank. Usually the liquid is not recirculated but runs off through an overflow pipe from the settling tank. Ordinarily a fresh water supply is required.

Impingement Type Collector

Impingement type collector (Fig. 2) is usually designed to direct air flow so as to create the desired water pattern and impinge the dust particles against wetted surfaces in one or more stages. The water is admitted near the top of the collector, just below the moisture eliminators. It descends against the air stream washing all the surfaces and carrying the dust downward.

A settling tank may be used or the effluent may be pumped or drained to the disposal area. Usually the liquid is recirculated unless the effluent is taken directly to the disposal site. A small amount of make-up is required to compensate for evaporation and sludge handling.

Combination Types of Collectors

In combinatiton type collectors (Fig. 3), the dust laden air is usually first subjected to a spray and then forced through nozzles or bubble caps so that the air passes through the liquid. Sometimes the liquid is treated with a wetting agent to form a foam bed through which the air passes. The dust then settles to a sludge tank. In some designs the liquid is recirculated and in others drained off. Make-up is required for evaporative and sludge handling losses.

Foundry Applications. Most wet collectors have a relatively high efficiency and are used for dust control of all foundry operations requiring collection including grinding and cleaning room equipment. They are especially well adapted for handling air at a wide range of temperatures and moisture content as in the case of shakeout and sand conditioning operations. Collectors have the additional operating features:

1—No fire hazard present.

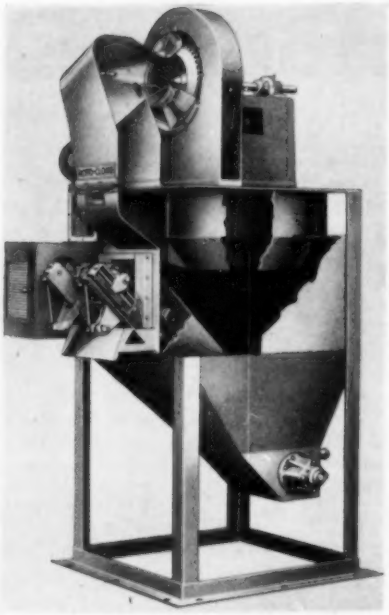
2—No health hazard in handling the collected dust since it is in the form of sludge.

3—Constant resistance for a given system eliminating fluctuations in air volume.

4—Continuous operation is possible since the collected dust is washed down constantly to the settling tank.

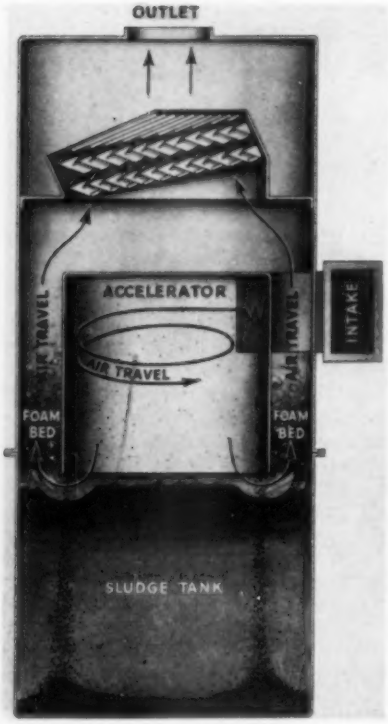
Collector installations. Collector installations are varied to suit requirements (Fig. 4 and 5). They range from a small completely self-contained unit for an isolated operation to a complete foundry system involving a number of collectors. In larger projects a study should be made to determine the most economical arrangement consistent with flexibility of foundry operations, satisfactory performance, and ease of maintenance. The tendency in our company has been to locate the collectors as close to the dust pick-up points as practical and to use a central dewatering system.

Presented at a Session sponsored by the Safety and Hygiene and the Plant and Plant Equipment Committees at the Fiftieth Annual Meeting of the American Foundrymen's Association at Cleveland, May 6, 1946.



(Courtesy, American Air Filter Co., Inc.)

Fig. 1 (Above)—Dynamic precipitator type collector with settling tank and sludge ejector.



(Courtesy, The R. C. Mahon Co.)

Fig. 3 (Above)—View of a combination type collector where dust-laden air is subjected to a spray treated with a wetting agent forming a foam through which the air passes (Patented).

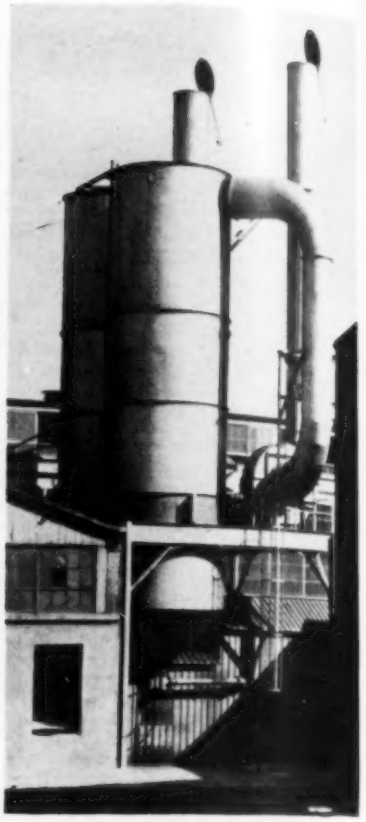


Fig. 5 (Above)—Collector outside of building to avoid loss of working space.

Fig. 2 (Below)—Cut away view of an impingement type collector. Water descends against the air stream washing all the surfaces and carrying the dust downward.

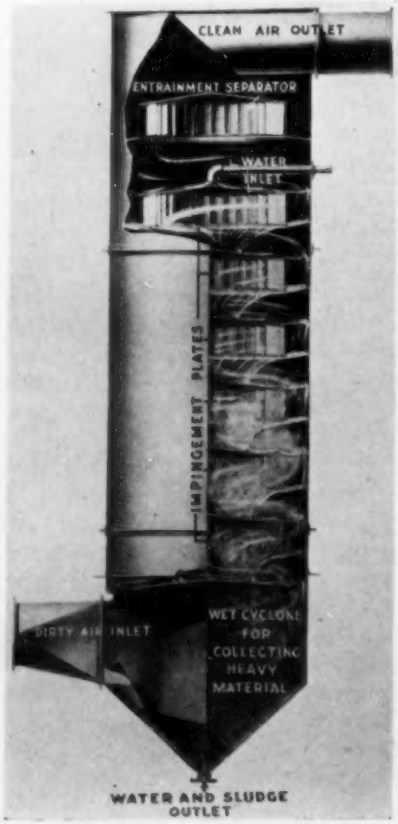
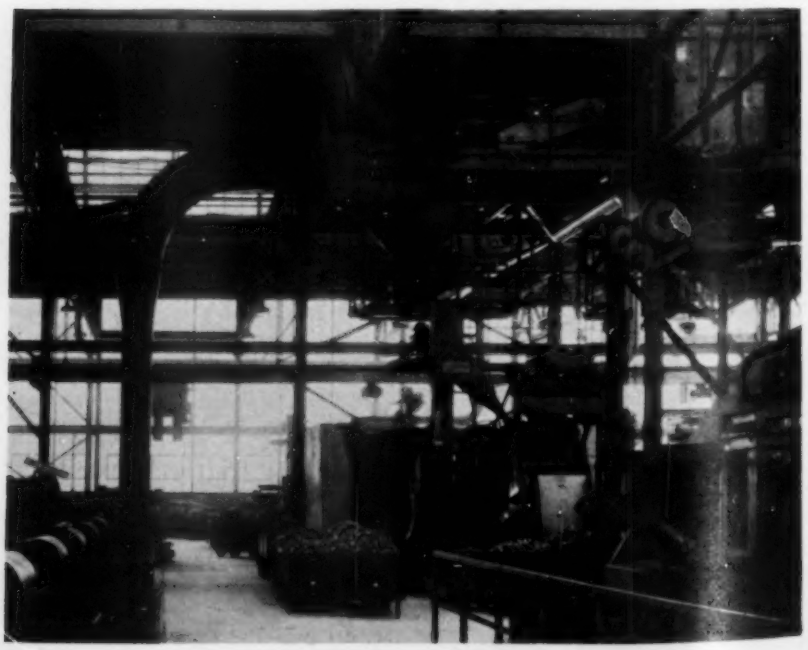


Fig. 4 (Below)—Cleaning room collector supported overhead to avoid loss of working space.



This has the advantage of saving power and reducing duct maintenance. Collectors should be located to avoid using valuable working space. They may be located above the floor or those types having a separate settling tank may be placed outside without danger of freezing provided the inlet air is well above a freezing temperature.

Collector operation. In operating wet type collectors care should be taken to follow the manufacturer's instructions. Otherwise, inefficient performance and increased wear may result. There should be an adequate supply of water (Fig. 6). Ordinarily the water supply should be interlocked with the fan motor to prevent the collector from being operated dry.

It often is advantageous to run water through the collector a short time after the exhaust system has been shut off to flush the collector and effluent lines when used. The collector should be inspected periodically, perhaps once a month. Moisture eliminators should be given especial attention as more build-up usually occurs there than in other portions of the collector. A small amount of preventative maintenance will save a lot of time later on (Fig. 7).

Sludge Handling. In general the collected dust has little value and can be readily transported to the disposal area. The method of handling is influenced by sludge quantity, relative location of settling tank to dump, and transportation facilities available. The solids are removed from the tank by hand, gravity, or conveyor or other mechanical means.

Equipment used for carrying the solids to the disposal site may range from a wheelbarrow to a railroad car. However, dump buckets are most frequently used. It is sometimes possible to eliminate most of the sludge handling by pumping the sludge from the settling tank to the dump. Such a method should be laid out carefully to avoid plugging the pipe.

Equipment Selection. Dust collecting equipment should be selected that will give required results with minimum initial and upkeep costs. Due to variations in installations and present conditions, no reliable cost information can be given for complete systems.

The cost may run from a few hundred dollars per thousand cfm. of air handled in large systems to several hundred dollars per thousand for small systems. For this reason dust pick-up points should be combined in as large a system as possible without sacrificing flexibility of foundry operations.

Summary

General types of wet collectors and their application and operation have been briefly discussed. There are several variations from those mentioned and their installation and operation may have to be modified accordingly. However, most wet type dust collecting equipment can be installed with a minimum of interference with foundry operations and will give satisfactory results.

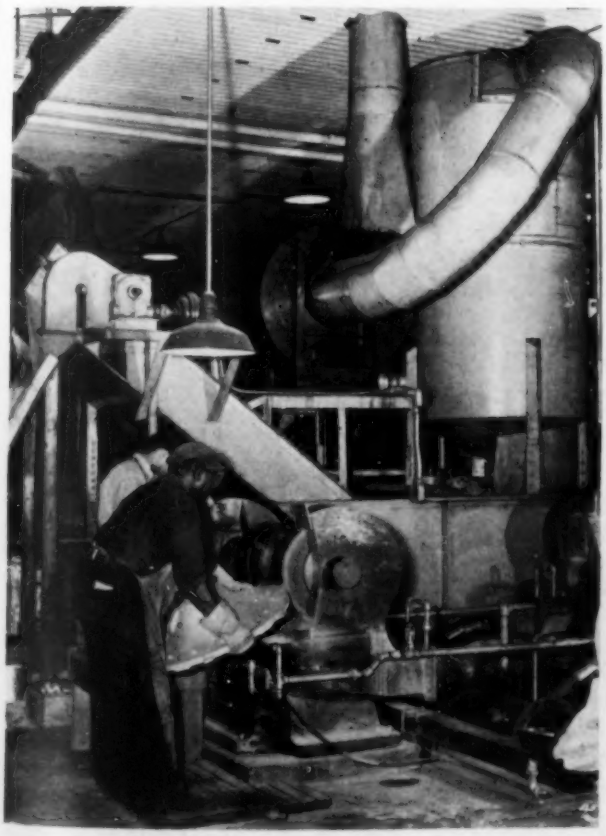
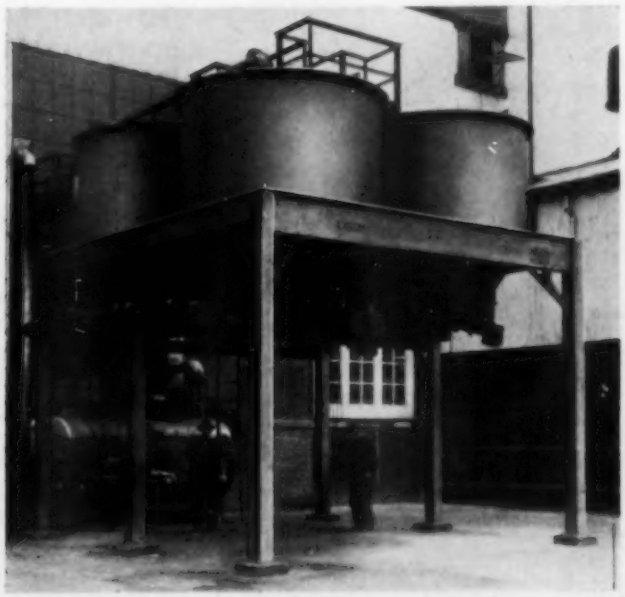


Fig. 6 (Left)—Conveyor type recirculating tank. Fig. 7 (Below)—Bottom dump recirculating tanks for loading directly into truck or railroad car.



BORE CRACKS IN CAST STEEL FITTINGS AND VALVE BODIES

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Bore cracks are hot tears which occur in the bores of valves and fittings under the flanges. They are caused by stresses which result when the ends of a valve or fitting and the comparatively thin barrel contract circumferentially and pinch down on the core, thereby restraining linear contraction from taking place. Bleeding tests show that the bore crack occurs after the barrel of the casting has solidified, at which time the stresses which prevent linear contraction exceed the strength of the metal skin beneath the flange. This defect can be eliminated by two means: (1) by the use of a hollow core, which can be deformed by the circumferential contraction of the casting, automatically relieving the stresses and permitting linear contraction to take place, and (2) by the use of graphite collars or properly designed steel chills in the core which chill and strengthen the metal in the casting so that its strength exceeds the imposed stress. Theories of the cause and formation of veins, which are closely allied to the bore crack, are also presented.

BORE CRACK IS THE TERM applied to a fissure-like defect which frequently occurs in the bores of valves and fittings. These so-called bore cracks extend circumferentially around the bore directly under the flanges and may vary in extent from a few degrees to the complete circumference of the bore.

These cracks occur in a large number of the valves and fittings made commercially and, unless the crack is repaired, leakage may result. Cases have been known where the defect extended so far into the flange that it reached the bolt holes. When such fittings were placed in service, steam leaked through the crack and out around the bolts.

Repairing the cracks is time consuming and difficult due to their inaccessibility, and naval vessels are sometimes put out of service for a day or more while a bore crack in a valve or fitting is repaired. The purpose of this work was to determine the causes of bore cracks and to develop foundry practices to prevent their occurrence.

General Considerations. Bore cracks, except in severe cases, are not visible to the naked eye and their prevalence in valves and fittings was not known until magnetic powder testing was used to detect them. The Boston Navy Yard foundry was among the first to test valves and fittings on a large scale by magnetic powder methods, and they found that a very large percentage of their castings contained the defect.

Other foundries manufacturing valves and fittings reported the same startling results. One manufacturer of valves and fittings has kept statistics on their castings and found that the crack occurs more frequently in fittings than in valves, and also that the prevalence and severity of the crack increases with the bore diameter of the casting.

Although this defect must have occurred in many thousands of valves and fittings made since the steel casting industry came into existence, practically no technical data with which to start in obtaining a solution to the problem were available because the prevalence and seriousness of the defect was not realized. A diversity of opinion existed among foundrymen as to the real character of the defect; some believed it to be a hot tear, others shrinkage, and some suspected that core gases were instrumental in forming the crack.

The most generally accepted theory and the one to which the authors subscribed was the defect was a hot tear. Hot tears are caused by hindered contraction and can occur only when (1) the design of the casting is such that the contraction of one member of the casting is restrained by other members of the same casting, or (2) sand or cores of the mold restrain contraction.

Since the design of fittings does not make them susceptible to this type of tearing, the first of these causes can be ruled out. A hot tear due to the design of valves may occur if the valve seat cools more rapidly than the rest of the casting and tears away from the valve body, but it is inconceivable that this could have any bearing on the tears under the flanges.

It was by no means apparent how the mold or core could cause the tears to occur at the position in which the bore cracks are always found. The most logical place for tearing to occur in a fitting would

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NOTE: The opinions expressed are those of the authors and do not necessarily reflect the views of the Navy Department.

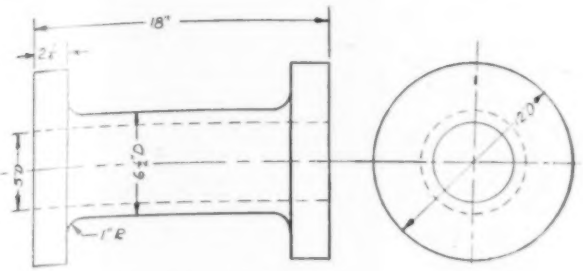


Fig. 1—Large test fitting.

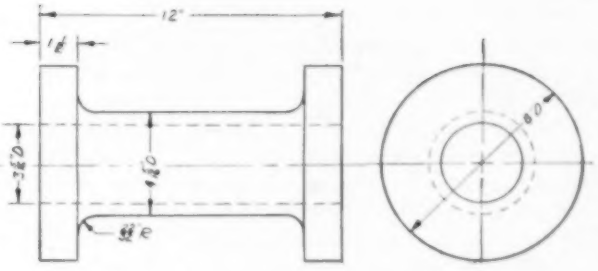


Fig. 2—Small test fitting.

be at the junctions of the flange and body on the outside of the casting. Had tearing occurred there, it would be apparent that the sand between the flanges prevented over-all contraction of the casting and that the barrel had simply pulled away from the flanges.

If the defect is a hot tear, hidden stresses which caused the crack must be present somewhere in the mold. These stresses need not necessarily be great. Korber and Schitzkowski[1] have determined that hot tears in a plain carbon steel occur in the vicinity of 2370° F. (1300° C.), at which temperature mold resistance begins to hinder contraction. K. Singer[2] and others agree that Korber and Schitzkowski's conclusions are correct.

Hall[3], in England, has determined the tensile properties of cast steels at elevated temperatures and found that at 2370° F. (1300° C.) a 0.26 per cent carbon steel has a tensile strength of about 1,600 psi. and negligible elongation. At 2552° F. (1400° C.) this steel has a tensile strength of approximately 200 psi. and no elongation. Hall's test bars were cast in sand molds contained in an elaborate machine and were pulled when the metal cooled to the desired temperature.

Dupuy[4] has also determined the high temperature properties of plain carbon steels. However, his specimens were machined from ingots and heated in a carbon resistance furnace mounted on his tensile machine. His tests show that at 2370° F. (1300° C.) steels containing between 0.15 and 0.55 per cent carbon have a very low tensile strength; but, contrary to the findings of Hall, these steels have extremely high ductility.

Dupuy found that as the steels were heated to about 2100° F. (1150° C.) they showed reduction of area values of 100 per cent and maintained this extreme ductility up to about 2550° F. (1400° C.); above this temperature the ductility dropped almost to zero. Dupuy's work was undoubtedly more accurate than Hall's, but his experiments did not reproduce the conditions of steel cast into sand molds as did Hall's. However, Hall and Dupuy agree that around 2550° F. (1400° C.) both the strength and ductility of steel are quite low.

Briggs and Gezelius[6, 7, 8] have made comprehensive studies on hindered contraction and the mechanism of hot tear formation and believe that hot tears occur at temperatures of 2370° F. (1300° C.) or higher. They have pulled tensile bars cast into molds contained in a tensile machine after the fashion of Hall, although their machine was of a completely different design and of simpler construction.

Hot Tear Temperature

They found that at 2370° F. (1300° C.) a Navy "Class B" steel of 0.25 per cent carbon had a tensile strength of approximately 1800 psi. and an elongation of about 5 per cent, while at 2550° F. (1400° C.) the strength was 900 psi. with less than 0.5 per cent elongation. It thus appears from work published to date that steel castings are most prone to tear between 2370 and 2550° F. (1300 and 1400° C.).

It is the authors' opinion that the hot tears occur at temperatures near 2550° F. (1400° C.) where both the strength and ductility of the steel are low. If the strength of the solidified section of metal were sufficient to cause deformation of the sand as the metal contracted, no tearing would occur even if the ductility of the metal was low.

If the metal had sufficient ductility to flow plastically as contraction occurred, it would not tear even if it had little strength. Thus, it is reasoned, both low strength and low ductility are requisite to the formation of hot tears. This condition exists at temperatures very near the solidus.

Method of Procedure. The test castings used for this work were the fittings shown in Figs. 1 and 2. A fitting was used rather than a valve casting primarily because it had been observed that the defect occurs more frequently in fittings.

Figure 1 shows the fitting with the 5 in. diameter bore which was used first, but was soon replaced with the smaller fitting of Fig. 2 which required only a quarter of the amount of metal needed for the large fitting and still exhibited the bore crack clearly. The sand used in making the molds was a New Jersey silica sand with an A.F.A. fineness number of about 70. The composition and properties of the molding sand and core sand were as shown in Table 1.

These compositions make hard, strong molds and cores, and severe bore cracks were obtained under both flanges of the test casting. The metal used in the tests, except where otherwise noted, was Navy "Class B" steel of approximately 0.25 per cent carbon, 0.60 per cent manganese and 0.40 per cent silicon. The melting for most of the castings was

Table 1

SAND COMPOSITION—PROPERTIES

Components, per cent	*Molding Sand*	*Core Sand*
Bentonite	2.5	—
Dextrine	1.2	—
Cereal Binder	0.3	0.5
Sand	96.0	97.5
Kerosene	—	0.5
Linseed Oil	—	1.5
Moisture	3.8	5.0
Properties		
Green Permeability	85	—
Green Strength, psi	3.5	1.5
Dry Permeability	120	90
Dry Strength, psi	300	1000
Hardness	60	—
Flowability	80	85

done in a 300-lb. basic induction furnace and a few heats were made in a 1,000-lb. basic-lined arc furnace.

In preparing the castings for examination, the flanges were removed from the barrel, either by cold sawing or flame cutting. The flanges were then sawed across the diameter to make the bore of the flange more accessible for magnetic powder testing. Figure 3 shows a pair of flanges from a test fitting with the black magnetic powder indicating the bore crack, and Fig. 4 shows a flange sectioned, with the magnetic powder indicating the depth of the crack.

In making the magnetic powder test a D.C. magnetic testing machine was used. Prods were placed on each sawed surface of the half-flange so that the current passed completely through its semi-circular section. A current of 400 amperes was used with low voltage. Figure 5 is a photograph of a casting being examined with magnetic powder.

Bore cracks cannot be readily detected radiographically, and the only effective means of observing them, other than by the use of magnetic powder, is hot acid etching.

Since it was believed that the defect was a form of hot tear, the known remedies for eliminating more common hot tears, such as chills, brackets, and unconventional gating methods were applied to the test fitting in the experimental work. Since stresses must be present during the formation of a hot tear, castings were made to determine the location of these stresses. The stresses could arise from three sources: casting design, mold resistance, and core resistance.

Experimental fittings were also made to determine how the bore crack was influenced by foundry variables such as pouring temperature, mold atmosphere, metal composition, and gating and risering methods. The results of these tests are described in the following paragraphs.

Gating and Risering. The ideal method of molding a pressure fitting such as is shown in Fig. 1 is to use a side riser on the face of each flange with gates leading into each riser. The riser may be either blind or open. Because the barrels of the fittings are of thinner section than the flanges and also because the metal in the barrel, being furthest from the gate, is the coldest, solidification begins in the barrel and proceeds to the flanges and then to the riser.

While this gating and risering method is best for obtaining a casting free from shrinkage, it aggravates the condition which produces hot tears in the casting. Before a hot tear can occur, two conditions must exist: a hot zone must be present, at which tearing occurs, and an external force, usually mold resistance, must be present to restrain the cooling contraction of the casting.



Fig. 5—Examining a flange with magnetic powder.

Hot zones in the pressure fitting are the flanges, since they are the heaviest section. The flanges, if gated in the manner prescribed by good foundry practice, are preheated by the metal flowing through them and the heat concentration in the flange areas is increased.

To eliminate this preheating effect a fitting was made, gated through 11 finger gates of ⅜-in. square cross section spaced along the entire length of the casting. With this gating procedure no part of the mold was appreciably preheated. However, examination of this casting showed that the bore crack persisted. The hot zone at the flanges was still of sufficient intensity to enable the bore crack to form.

Test Gating Method

A test fitting was then made by gating through four small blind risers, two on each side of the barrel, with open risers on the flanges. This gating method further reduced temperature gradients since the barrel was thus preheated by the metal flowing through it, and the flanges, which normally have a high heat content because of their mass, were filled with the coldest metal.

This gating method did not prove to be the solution to the problem although, in the several castings poured in this manner, the bore crack was less severe than it was in castings gated through the flanges.

Some of the large test fittings were made by gating into the parting line

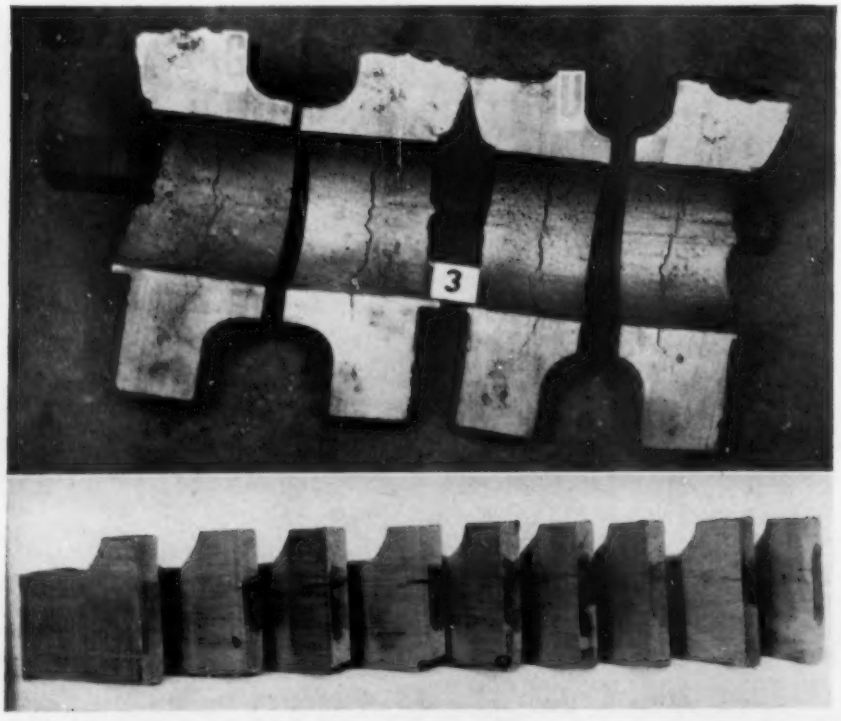


Fig. 3 (top)—Typical bore crack indicated by magnetic powder. Fig. 4 (bottom)—Flange sectioned to show depth of crack.

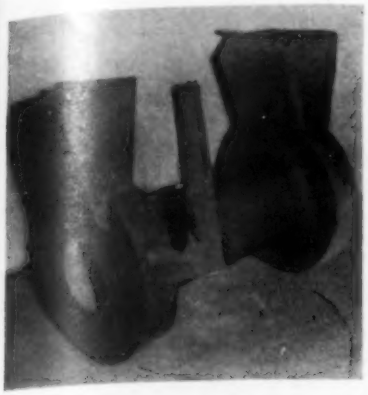


Fig. 6—Fitting as removed from mold showing gating and risering method used in making test castings.

with padded open risers on top of the flanges, as shown in Fig. 6. Although, theoretically, top risers do not feed as completely as do side risers, no difference is apparent in the susceptibility to bore cracking of fittings risered at the top or at the side. It has always been noted that the bore crack has a tendency to form in the bore opposite from the riser. With top risers the crack usually is on the bottom of the barrel, and with side risers the crack is found on the side opposite the riser.

It was decided to use the top riser method in subsequent tests because it permitted the use of a smaller flask and less sand. The small test fitting of Fig. 2 was also adopted at this time since it was found that it exhibited the bore crack very clearly and required only 25 per cent as much metal as the larger one.

It was noted that if only one of the two flanges of the test fitting was gated, the bore crack under the gated flange was more severe than that under the ungated flange. The cracking under this flange was also more severe than that in fittings in which the metal was introduced through both flanges.

Many of the castings in subsequent tests were gated into only one flange on the supposition that this was a more severe test and that any corrective measure found under these conditions would work under any molding conditions.

To test the possibility that, because of compression of the casting by the head of molten metal, bore cracking might not occur in a fitting made in a vertical position, the small test fitting was molded in this manner with two blind risers on the bottom flange and two open risers on the top flange, with the gate leading into the bottom.

Severe tears were present in both flanges, as shown in Fig. 7, with the bottom flange having the most severe tears. The pressure of this column of metal does not act to prevent bore cracks.

Effect of Molding Sand Between the Flanges. Steel begins to solidify as a skin on all mold surfaces and, as the metal cools, the skin becomes increasingly thicker. One theory of the cause of bore cracks, which was disproved in tests, was that the sand between the flanges of a fitting prevented contraction and that the stresses on the flanges were transmitted from the inner flange face to the bore through the solid skin of metal around the casting.

The inner part of the casting, of course, was still molten while the stress transmission was thought to take place. It was reasoned that when the stresses in this envelope of metal became sufficiently large, tearing occurred at the weakest point in the system, which was the hot zone under the flange.

To help the casting resist the stresses imposed upon it by the molding sand, four long webs were cut along the length of the mold between the flanges, extending about 2 in. radially from the outer wall of the barrel. The webs were spaced equally around the mold cavity.

These thin members solidified and cooled rapidly in relation to the remainder of the fitting. Therefore, they had developed considerable strength when the casting was contracting and assisted in counterbalancing any resistance which was offered by the molding sand confined between the flanges. No cracks were formed in the webs, but severe bore cracks were formed as usual.

Another casting was made with ashes in the mold between the flanges to increase the collapsibility of mold; this casting also contained bore cracks.

To positively insure adequate elimination of mold resistance, four molds were made of the fitting with relieving cavities completely encircling the flanges and risers. The patterns for the relieving cavity were molded around the inner face of both flanges so as to leave about ½-in. of sand between them and the flanges. The metal broke through the thin wall of sand separating the mold from the relieving cavity in two of the tests, but in the other two tests the sand sustained the metal pressure.

The sand wall could be seen cracking inward soon after pouring, indicating that no appreciable stresses were built upon the inner flange faces; nevertheless, the fittings contained bore cracks of the same severity as others made with no relieving cavities. From these experiments it was concluded that the

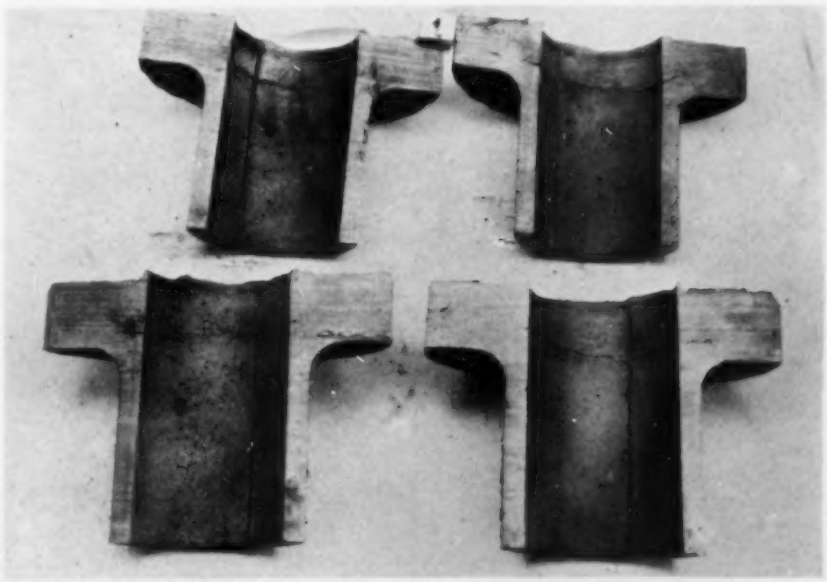


Fig. 7—Flanges of a fitting cast in vertical position. Lower flange in photograph was the bottom flange of casting.

mold itself had no influence on bore cracking and that the cause of the defect must be found in the core.

Brackets. A common method of preventing hot tears in steel castings is to employ chills or brackets at the point in the mold where the tears occur, and it was though that these devices might prove helpful in preventing bore cracks.

Several molds were made for the fitting with the 5-in. bore using cores which had 1½ in. long brackets cut into them at positions under the flanges. The brackets were cut parallel to the axis of the core and were placed about 2 in. apart completely around the core. Examination showed that the bore crack had shifted to the ends of the brackets toward the barrel of the fitting.

When the length of the brackets was increased, small tears were found between the brackets in their normal position under the flange, as seen in Fig. 9. Perhaps more closely spaced brackets would eliminate the defect completely, but the cost of removing them from the bore of the casting would make their use impractical.

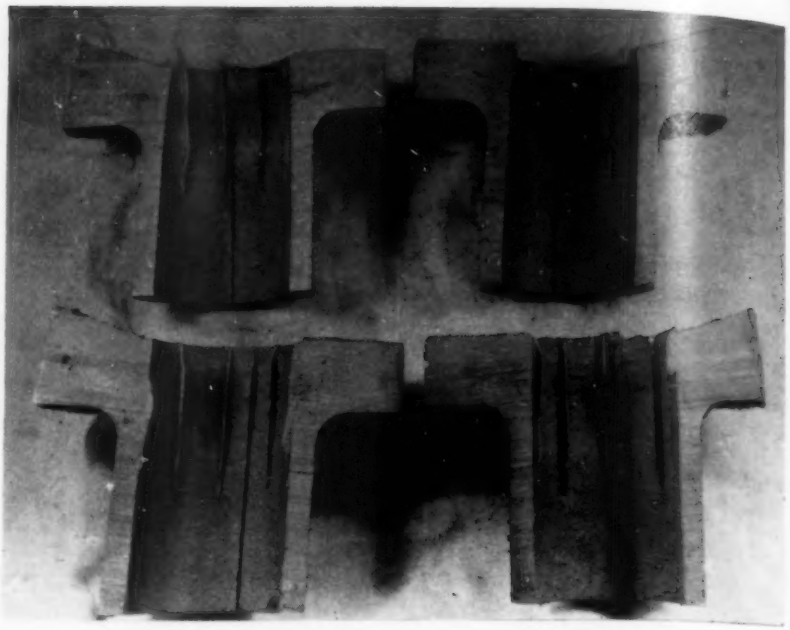


Fig. 9—Fitting made with brackets in core.

Steel Chills. The Dodge Steel Co., Philadelphia, has used steel bar chills molded in the core at positions under the flange to prevent bore cracks. This expedient was tried on the large test casting of Fig. 1, using rectangular bar chills of ½x½x1¼ in. They were placed in the core at the position where the bore crack occurs, spaced about one in. apart and extending completely around the core. This casting was gated into only one flange.

When the casting was examined no evidence of tears under the flange of the ungated end appeared, but under the gated flange it was found that the tear had been shifted from its usual position to the end of the chills into the barrel. When still longer chills were used, the tear continued to move to their ends further into the barrel.

Sometimes the tear would occur at the ends of the chills near the flange face of the casting. In one case a bore crack was found less than ¼-in. from the external face of the flange. It was believed that the reason tears occurred at the ends of the chills was that the transition from the chilled to the unchilled metal was too abrupt, so a number of chills were made (type *A*, Figs. 10 and 11) which tapered off to a feather edge on one end.

These chills were incorporated into the core at the same position as were the rectangular chills, with the feather edge extending into the barrel and the opposite end extending into the core print. The castings were free from bore cracks when these chills were used. The chills need not be placed in the section of the core immediately adjacent to the riser since it has frequently been observed that the tendency for bore crack occurrence in this location is small.

It is desirable to stagger the tapered chills so that the feathered ends do not terminate in the same circle. This practice has not been found necessary to eliminate the bore crack, but helps to make the transition between the chilled and unchilled parts of the casting more gradual and gives added insurance of obtaining the desired results.

However, it has been found abso-

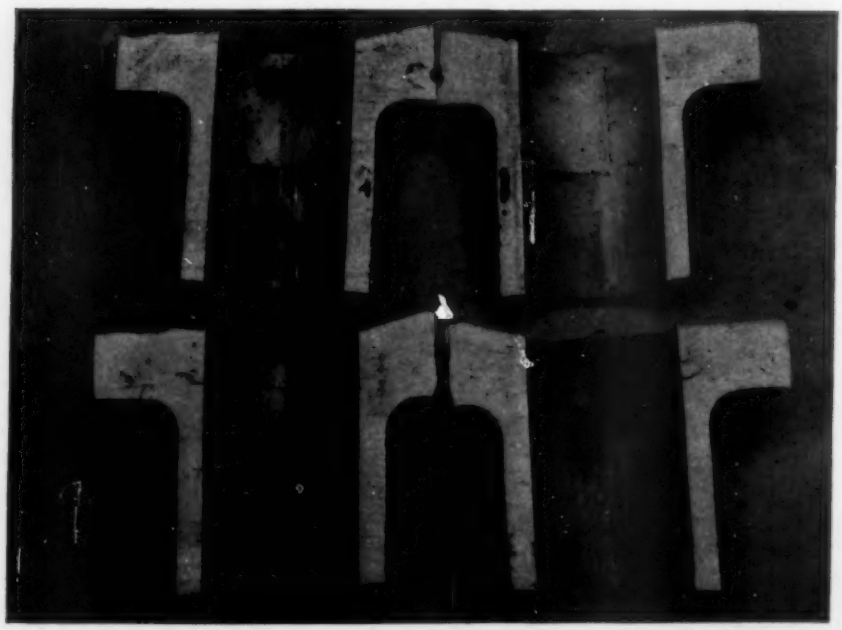


Fig. 8—Two flanges from a test fitting. A pyrex tube was molded under the flange at top of photograph and acted as a mild chill.

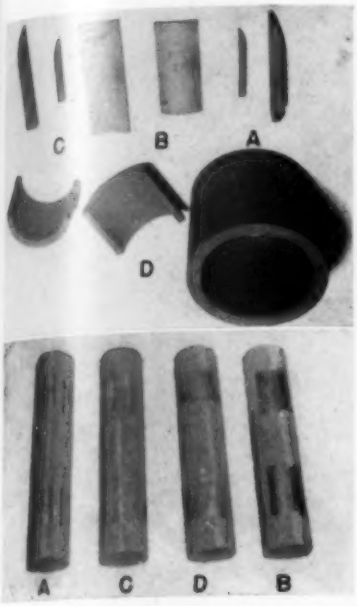


Fig. 10 (top)—Types of chills used to prevent bore cracks: A and B—Steel chills, C and D—Graphite chills. Fig. 11 (bottom)—Cores containing the chills shown in Fig. 10.

lutely necessary that the untapered ends of the chills extend beyond the casting into the core print. This precaution is necessary to preclude the possibility of tears forming at the ends of the chills near the flange face. The tapered ends of the chills should extend into the casting to approximately the line where the fillet from the flange joins the barrel of the fitting.

The smaller test fitting of Fig. 2 can be made without the occurrence of bore cracks by using chills of the same design as those discussed in the foregoing but with cross-sectional dimensions of ¼x¼-in. To determine how many chills were necessary in the core to prevent tears, several molds of the small fitting were made with the gate leading into only one flange.

Varying numbers of chills were placed in the cores under the flange to which the gate was connected, and the opposite end was molded entirely in core sand. Since the most severe bore cracks have always occurred at the gated ends of the test fittings, the chills were tested under adverse conditions. Ten chills, six in the drag and four in the cope, were sufficient to prevent tearing in this particular fitting, using the core sand previously described.

However, a severe bore crack, extending completely around the casting, occurred under the opposite flange where no chills were located. Figure 12 shows the two flanges of the test fitting after they were subjected to magnetic powder testing.

Another type of chill used in the cores was the thin chill shown in *B* of Figs. 10 and 11. These chills were 4 in. long, 1¾ in. wide, and only 1/16-in. thick, and were made to conform to the contour of the bore of the fitting. They were molded side by side in the core with nails through the holes for anchorage, and they extended around the core except for the position under the riser.

As in the case when bar chills are used, the ends of these chills should extend into the core print. For the small test fitting, three chills in the drag and two in the cope of the core prevented bore cracks. No danger of developing longitudinal tears in the fitting exists when using these chills since they are thin and sufficiently narrow to yield independently as contraction takes place.

A disadvantage in using these chills is that of frequent fusing to the casting, even when coated with a refractory wash, and it often is necessary to remove them by grinding or chipping. However, the added effort of removing the chills is certainly less than would be required to remove a bore crack and repair it by welding.

A chill which is only 1/16-in. thick, as these are, has only a slight chilling effect since a section of this size quickly reaches the temperature of the metal in the casting. However, a thin chill strengthens the metal skin sufficiently to enable it to resist the tearing stresses.

Graphite Chills. Chill studies at the Naval Research Laboratory have

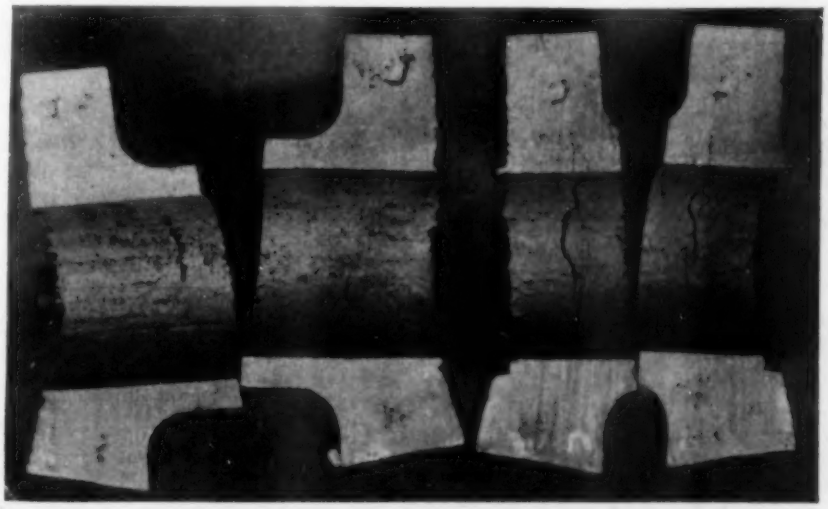


Fig. 12—Flanges from a test fitting. Steel bar chills were molded in core beneath flange on the left.

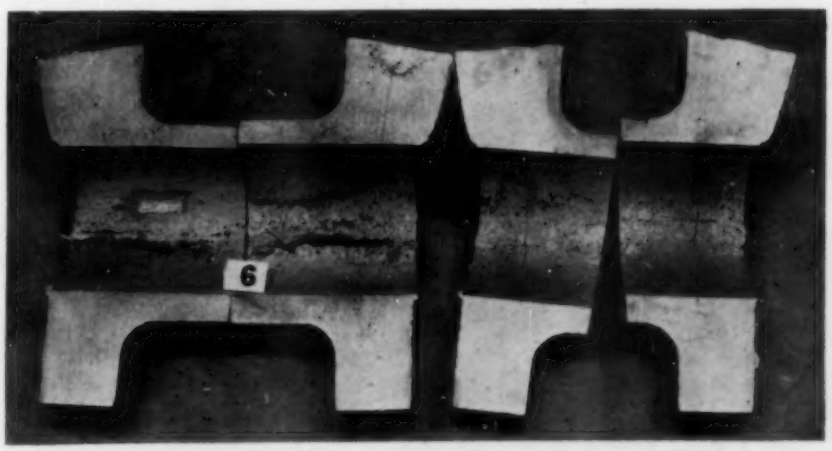


Fig. 13—Fitting made as in Fig. 12, but with graphite chills in the core.

shown that, on a volume basis, graphite has approximately two-thirds of the chilling effect of steel. Since a mild chill is desirable for use in preventing bore cracks, it was thought that the graphite chills could be used to advantage. The chills were used in two forms as shown in *C* and *D* of Figs. 10 and 11.

One type of chill was the rectangular stick ½x½ in. in cross section and 2½ in. long; the second type was the circular graphite collar. The sticks were molded in the core in the same manner as were the steel chills. The graphite collars were molded in the core, one-half in the cope section and one-half in the drag section so that, when the core was pasted together, the graphite collar completely encircled the core.

A series of small test fittings was poured to determine the value of graphite chills for preventing bore cracks. The fittings were gated into one flange and fed by open risers. Some of the cores contained varying numbers of graphite sticks and others contained the collars, which varied in thickness from ⅛ to ⅞-in.

It was found that for this particular fitting, five of the ½x½x2½-in. sticks in the drag and four in the cope were sufficient to prevent bore cracks. None of the fittings with the graphite collar chills had the defect.

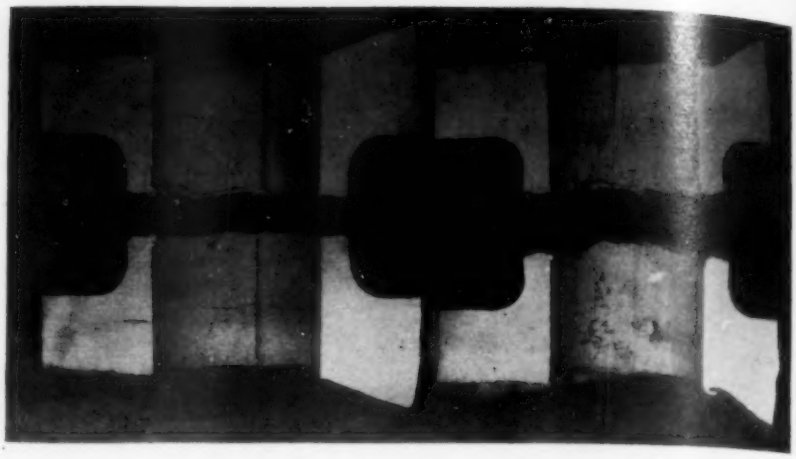


Fig. 15—Two flanges from a fitting. Flange on right chilled with graphite collar, defective flange unchilled.

However, when the molten metal came in contact with the graphite chills it became locally enriched with carbon unless a heavy refractory wash was applied to the chill. Unless such a wash was applied to the chill, the metal adjacent to it absorbed carbon in quantities usually sufficient to give the metal surface an appearance of gray iron under microscopic examination.

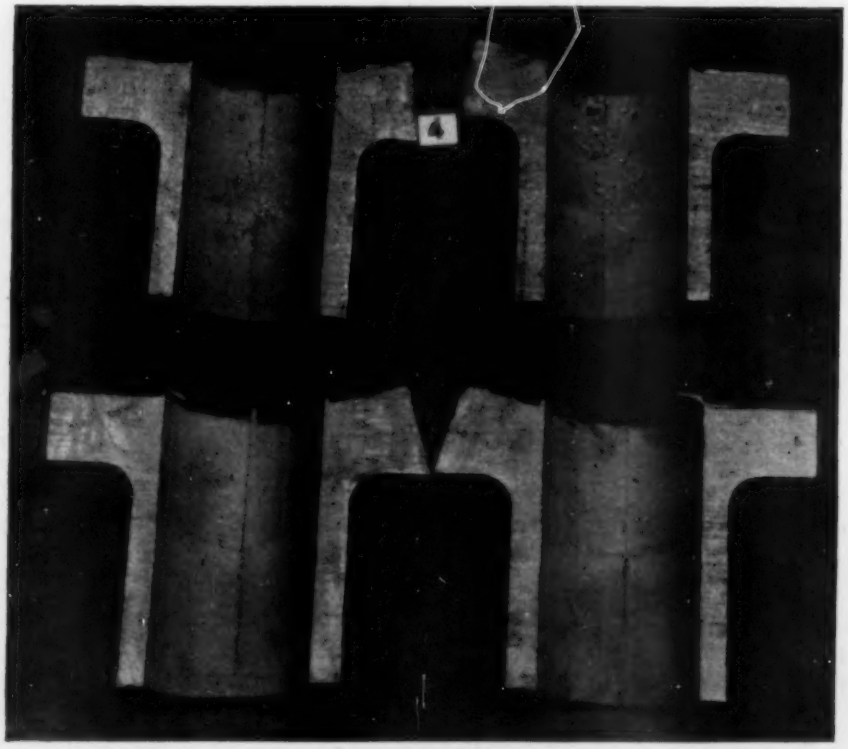


Fig. 14—Fitting flanges cast with graphite collars in core.

Although a graphite collar of ⅛-in. thickness has sufficient chilling effect to prevent tearing, it leaves a poor surface on the casting. Increasing the chill thickness improves the surface appearance and reduces the tendency of the metal to absorb carbon. Apparently, the thin chills become too hot and a gas is evolved from them which causes a poor surface.

For the small test fitting used in these experiments it was found that, when the chill thickness was ½-in. or more, the surface next to the chill was as good as that with sand. Graphite is expensive and a graphite chill can be used only once whereas metal chills can be used repeatedly so, for reasons of economy, metal chills are preferable.

Figure 13 is a photograph of the flanges of a fitting in which one flange was chilled with graphite sticks and the other flange was unchilled. The graphite sticks act in the same manner as do the steel bar chills, although with less chilling action. The magnetic powder indicates the bore crack in the unchilled end of the casting.

Graphite Collars

Figures 14 and 15 are photographs of flanges cut from a fitting made with graphite collars in the core. The casting in Fig. 14 was made with graphite collars under both flanges, while the casting in Fig. 15 was made with a collar under only one flange. This photograph illustrates the effectiveness of the graphite in preventing bore cracks since the chilled flange is completely free from bore cracks while the opposite

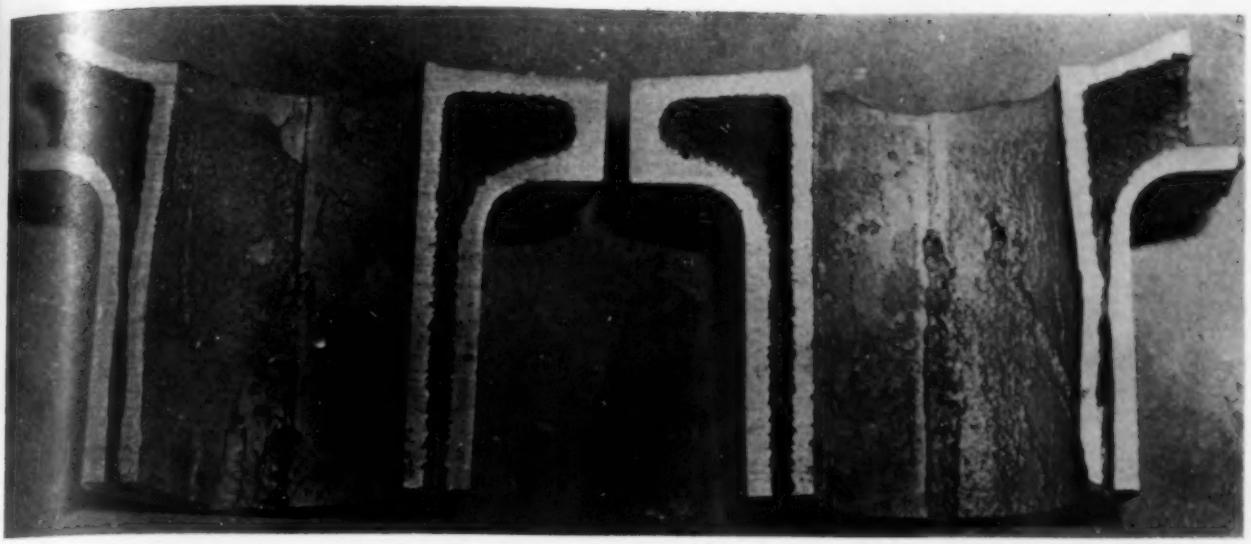


Fig. 16—Casting bled 15 sec. after pouring.

flange without the collar exhibits a severe case of bore cracking.

Many foundrymen who have discussed this problem with the authors are wary of chills as a remedy for bore cracks. One objection is the belief that the chills will choke off feeding to the barrel of the casting. No abnormal shrinkage has been detected in the barrels of the test castings made with chills in the core.

Some centerline shrinkage was present, but it is a safe prediction that 90 per cent of all fittings and valves have centerline shrinkage. Brinson and Duma[9] have shown clearly that feeding can take place over only a short distance in thin sections such as are usually found in valve and fitting bodies. For example, a ½-in. section can be completely fed for a distance of only one in.

Also, since no chills are needed in the part of the core beneath the riser, feed metal is readily accessible to the body of the casting through this area. Any material that acts as a mild chill and withstands the temperature of molten steel can be used to prevent bore cracking.

Figure 8, for example, is a photograph of the flanges of a casting made with a pyrex tube, having an outer diameter equal to the bore diameter, molded into the core so that when the mold was assembled the tube was under one of the flanges. The upper flange, in this photograph, which shows no tear, is the one under which the pyrex was used as a chill. The other flange of the casting had only core sand beneath it and was severely cracked.

Although the proper use of chills was found to eliminate bore cracking, the fundamental cause of the defect remained obscure. Also it was not clear how and when bore cracks formed. It was felt that simpler remedies for bore cracking could be found if more information as to its cause was available. In an effort to obtain such information, a series of bleeding experiments was undertaken.

Bleeding Experiments. "Bleeding" of castings is accomplished by inverting the molds after they have been poured so that still molten metal can run out through the risers. "Slush casting" is a term also

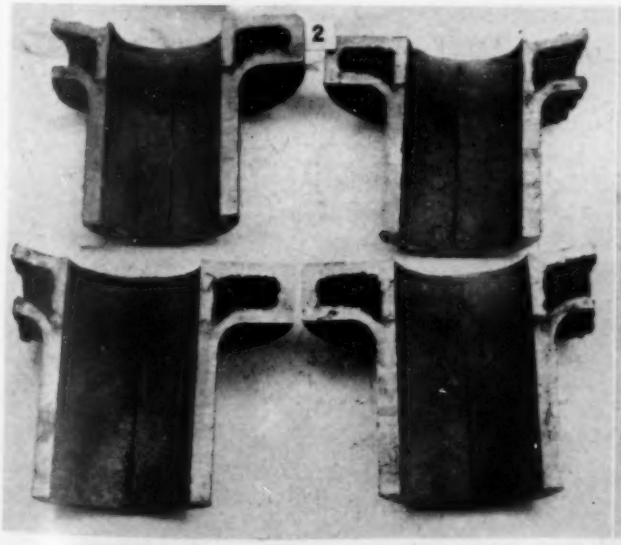


Fig. 17—Casting bled 25 sec. after pouring.

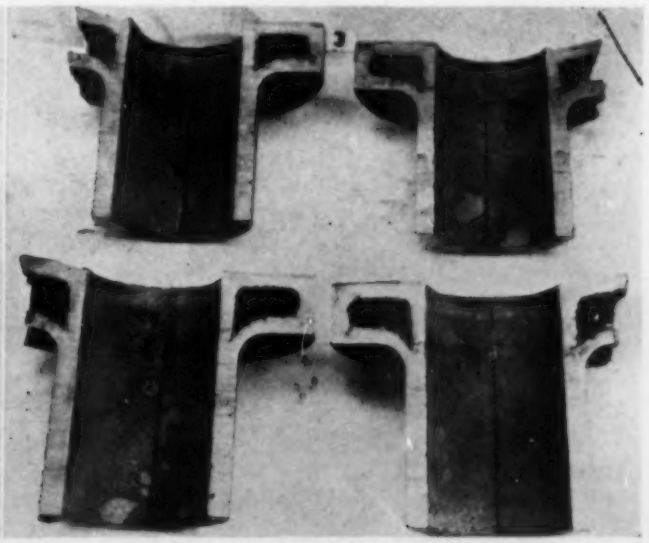


Fig. 18—Casting bled 35 sec. after pouring.

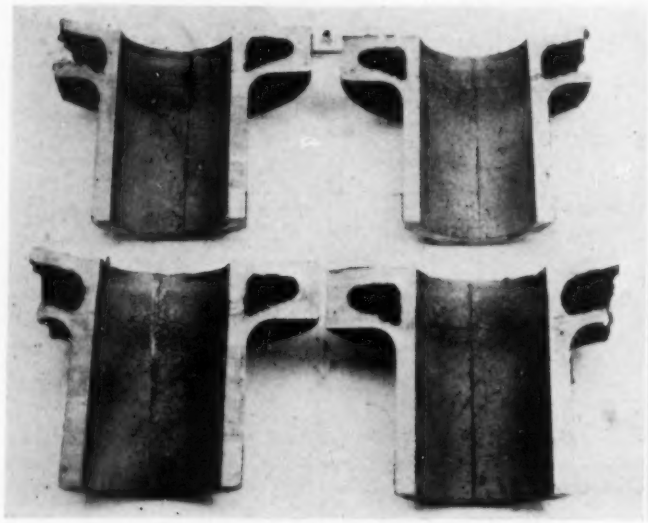


Fig. 19—Casting bled 45 sec. after pouring.

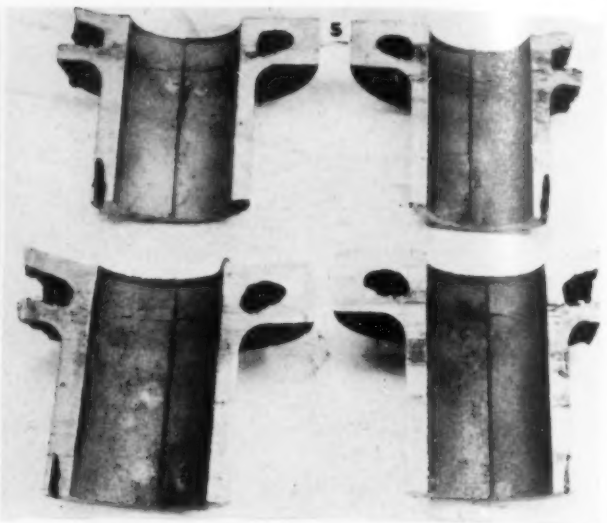


Fig. 20—Casting bled 60 sec. after pouring.

applied to processes involving the bleeding of castings. By bleeding the molds at various times after pouring, different skin thicknesses were obtained in the castings and these shells of metal were examined for bore cracks.

The castings were gated into the barrel and had sand cores in the bottom of the flanges to permit air to enter during bleeding, and thus prevent collapse of the walls. Bleeding tests were made on both of the test fittings. Pouring temperatures of approximately 2950° F. (1620° C.) were used.

It was found that in the large fitting the bore crack did not begin to form until about 60 sec. after pouring was completed. In the smaller fitting, bore cracking began 35 sec. after pouring. Photographs of the flanges of the small fittings after they were bled at various stages of solidification are shown in Figs. 16 to 20, inclusive.

In all of the bleeding tests it was found that bore cracking did not begin until after the barrel of the casting had completely solidified. Bore cracking starts earlier in the smaller fitting than it does in the larger fitting since the barrel is thinner and solidifies in a shorter time than does the barrel of the larger fitting. In the bled casting of Fig. 16, the bore has not completely solidified and no cracking is present.

In the casting of Fig. 17, bled after 25 sec., the bore has just solidified, but no cracks are present. The reason for the absence of cracks, one that will be considered later as to mode of action in connection with a discussion of stresses, is that no hot zone is present under the flanges at this stage of the solidification.

The magnetic powder on the sawed part of the barrel of the casting near the inner faces of the flanges (Fig. 17) should not be construed as an indication of a defect; the prods of the testing machine were placed there and attracted some of the powder. In these first photographs the skin thickness, with the exception of that of the external corners and the mild internal corner made by the fillet at the junction of the flange and barrel, is uniform.

Skin Thickness

No indication of the skin thickness being less beneath the flange is shown, which would be the case if there were a hot spot at this point. After 35 sec. the bore crack had started to form, as can be seen from the photograph of Fig. 18. Figures 19 and 20 are castings bled at later intervals after pouring and show how the bore crack grows as solidification proceeds.

Stress Factor in Bore Cracking. External corners in castings always cool and solidify at a more rapid rate than do other sections. This is due to the fact that the greatest volume of sand, which can conduct heat away from the casting, is around an external corner. An external 90° corner has three times as much sand around it than does an internal 90° corner. Greater thickness of metal at the external corners of the bled castings of Figs. 16 to 20 is apparent.

Since external corners cool more rapidly, they also begin to contract earlier than the rest of the casting. In the fitting, the external corner around the casting made by the outer flange face and the bore contracts earlier than the rest of the casting and, in contracting, pinches down on the core as shown in *B* of Fig. 21.

Further, Briggs and Gezelius[5] have determined that, if pouring temperatures and molding materials are constant, the amount of metal solidified on a mold surface at a given time after pouring is always the same irrespective of the size of the casting.

While this theory may not be correct for all cases, it holds for the section sizes and time intervals of solidification with which this discussion is concerned. In other words, at some time after pouring, a one-in. cube should have the same skin thickness as would be found on a 3-in. cube. A pressure fitting consists of a barrel which is relatively thin and a flange which is approximately three times as thick as the barrel.

In accordance with the theory of skin formation set forth by Briggs and Gezelius, the skin thickness along the bore of the test fitting should always be the same under the barrel as it is under the flange until the barrel is completely solidified. The heat extracted by the core during solidification of the barrel is largely heat of fusion.

Since the skin thickness of the casting is uniform, no hot zone is

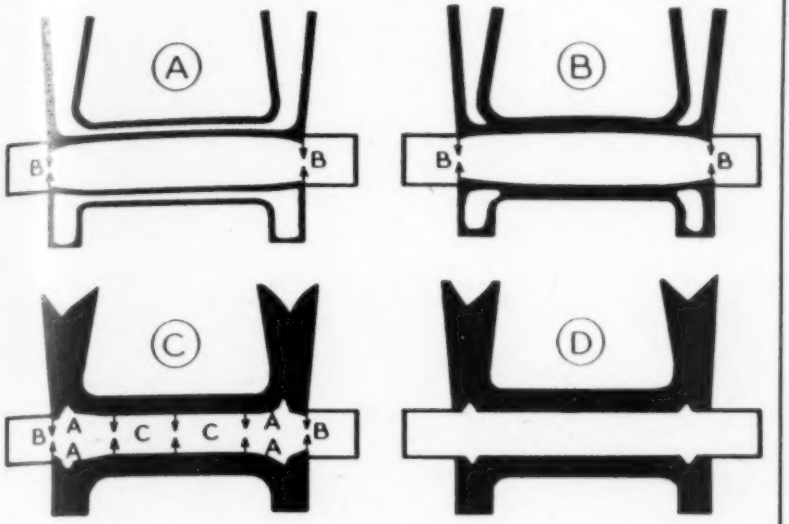


Fig. 21—Diagrammatic representation of bore crack formation. A—Metal skin around casting shortly after pouring. External corners made by junction of flange face and bore of casting pinching on core at "B." No hot zone present. B—Metal skin around casting at the time barrel has just completed solidification. No hot zone present, but as solidification progresses from this point the formation of hot zones and bore cracks begins. C—Flanges have just completed solidification and bore cracks are formed. During this time the barrel has been cooling and contacting onto the core at "C," area at "A" has been dissipating heat of fusion from the flanges with no appreciable temperature drop and this area becomes the hot zone. Linear contraction has been restrained at "B" and "C" causing bore cracks to form at "A." D—Casting cooled to room temperature. Flanges in cooling have also shrunk onto core.

present in the wall of the bore or anywhere else in the casting. However, after the barrel of the casting has solidified, further extraction of heat from the barrel by the core serves entirely to cool the barrel. At the same time, the heat extracted from the flange by the core is still largely heat of fusion lost in further solidification of the flanges.

Since heat of fusion is lost without a lowering of temperature, the flange does not cool as rapidly as the barrel and a hot zone develops. When the barrel of the fitting begins to cool at a more rapid rate than the flange it also contracts circumferentially at a faster rate and, in so doing, clamps onto the core with a greater force than does the flange. These contraction forces are illustrated in exaggerated form in Fig. 21.

Bore cracking then would not be expected to occur until the barrel is completely solidified because, although the corners made by the bore and the flange face do pinch on the core, no hot zone is present, and before conditions are favorable for tearing a hot zone must exist on the surface of the casting.

That the bore crack does not occur until the barrel of the casting has solidified has been proved by the bleeding experiments discussed earlier. After the barrel has solidified, longitudinal contraction is restrained by the pinching action of the barrel and by the external corner of the flange at points *B* and *C* in Fig. 21, and tearing occurs.

The casting shown in Fig. 18, bled after 35 sec., has a small bore crack. In this casting the hot zone under the flange is coming into existence. The barrel at this point is cooling more rapidly than the flange and is pinching on the core. Also, at this time, longitudinal contraction is exerting its forces along the barrel of the casting. The barrel of the casting and the corner at the flange face are clamping onto the core and resisting this contraction, and the casting begins to tear.

Both the longitudinal contraction forces and the circumferential contraction forces build up as the casting cools and, after 45 sec., as shown in Fig. 19, the bore crack becomes more severe. The crack continues to open up as the casting completes its solidification.

The circumferential contraction of the flange corner and the barrel is slight, and the force with which it binds on the bore is minute, but when it is remembered that at the high temperature at which tearing occurs the tensile properties of steel are extremely low, it is reasonable to believe that these stresses are sufficient to cause tearing.

In valves and fittings, each flange and a short part of the barrel adjoining it is a single entity as regards bore cracking, and it does not matter whether the barrel attached to the flange is 10 ft. long or 6 in. long. A mold of a test fitting was made and one flange was blocked off with sand so that the casting consisted of a barrel with a flange on one end. Severe bore cracks were found under this flange.

Varied Length Barrels

A series of castings was also made, each casting consisting of a flange with a barrel extending from each side. The barrels on these castings varied in length from 8 in. to only one in., and in every case bore cracks were found under the flange. Figures 22 and 23 are photographs of the casting with the 6-in. barrel, with the magnetic powder indicating the crack in the bore.

Figure 24 shows the casting with the one in. long barrel as sectioned and tested with magnetic powder. This photograph, which shows the entire casting with only the gate and riser removed, illustrates the susceptibility of a flange design to bore cracking.

Figure 25 shows a casting which had a flange equal in thickness to the barrel, had only a minor bore crack because the flange and the barrel completed solidification at the same time and the only hot zone which existed was a small one made by the junction of the two sections. The pinching action discussed occurs at the extremities of any casting in which a core is surrounded by metal because, with rare exceptions, the casting must end with an external corner.

The one condition which must exist in such a casting before tearing can occur is a hot area around the core. A long cast cylinder is not

susceptible to circumferential tears because no hot area is present.

A cylinder with a boss on it, although the boss does form a hot spot, may not tear because the remainder of the cylinder in the same plane as the boss is strong enough to resist the tearing stresses. If the boss, or it may be a flange or other protuberance, extends all or a majority of the way around the casting, the conditions become favorable for tearing.

If the external corner of metal which pinches on the core could be partially eliminated, the severity of the bore crack should be correspondingly reduced, since then the casting would be free to contract toward its center. A casting, designed so that the bore flared out at the flanges, was made to test this theory. The flanges of one of these castings is shown in Fig. 26 and, as can be seen from the magnetic powder indications, the tears are not as severe as those found in straight bores.

By flaring the bore, as was done in this case, instead of a 90° corner in the casting with a 270° corner of sand to conduct heat away from it, the metal corner of about 120° has only a 240° corner of sand to remove heat.

Several castings of only the cope half of the test fitting were made, and not the slightest indication of a bore crack could be found. In this case, although early contraction occurred at the external corners and the barrel, the casting could deform slightly in the manner of a "C-spring" so that no actual pinching occurred.



Fig. 24—Single flange casting with one inch long barrel on each end. Photograph shows entire casting as sectioned, magnetic powder indicating crack.

A test fitting was made in which one flange had a ½-in. slot cored in it from the bore to the outer circumference of the flange to make the flange discontinuous. The surfaces marked *A* in Fig. 27 are the cast surfaces of the slot. This flange was found to be free from tears, whereas the other unslotted flange had bore cracks.

The explanation for this is that, when contraction of the external corner of the slotted flange occurred,



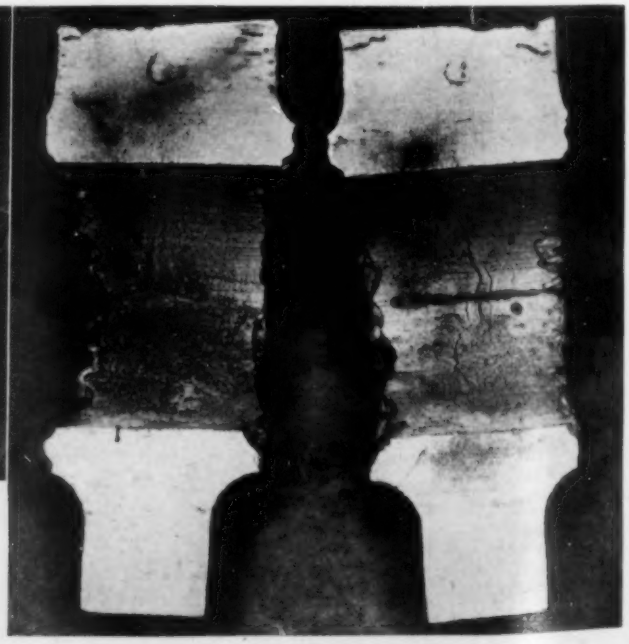


Fig. 22 (above)—Single flange casting as removed from mold. Fig. 23 (right)—Flange of casting in Fig. 22 as sectioned and tested with magnetic powder.

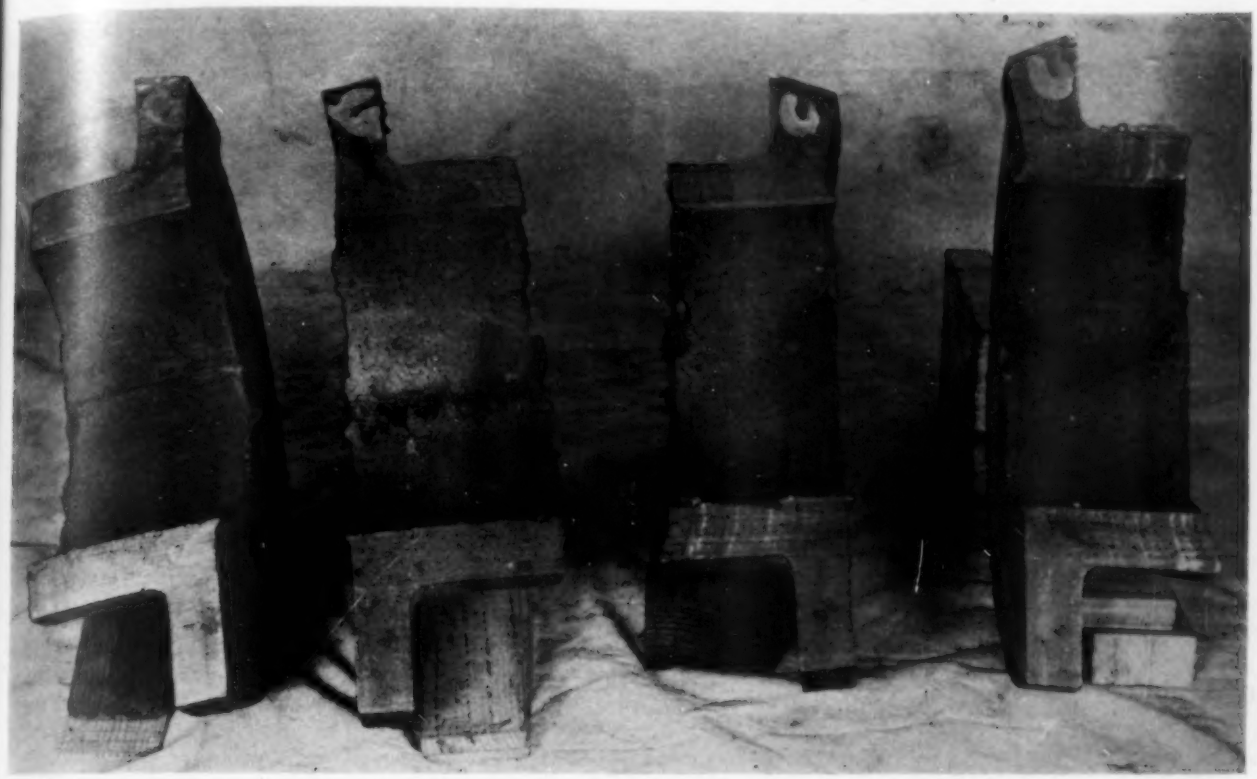


Fig. 25—Sections of a fitting designed with barrel and flanges of equal thickness. Magnetic powder indicates small crack on section at the left.

the flange was able to spring apart slightly at the slot and creep around the core, whereas the opposite flange, being continuous, could shrink only onto the core.

The theory of stresses explains why the severity and frequency of bore cracking increase directly as functions of the bore diameter of the casting. As the circumference of the bore increases the amount of circumferential contraction also increases and, because of the increased circumferential contraction, the flanges and barrels must pinch deeper and more firmly into the core.

Chills prevent bore cracking because they tend to eliminate the hot, weak area under the flange. Another solution to this problem, which is apparent if stresses are accepted as the cause of the bore crack, is the use of highly collapsible cores.

Hollow Cores. Many proprietary core binders and many compositions were used in making cores for the

Fig. 26—Sections of fitting designed with a flaring bore to reduce stresses on the core.



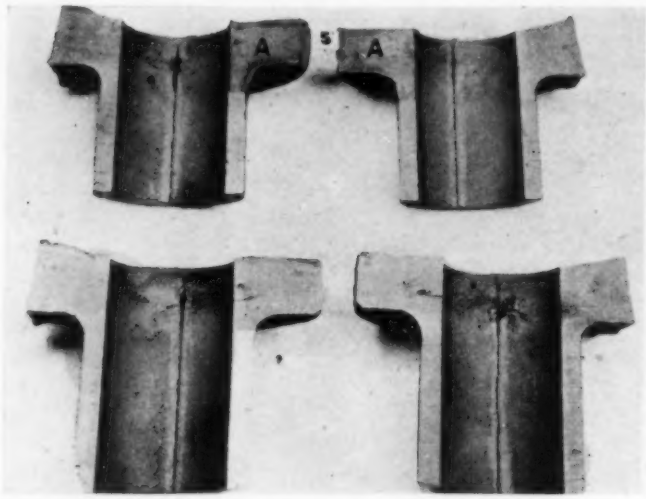


Fig. 27—Sections of a fitting made with one flange discontinuous to reduce stress (top flange in photograph). Surfaces marked "A" are cast surfaces.

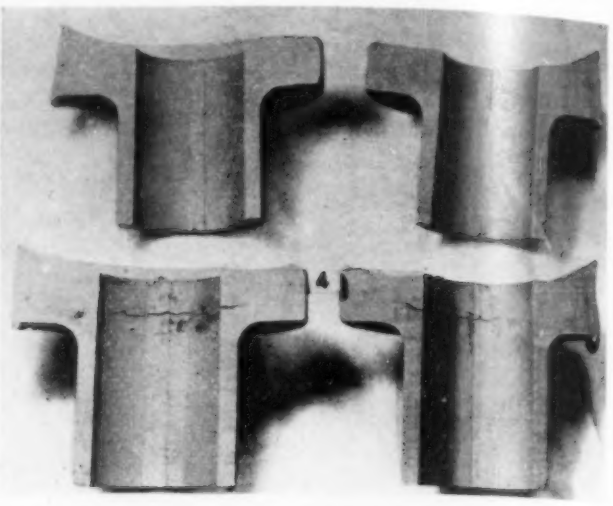


Fig. 28—Flanges of a casting made using a core with one end hollow and the other end solid. The lower flange containing bore crack was around the solid end.

test fitting in an endeavor to find a mixture that could be used in producing castings free from bore cracks. Green sand cores and cores with low hot strength did not produce the desired results. It became apparent that to obtain the desired results, the core must be made of such materials or designed in such a manner that it collapses readily, and that space must be provided in which the core can collapse.

In other words, a solid core occupies a certain volume and, even though it becomes weak, it will still occupy the same space and resist any tendency to compress it. The binders in the core will eventually burn out to reduce the volume needed for the core, but this does not take place early enough after pouring to prevent bore cracking.

Lucite, which depolymerizes above 400° F. (200° C.) and burns very rapidly, was tried as a core binder, but the core decomposed as soon as the hot metal touched it. To circumvent this difficulty, a special core was made into which a cylindrical shell of lucite of ¼-in. thickness was rammed so that its outer surface was approximately ¼-in. beneath the external surface of the sand.

When this core was used a casting free of bore cracks was obtained. The lucite burns out by the time cracking begins and leaves a void into which the outer sand layer can collapse. These considerations led to the use of hollow cores. When the pressure of the contracting casting is exerted on a hollow core, the core yields inwardly toward its center and relieves the stresses imposed upon it.

Several of the small test fittings were made with hollow cores of different wall thicknesses. The cores were made of the mixture containing 0.5 per cent cereal binders, 0.5 per cent kerosene, 1.5 per cent core oil, and 5 per cent water, and the wall thicknesses of the cores varied from ¼ to 1 in. It was found that when the wall thickness of the core was between ¼ and ⅜-in., the castings, which were poured at about 2950° F. (1620° C.), were free of bore cracks.

When the wall thickness was greater than ⅜-in., bore cracks were present. A core of this composition with a ½-in. wall or a wall of greater thickness does not break down soon enough to relieve the stresses which produce bore cracks.

However, when the larger test casting with the 5 in. diameter bore was made, a core with a ⅝-in. wall collapsed rapidly enough to prevent the formation of bore cracks. A core with a ⅜-in. wall in the large casting collapsed too rapidly, and molten

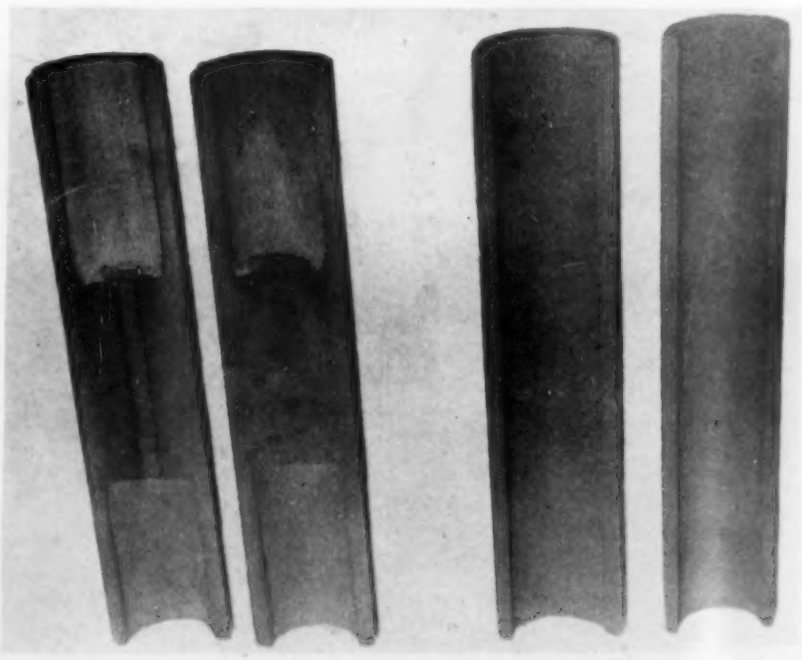


Fig. 29—Hollow cores used in making test fittings (Fig. 28).

metal broke through the wall and entered the core.

A correct wall thickness of the core must be established for each core composition. For example, cores made of the composition described and with a ¼-in. wall thickness prevented bore cracking in the test casting and still were strong enough to withstand the metal pressure, whereas a core made with 1.0 per cent cereal binder, 1.0 per cent core oil and 4.0 per cent water, with a wall thickness of ⅜-in. did not withstand the metal pressure and broke. With a ½-in. wall, this core composition worked satisfactorily.

Procedure for using hollow cores must be worked out in each foundry. Core composition, pouring temperature, and the size of the casting will determine the correct wall thickness of the core. In a commercial foundry the pouring temperature will be fairly constant, and a few tests can be made on the core sand being used to determine the necessary wall thickness for each size casting.

Cores can be rammed to size with a suitable pattern, or they can be made hollow with a sweep. Both methods have been found satisfactory. The cores need only be hollow in the parts located under the flanges, but for economy a completely hollow core is recommended. Figure 29 shows some hollow cores before they were pasted together.

Figure 28 clearly illustrates the advantage gained by the use of hollow cores. This casting was made with half of the core hollow and the other half solid and gated into each flange. The flange around the hollow end was free from tears, while the opposite flange had a bore crack which extended almost completely around the casting.

To make possible the observation of the inside of a hollow core during the casting and solidification of a test fitting, the core print of the pattern was extended to the side of the molding flask, and a hole was cut into the flask at this point equal in diameter to the core. In those experiments the fitting with the 3 5/16 in. diameter bore was used.

Several molds were poured with this arrangement. In each case the core burned violently soon after pouring and, about 30 sec. after pouring was completed, metal could be seen bleeding slowly through the core walls into the hollow center of the core. This failure of the core always occurs at the lowest part of the mold where the greatest metal pressure exists.

Preventing Core Burning

Since, under the conditions for observation described, the core interior is open to the atmosphere, the core burns and breaks down much faster than it normally would when it is enclosed in a mold. In subsequent tests, a glass window was placed between the end of the core print and the molding flask in order to prevent oxygen from entering the core and yet permit observation of the core interior. The cores then satisfactorily resisted the pressure of the metal.

With the window in the mold, no burning occurred in the core although smoke was given off and the core soon became red hot. No actual cracking or deformation of the core could be detected by the eye during solidification of the casting, the deformation being too small to be observed. A few minutes after pouring was completed the core had completely decomposed, and jarring of the mold caused the core to flow like a dry, unbonded sand.

To prove that the beneficial effects of the hollow core were ascribable to its ability to deform rather than to an insulation effect, a core for the test fitting was made, one end of which was hollow with ⅜-in. wall thickness, while the other end had a tin can molded into it. The can was ½-in. smaller in diameter than the core itself so that ¼-in. of sand was between the can and the casting.

It was assumed that the can would resist the contraction of the casting, but because of its thin walls would not appreciably alter the thermal conditions which exist in the hollow core, so that if the bore crack was prevented by an insulation effect, none should occur in the end of the fitting containing the can.

This casting was gated into the center of the barrel to make the metal conditions in each flange equal. The flange under which the can was located had a bore crack, as shown at the top of Fig. 30, while the other end was sound.

It appears that if any insulating effect due to the hollow core is present, it does not exist during the early stages of solidification when the bore crack begins to form. Figure 31 is a photograph of a casting

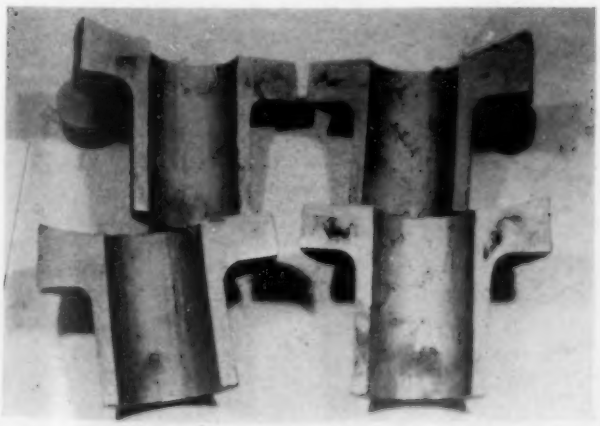


Fig. 30—Flanges of a fitting made with a core having a tin can bedded ¼-in. below its surface on one end while the other end was hollow. The sound flange was around the hollow end.

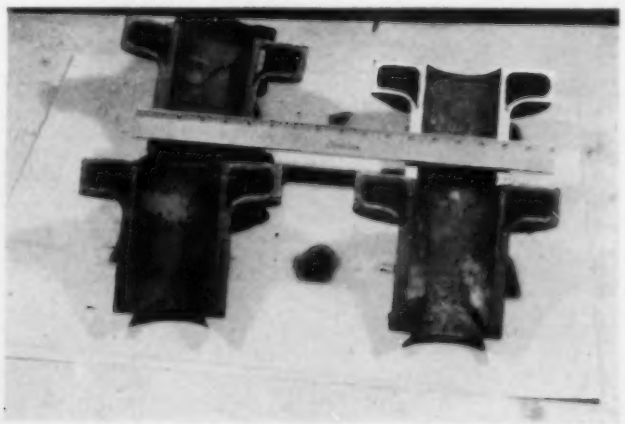


Fig. 31—Bled casting containing a half hollow core. The skin thickness on both ends of the casting is the same.

bled 33 sec. after pouring. This casting contained a core, one end of which was hollow and the other end of which was solid. The casting was gated into the center of the barrel.

The top flange (Fig. 31) was around the solid portion of the core, and the bottom flange was around the hollow end. Both flanges have the same wall thickness at the barrel, which would not be the case if there had been any difference in the heat conductivity of the two ends of the core.

Shrinkage as a Factor in Bore Cracking. In making test fittings in the course of these experiments, it has been frequently noted that the surfaces in the bores of the castings beneath the flanges have a pitted and somewhat spongy appearance. Because surfaces such as these in some non-ferrous alloys are associated with shrinkage, it was believed that this condition might be a form of shrinkage and be contributory to bore cracking.

Fittings used throughout these tests were gated at the parting line and fed by risers on top of the flanges. Although they were risered sufficiently to yield commercially sound castings, they were not perfectly fed. Accordingly, several castings were made with large risers in order to approach the ultimate in feeding. This risering method consisted of placing two heavy risers on each flange, as shown in Fig. 32.

Metal was gated directly into all four risers to insure proper temperature gradients. These castings, one of which is shown in Fig. 33, had small cracks in the cope which were barely detectable, and the surface condition was improved. However, all of the improvement in these castings in regard to the bore crack should not be attributed to the fact that they were more heavily risered.

Risers were placed on the casting in such a manner that the external corner, made by the bore and the fact of the flange, was eliminated at the points of contact. With two risers on each flange the riser contacts cover about one-third of the flange and, by thus partially eliminating the corners, cause a proportional decrease in the contraction upon the core.

As previously described, hollow cores do prevent bore cracks, but when the fitting is made without risers, although a hollow core is used, tears are present. Figure 34 shows the flanges of a test fitting made with a hollow core but without risers. This photograph indi-

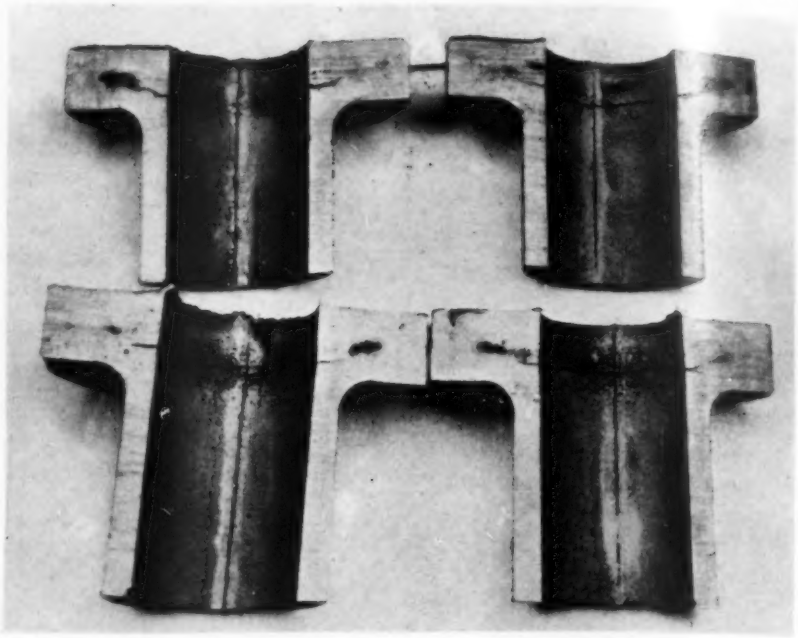


Fig. 34—Fitting made without risers and using a hollow core.



Fig. 32 (above)—Casting gated and risered to obtain complete soundness.

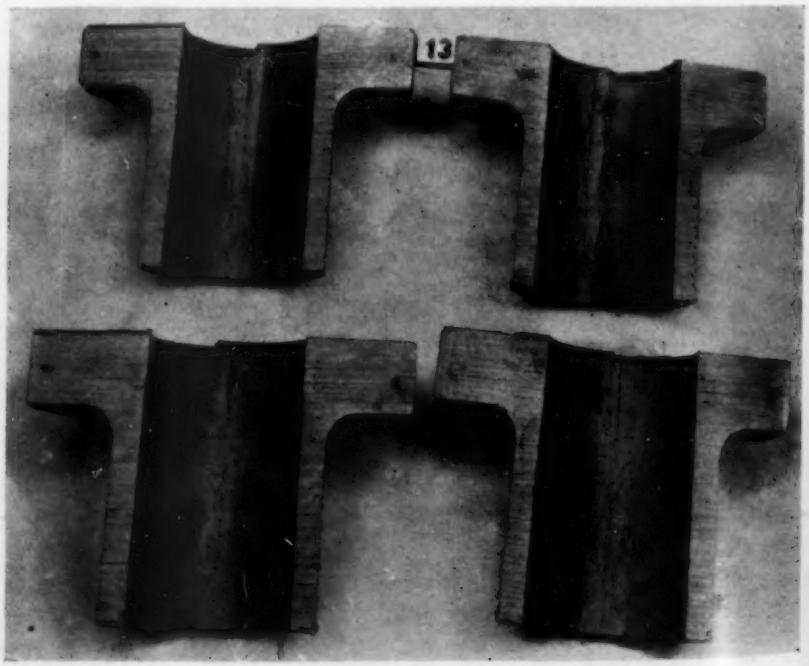


Fig. 33 (right)—Flange sections from casting of Fig. 32 after testing with magnetic powder.

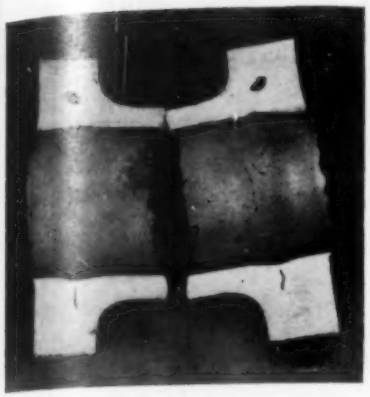


Fig. 35—"Dishing" of bore beneath the flange due to inadequate risers and atmospheric pressure.

cates that incomplete feeding is a factor in causing or at least aggravating bore cracks.

It has been noted that if the riser fails to function properly, dishing may occur in the bore beneath the flanges, and if a shrinkage cavity is present within the flange, the crack invariably will lead to it. Figure 35 shows an exceptionally pronounced dishing effect in an improperly fed flange.

Detrimental effects of inadequate risering can be blamed upon atmospheric pressure. If the riser freezes before feeding is complete, a shrinkage cavity which is a vacuum begins to form. Atmospheric pressure, acting on the external surfaces of the casting with a force of 14.7 psi., tries to relieve this vacuum within the casting.

The wall of the casting under the flange, since it is the hottest and weakest area, will be forced inward in much the same manner as the walls of a tin can collapse when it is evacuated. The stresses induced in the wall of the casting by this localized distortion could conceivably cause the wall to rupture and permit access of air to the shrinkage cavity.

If a bore crack has already started when the shrinkage cavity begins to form, the force of atmospheric pressure, in trying to break through to the shrinkage void, will extend the crack to the shrinkage cavity, and also enlarge it. Figure 36 is a photomicrograph of a bore crack. The metal adjacent to the tear is decarburized indicating that oxygen from the atmosphere has been present in the crack.

In a further investigation of the effect of atmospheric pressure on bore cracking, a mold of the test fitting was made with the gate attached at the midlength of the barrel to equalize the temperature of the metal in the flanges. This casting was made without risers. However, in one flange, cracker cores, similar to those used in blind risers to permit access of the atmosphere to the riser, were inserted into the flange at the fillet.

The sectioned casting is shown in Fig. 37, and the void left by one of the cores can be seen extending to the shrinkage cavity. The flange containing the cores was not torn as severely as was the opposite flange, since the cores afforded easy passage of air to equalize the pressure between the inside and outside of the casting.

Veins and Their Relation to Bore Cracking. Veins are ragged encrustations occurring in irregular lines on the surfaces of castings. Their origin usually is ascribed to the occurrence of cracks in the core or sand of the mold which subsequently are filled with molten metal to form the veins. Veins frequently have been noted in the bores of valves and fittings under the flanges and, invariably,

Fig. 36—Photomicrograph of a bore crack. 500X.



when the vein is ground away and the area subjected to a magnetic powder test, a bore crack is found.

Figure 38 shows a section of a fitting with a vein in the bore. Figure 39 shows the same section after the vein has been ground away and the part subjected to magnetic powder inspection. Since the bore crack follows the contour of the vein so accurately, it is a certainty that they are related. Singer and Benek[10] have supplied the answer to the relationship between hot tearing and veining.

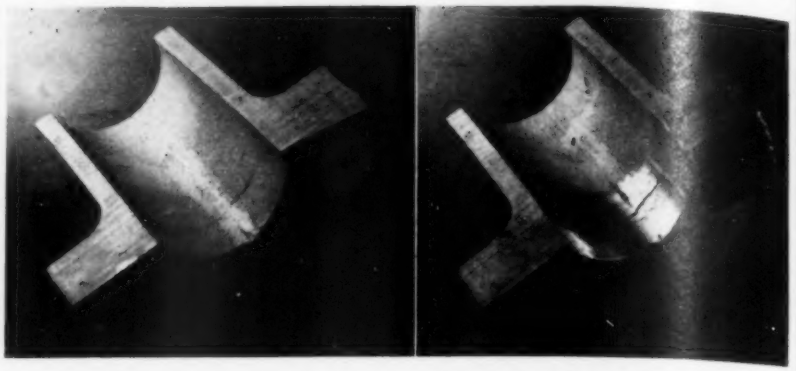


Fig. 38 (left)—Vein in the bore of a fitting. Fig. 39 (right)—Bore crack in the same casting after removal of vein.

They concluded from their work that when the frozen skin of the casting tears, the mold wall adjacent to and adhering to this metal skin is also torn. Molten metal from the interior of the casting flows through the crack in the metal skin and into the crack in the mold, thus forming a vein. They found, by metallographic examination of metal sections cut through a vein, that the metal leading to it was richer in carbon than the adjoining metal.

Sulphur prints indicated that the original crack could be traced by the phosphorous and sulphur-rich metal which had filled it. Since the last metal to solidify in a casting is richest in carbon, phosphorous, and sulphur, the metal in the crack unquestionably came from the casting interior. Singer and Benek's work is supported by that of the authors.

Figure 40 shows an unetched photomicrograph of a section through a vein, and stringer inclusions can be seen leading from the casting interior to the vein. Figure 41 shows the same section after etching, and the path now is indicated by the difference in metal structure. These samples were taken from a small cast anchor shank and, although the crack was completely filled at the point where the sample was taken, a hot tear was visible a short distance further along the same vein.

Veins under the flanges of valves and fittings occur after tearing and as a result of the tear. The fact that the bore cracks are not completely sealed may be attributed to the fact that bore crack formation is a continuous process.

The crack when it first forms may be completely filled but, as the contraction stresses reopen the crack, the metal becomes too sluggish to refill the crack or the metal skin becomes cooled sufficiently to chill the molten metal before it can refill the crack.

In one of the castings made with a hollow core, the core cracked while it was being placed in the mold. The mold was poured and the crack in the core was filled with metal. The filled crack in the core had the appearance of a vein on the casting, but had no bore crack beneath it.

This accidental test tends to refute a belief that the core cracks first to permit a vein to form, and the vein, since it forms a small hot spot, then makes conditions in the mold favorable for a hot tear to occur beneath it.

Effect of Pouring Temperature on Bore Cracking. A series of test fittings was poured at temperatures ranging from 2850° F. (1565° C.) to 3500° F. (1925° C.) to determine pouring temperature effect on bore cracking. The most severe cracks were found in the castings poured between 2900° F. (1595° C.) and 3050° F. (1675° C.).

On either side of this range the extent of the tear decreased. It may be that with lower pouring temperatures a rapid development of skin strength takes place due to the fast freezing of the cold metal, and that this metal skin is more capable of resisting the tearing stresses.

With the higher pouring temperatures the metal does not begin to

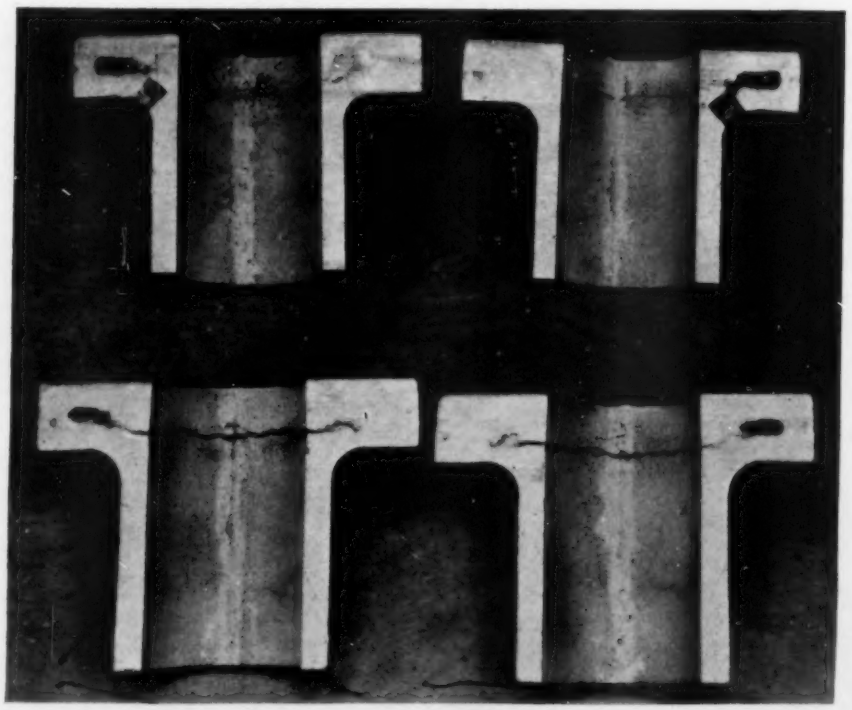


Fig. 37—Fitting made without risers. The flange on the left had cores inserted in it to permit entry of atmosphere to the shrinkage void.

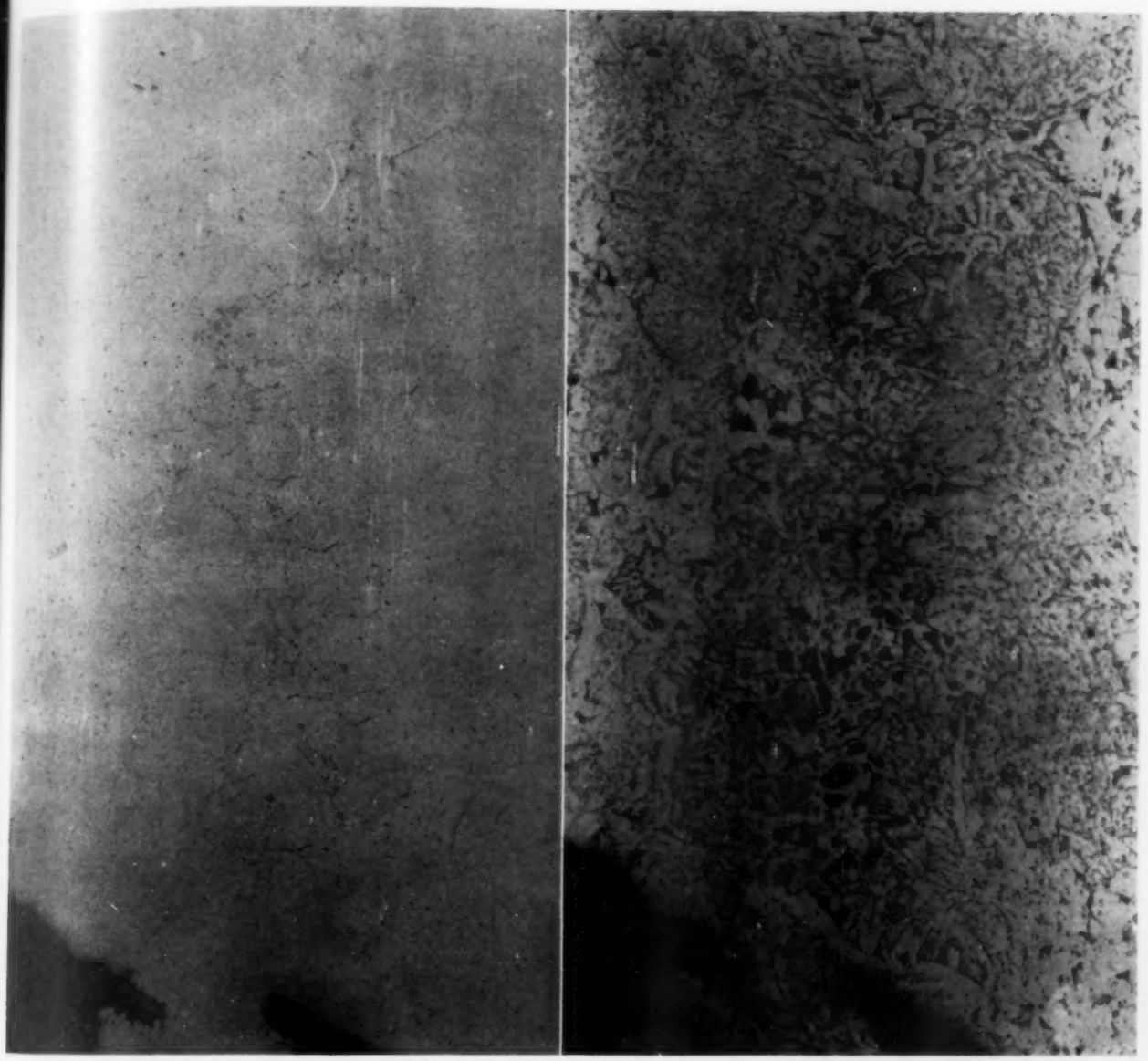


Fig. 40—Photomicrograph of section through vein, unetched. 100X.

Fig. 41—Photomicrograph of same section (Fig. 40) through vein, nital etch. 100X.

solidify immediately after pouring because the superheat must first be extracted by the core, and when solidification does begin, it proceeds slowly. During this time the binders in the core may have time to burn away permitting the core to partially collapse by the time bore cracking begins.

Another possible explanation is that the higher pouring temperatures cause rough surfaces on the casting, and it has been observed many times that the bore crack is less severe on a rough surface. It appears that all of the irregularities on a rough surface grip onto the core and thereby cause distribution of the linear contraction.

Theoretically, the linear contraction is broken down into small increments, each increment being the span between two adjacent projections of the rough surface. However, with smooth surfaces the clamping on the core by the casting is localized at the external flange face at one end and the barrel at the other end, and the area between these points contracts as a unit, with the result that the total contraction becomes localized and causes a tear.

Effect of Metal Composition. The Boston Navy Yard has found that when the carbon content of a cast steel is lowered below 0.10 per cent and the manganese raised above 1.25 per cent the frequency and severity of the bore cracks in valves and fittings are reduced.

A number of alloys, including the high-manganese alloys, were cast into test fittings. The alloying element and the severity of the bore crack found in some of the castings are listed in Table 2. The base composition of these steels unless noted otherwise is 0.25 per cent carbon, 0.40 per cent silicon, and 0.60 per cent manganese.

Beneficial effects of the high manganese content in these steels apparently depend upon a proper balance between the carbon and manganese. Low-carbon steels containing no manganese are susceptible to the formation of bore cracks. A casting made with electrolytic iron and deoxidized only with aluminum had severe cracks. As the carbon is in-

creased above 0.10 per cent, the manganese must be increased proportionally to eliminate the bore cracks.

Chrome-manganese steels are less prone to bore cracking than are "Class B" steels; chromium has an effect similar to that of manganese in preventing the defect. When chromium is present in the steel, the amount of manganese can be reduced. The other common alloying elements in the amounts used in these tests had no effect on the bore crack.

Degree of deoxidation of the steel apparently is not a critical factor in bore crack formation. The addition of large amounts of titanium to the steel appeared to reduce the severity of the crack occasionally, but the results were inconsistent.

Effect of Gas Atmosphere on Bore Cracks. The possibility that bore crack formation might be influenced by the atmosphere of the mold was considered. It is conceivable that oxygen would, by a mild bessemerizing action, enlarge and extend an incipient crack. Conversely, a reducing gas would decrease this tendency. Three molds containing hollow cores were made and gas was introduced into the core.

The wall thickness of the cores was ½-in., which is too great to prevent the formation of hot tears. The gases, which were oxygen, nitrogen, and city gas, were introduced in the mold for 15 min. prior to pouring, thus allowing adequate time for the gas to permeate the mold. These gases had no influence on the bore crack. All three castings tore in the same manner with the same severity.

Summary and Conclusions

Bore cracks are caused by stresses. These stresses are ascribable to core resistance which hinders contraction of the casting.

Bore cracking occurs at the hot zone in the bore under the heavy flange. If valves and fittings are designed so that the flange thickness is equal to or less than the barrel thickness of the casting, the frequency and severity of the bore crack is lessened. Bore cracks do not begin to form until the barrel of the casting has solidified as has been shown by bleeding tests.

Bore cracks occur in castings containing only one flange, and occur in single flanges even though the at-

Table 2

EFFECT OF ALLOYS ON BORE CRACK FORMATION

Alloying Element	Per Cent	Bore Cracks** Gated End Degrees	Gated End Inches	Ungated End Degrees	Ungated End Inches	Remarks
Electrolytic Iron..	100	300	¾	180	¼	1¼ lb. Al used per 100 lb. Fe.
Chromium	1.03	300	½	180	⅜	
Columbium	0.10	180	1	180	½	
Molybdenum	0.47	300	¾	180	1¼	Mo up to 2.00 per cent used. Cracks quite severe
Tungsten	1.47	180	½	180	½	
Vanadium	0.24	300	1	300	¾	
Nickel	4.68	360	1	200	1	
Copper	2.50	360	½	180	¾	
Boron	0.01	360	1½	180	1½	
Tin	0.06	300	¾	200	1	
Silicon	2.10	180	½	180	1	
Cobalt	0.88	180	1¼	270	1¼	
Titanium	0.44	45	¼	no crack		Crack in front of gate
Phosphorus	0.10	360	1	360	1	
Sulphur	0.10	360	1	360	1	
Carbon	0.08†	300	½	‡300	¼	Cracks more prevalent as carbon content increases
	0.82	360	½	360	¼	
Carbon Manganese	0.11 0.07	200	1	180	1	
Carbon Manganese	0.07 1.28	small cracks		small cracks		Probably good enough for production
Carbon Manganese	0.12 1.47	small cracks		small cracks		
Carbon Manganese	0.18 1.46	small cracks		90	¼	
Carbon (avg.) Manganese	0.27 1.42-3.73	90 small cracks	½	270 small cracks	½	Cracks less prevalent as Mn content increases
Carbon Manganese	0.10 1.98	small cracks		no cracks		Probably good enough for production
Carbon Manganese	0.15 2.90	180	½	200	¼	
Carbon Manganese	0.18 2.76	200	1/16	180	1/16	
Carbon Manganese	0.19 3.62	10	½	no cracks		Cracks less prevalent as Mn content increases
Carbon (avg.) Chromium Manganese	0.25 1.17 1.56	180	¼	180	¼	
Chromium Manganese	1.42 2.25	150	¼	60	⅛	
Chromium Manganese	0.92 2.35	30		150		
Chromium Manganese	1.35 2.46	no cracks		no cracks		Probably good enough for production
Chromium Manganese	1.80 2.55	no cracks		no cracks		

*Base composition (percentage) unless otherwise noted—C, 0.25; Si, 0.40; Mn, 0.60; P, < 0.03; S, < 0.03.
**Degrees (°)—Circumferential extent of crack. Inches—Depth of crack.
†Two to four parallel rows.
‡Two to three parallel rows.

tached barrel extends only one in. from each side of the flange. The system from which bore cracks result consists only of the flange and a short length of the barrel attached to it.

Severity of the bore crack can be reduced by gating into the barrel of a valve or fitting rather than the flanges. However, this is not desirable because the principles of controlled directional solidification cannot be applied.

Chills suitably placed in the core under the flange strengthen the metal skin, tend to eliminate the hot spot, and thereby prevent bore cracks. Steel bar chills, with tapered ends extending a short distance under the barrel and with the other ends extending into the core print, have been satisfactory.

Thin, flat chills in the core under the flange will also eliminate bore cracks, but such chills tend to fuse to the casting. Graphite chills in the form of collars or sticks can also be used but must be carefully washed to prevent carbon pickup in the metal.

Hollow cores used in molds for valves and fittings also will yield a casting free from bore cracks. Hollow cores can deform inwardly to relieve any stresses which may form.

The wall of the core must be sufficiently thin to decompose and weaken by the time the barrel has solidified and bore cracking begins, and the walls must also be of sufficient thickness or strength to withstand the metal pressure upon the core. The correct wall thickness of the hollow core depends on the core composition, pouring temperature, and the diameter of the bore of the casting.

Liquid shrinkage is a minor factor in causing bore cracks. When shrinkage is formed within the flange a vacuum exists in the shrinkage cavity, and atmospheric pressure, in attempting to relieve this vacuum, may enlarge an incipient crack.

Pouring temperature has a minor effect on bore cracking. Castings poured at extremely low or high temperatures are less prone to tear than those poured at intermediate temperatures.

Veins and bore cracks are closely allied. A vein in the bore of the casting invariably indicates a bore crack beneath it.

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DISCUSSION

Chairman: H. H. BLOSJO, Minneapolis Electric Steel Castings Co., Minneapolis.

Co-Chairman: CHARLES LOCKE, West Michigan Steel Foundry Co., Muskegon, Mich.

J. R. GOLDSMITH AND W. L. MEINHART[1] *(written discussion):* The authors are to be congratulated on this thorough presentation of the subject of bore cracks in steel castings. It represents a large amount of research, and is a valuable contribution to this perplexing problem.

The writers would like to supplement this paper with some remarks on work carried on in their laboratory while investigating the same problem. Their method of studying the occurrence of bore cracks in steel castings is substantially the same as used at the Naval Research Laboratory.

Their test casting is the same shape, and the size and weight are about half way between that of the two test fittings mentioned by the authors. The casting was molded with a side riser at the face of each flange, and a gate running from the face to each riser.

In their study of the effect of manganese on bore cracks, it was decided to make the steel in three carbon content groups: (1) low carbon, containing less than 0.10 per cent carbon, (2) intermediate carbon, with approximately 0.15 per cent carbon, and (3) high carbon, with about 0.30 per cent carbon. The range of manganese was varied from 0.50 per cent to 3.0 per cent manganese in each series. In all castings in which the manganese content was above 1.5 per cent, regardless of carbon content, many veins and bore cracks were present.

The degree of cracking became much worse as the manganese content approached 3.0 per cent, and only a very slight difference in the intensity of the bore cracks between low carbon and high carbon castings was evident. The lower carbon and intermediate carbon castings were only slightly less affected.

A similar study was made to determine the effect of silicon. The exact opposite of the manganese series was found. The high carbon (0.30 per cent), high silicon (2.0 per cent) castings were entirely free from cracks and veins, but as the silicon was lowered to the usual 0.5 per cent, the intensity of the bore cracks increased. Again, it was the silicon rather than the carbon that seemed to influence the bore cracks.

Castings made from 4 to 6 per cent chromium steels and 8 to 10 per cent chromium steels were very susceptible to bore cracks. Upon deep etching, these castings showed cracks comparable to those in plain carbon steel castings. Austenitic stainless steel test castings were entirely free from bore cracks. This gives an indication that transformation may have some effect on hot tear formation.

The writers obtained substantially the same results that the authors did with their hollow cores by using a migrating type core binder to produce a shell hard enough to withstand the molten metal, and nothing but dry sand beneath the shell. With practically no hardness in the center of the core, it had very good collapsibility. The core that gave good results, however, had a shell that was softer than any core that could be used practically in present day production.

The brief mention about the effect of gases by the authors is an approach that in these writers' opinions warrants more study. On two different occasions, the writers have had periods during which no hot tears could be produced in their test casting. Each of these periods was during two successive summers at which time the humidity was extremely high.

Attempts to duplicate these results by artificial means have failed up to this point. Previous tests and knowledge of behavior of bases in steel indicate that they should have a detrimental effect on steels and promote bore cracks. Inasmuch as the results were encountered twice, the writers question whether the high humidity may have had some chilling effect on the mold.

As some of their cores were stored for long intervals before using, perhaps the core absorbed enough moisture to have a chilling effect. More accurate work along this line and also of the effect of dissolved gases upon steel would con-

tribute to the answer of this problem.

The writers' metallographic examination of bore cracks also tends to confirm the authors' statements that stresses caused by restrained contraction of the metal contribute to the formation of the defect. Bore cracks are interdendritic which suggests that the metal had pulled apart just under the freezing point when the dendrite filling was mushy or nearly so. The edges of the dendrites are not as sharp as those usually found in shrink holes.

[1]Crane Co., Chicago, Ill.

MR. BISHOP: In answer to Messrs. Goldsmith and Meinhardt's statements that the hollow cores which they found successful had shells too soft to be practical, we have tried to use just as strong a core mix as possible and obtain collapsibility by closely controlling the thickness of the shell. We find that by using fairly constant pouring temperatures the cores can be hollowed to a fixed amount so that they will sustain the pressure of the molten metal and will not collapse until the barrel is solid and the stresses begin to act.

There are apparently conflicting results on the effect of chemical composition as an influence on bore cracking. We are of the opinion that composition is important insofar as it effects the properties of the steel at very high temperatures. If the steel is strong enough to resist the contraction stresses, or ductile enough to elongate under these stresses, bore cracks will not occur.

MALLEABLE FOUNDRY COREMAKING PRACTICE

Core sand in a mechanized malleable foundry making pipe fittings and small castings must primarily meet the requirements of the core shop in making cores which, after performing their functions in the molds, become the basis for a satisfactory molding sand

D. F. Sawtelle
Metallurgist
Malleable Iron Fittings Co.
Branford, Conn.

IT IS CUSTOMARY in many malleable foundries making castings varying in size from several pieces to the ounce to several lb. each to use two or three grades of core sands, which vary in A.F.A. fineness from 25 to 162.

When the author presented a paper in 1942* three such widely varying sands were being used in the core shop. They were listed in that paper under Table 1 as, Wareham, Providence, and Jersey No. 52, with A. F. A. Grain Fineness Nos. 72, 160, and 23, respectively.

These sands had been used successfully for several years prior to mechanization, and was thought best to continue their use during the first few months of mechanized operation. At that time, a little more than half the cores were being made in core blowing machines and the remainder by hand.

Coarse Jersey No. 52 sand was considered essential by the core shop, was used in several core mixers to give the cores greater permeability, collapsibility, and help in blowing the cores. It was regarded as detrimental to the system molding sand and was soon replaced by a sand of A.F.A. Grain Fineness No. 35.

Fineness No. 35 sand was not considered as detrimental to the molding sand and, by slight changes in the proportions with which it was mixed with the Wareham and Providence sands, it gave good results in the core shop and in the molds.

The next change in the core sands was the substitution of a so-called "Marion" sand of Grain Fineness No. 119 for both the Providence and Wareham sands. The final change was the substitution of a sand of Fineness No. 48 for the No. 35 sand.

Screen Tests

Table 1 shows the screen tests of the three core sands in use when the mechanized foundry as started on June 23, 1941, and the two sands which are being used at the present time.

Use of the Marion sand and the Jersey No. 45 sand in the core shop has decreased the percentage of coarse grains in the system molding sand. The total percentage on screens 6-12-20-30 has dropped from 7.45 to only 2.03 per cent.

Type of Binders. The malleable core shop uses linseed as an oil binder in combination with a cereal binder known as dextrine sweepings. In past years many types of core oils and dry binders have been tried with varying degrees of success, but under the conditions existing in this core shop and foundry, it is believed that the best results are obtained when using linseed oil and dextrine sweepings.

Other Materials Used. A fine grade of sawdust is added to some core mixtures solely to attain high collapsibility. This is necessary only in cores for some of the largest fittings and as most of these are made on side floors, the sawdust does not enter the system molding sand.

Kerosene is also used in amounts of 0.1 to 0.2 per cent by weight. This does not appear in the core mixtures which follow. Its purpose is to prevent the core sand sticking to the core boxes, to some extent, and to retard the air hardening of a mixed

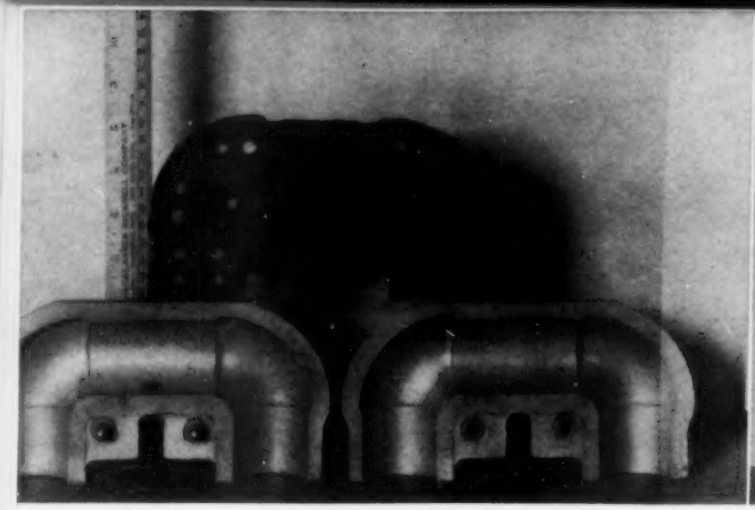
Table 1

SCREEN ANALYSIS OF CORE SANDS

Screen No.	*Remaining on Screen, per cent*				
	Jersey No. 52	*Wareham*	*Providence*	*Marion*	*Jersey No. 45*
6-12-20	8.16	1.02	—	0.09	0.20
30	35.56	0.98	0.04	0.34	2.00
40	47.24	2.28	0.07	0.93	12.10
50	6.68	6.72	0.38	2.85	31.50
70	0.66	21.08	2.30	12.60	36.80
100	0.46	29.12	12.09	24.72	15.20
140	0.22	23.84	24.80	30.16	1.90
200	—	10.04	25.05	16.44	0.30
270	—	2.88	13.21	5.28	—
Pan	—	2.00	22.06	6.59	—
A.F.A. Grain Fineness No.	23	72	162	119	48

This paper was presented at a Malleable Foundry Practice Session of the Fiftieth Annual Meeting, American Foundrymen's Association, at Cleveland, May 7, 1946.

*D. F. Sawtelle, "A Sand Control Program in a Mechanized Malleable Foundry," TRANSACTIONS, American Foundrymen's Association, vol. 50, pp. 830-844 (1942).



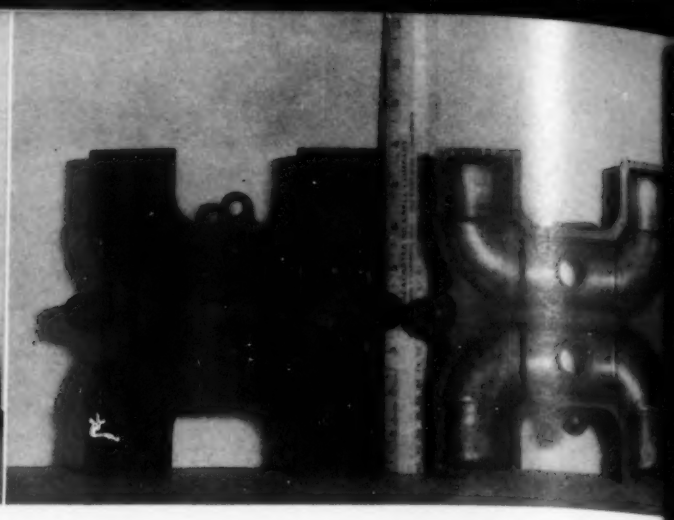


Fig. 1 (left)—Two halves of a core box and a dryer into which core is transferred for baking. Fig. 2 (right)—Multiple system in which the core dryer (left) constitutes one-half of core box and "mother" (right) the other half.

Table 2

MIXTURE FOR SMALL CORES

	Cores	
Material	*Hand-made*	*Blown*
Marion sand, bu. ..	6	5
Jersey sand, bu.	—	1
Linseed oil, qt.	2½	4
Dextrine, qt.	2	7
Water (approx.), per cent	4-5 (by wt.)	2 (by wt.)

batch of core sand before being used.

Kerosene also assists if the moisture in the mixed sands is on the high side. Sometimes this is the case, for while the Jersey No. 45 sand is purchased dry, the Marion sand is not. Water content will be discussed in more detail under core mixes.

Core Mixtures and Consumption of Materials. Six different mixtures in daily use are listed in Tables 2, 3, and 4. Several trends in these core mixtures are worthy of close attention. The sand mixtures for blown cores have less water, more oil, and much more dextrine than the corresponding mixtures for handmade cores. They also contain the coarse Jersey sand which is not used in handmade cores except as in

Table 3

MIXTURE FOR MEDIUM SIZE CORES

	Cores	
Material	*Hand-made*	*Blown*
Marion sand, bu. ..	6	5
Jersey sand, bu.	—	1
Linseed oil, qt.	1¼	3
Dextrine, qt.	2	7
Water (approx.), per cent	5-6 (by wt.)	3 (by wt.)

Table 4, for large collapsible cores. The addition of the coarser-grain sand is an important factor in the successful blowing of these cores.

It is felt that a great deal of the success in blowing cores is due to these mixtures, which has been achieved only after many changes and experiments. The percentage of cores made by this method has increased from 50 per cent in 1941 to 80 per cent at the present time.

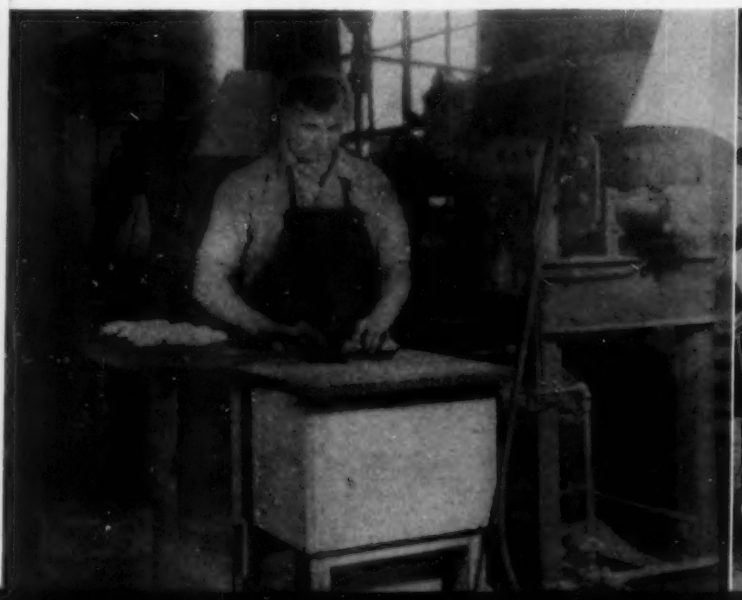
The foregoing core mixes are tabulated just as they are used, with no attempt to convert them to percentage by weight. The amount of

Table 4

MIXTURE FOR LARGE COLLAPSIBLE CORES

	Cores	
Material	*Hand-made*	*Blown*
Marion sand, bu. ..	3½	4
Jersey sand, bu.	1½	1½
Sawdust, bu.	1	½
Linseed oil, qt.	1	1½
Dextrine, qt.	1	3
Water (approx.), per cent	6-8 (by wt.)	4 (by wt.)

Fig. 3 (left)—Blowing cores directly into dryers (multiple system). Dryers are placed on rack for transfer to baking ovens. Fig. 4 (right)—General view of core shop. Belt elevator (left) for sand hoppers supplying two blowing machines. Operator is making cores by unit system.



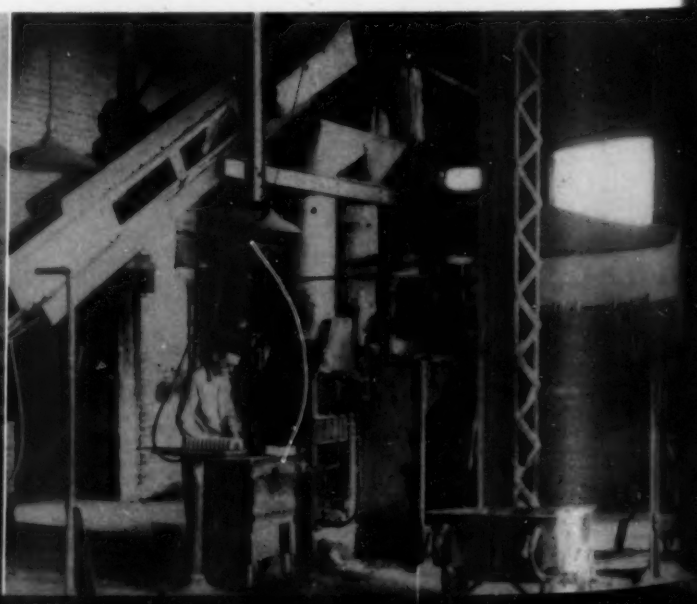




Fig. 5 (left)—Cores stacked on racks being transferred to stationary, batch type, oil-fired core ovens. Fig. 6 (right)—Outlet end of core oven with rack of cores suspended from overhead conveyor.

materials used by weight and the quantity of cores to tons of metal poured are shown in Table 5.

From these figures, it is seen that one ton of cores is needed for every 9 tons of metal poured; that the ratio by weight of core binders to sand is one to 45; that the linseed oil amounts to 48 per cent, and the dextrine sweepings to 52 per cent of the binders used.

Core Making Methods. It is not the purpose of this paper, nor within the scope of the writer to attempt to give in detail the many methods and tricks of the trade used in making cores by hand. Intricate cores, many of the larger cores, and those which could be blown but only a small number required, usually are handmade. The core shop foreman selects the least expensive method available.

Several standard fitting cores are made both by hand and by blowing machines, with costs running equal, but these instances are rare. Where the number required is in the thousands per run, and sometimes a million or more per year, blowing is ideal and the cost is only half that of handmade cores.

In blowing cores, No. 2 machines at 125 lb. air pressure are used. It is important to have full air pressure. There are two methods or systems for blowing cores, different equipment being required for each.

One method is known as the unit system, and the other as multiple. In the unit system a split core box is used, both halves of which are actually machined to the dimensions of the cores to be made and equipped with the necessary air vents, air and sand inlets. The sand is blown into the box and then either transferred to a core dryer in which it is baked, or, in some instances, dried on a tray.

In all cases the cores must be removed from both halves of the core box, which is then ready for re-use. Figure 1 shows the two halves of a core box and a dryer into which the core is transferred for baking.

The core box is cast of low-carbon gray iron which offers better resistance to wear than softer high-carbon gray irons. The box is finished to dimensions in the pattern shop and the necessary air vents inserted. The correct size and placing of the vents, is learned only by experience.

Core dryers are cast of aluminum, or gray iron, a little oversize, sometimes perforated, and are used as cast. A complete set of equipment

Table 5

CONSUMPTION OF COREMAKING MATERIALS

Period Covering Pouring of 14,000 Tons of Metal

Material	*Tons*
Sand	1,600
Linseed oil	17
Dextrine	18.5
Kerosene	0.5
Sawdust	1

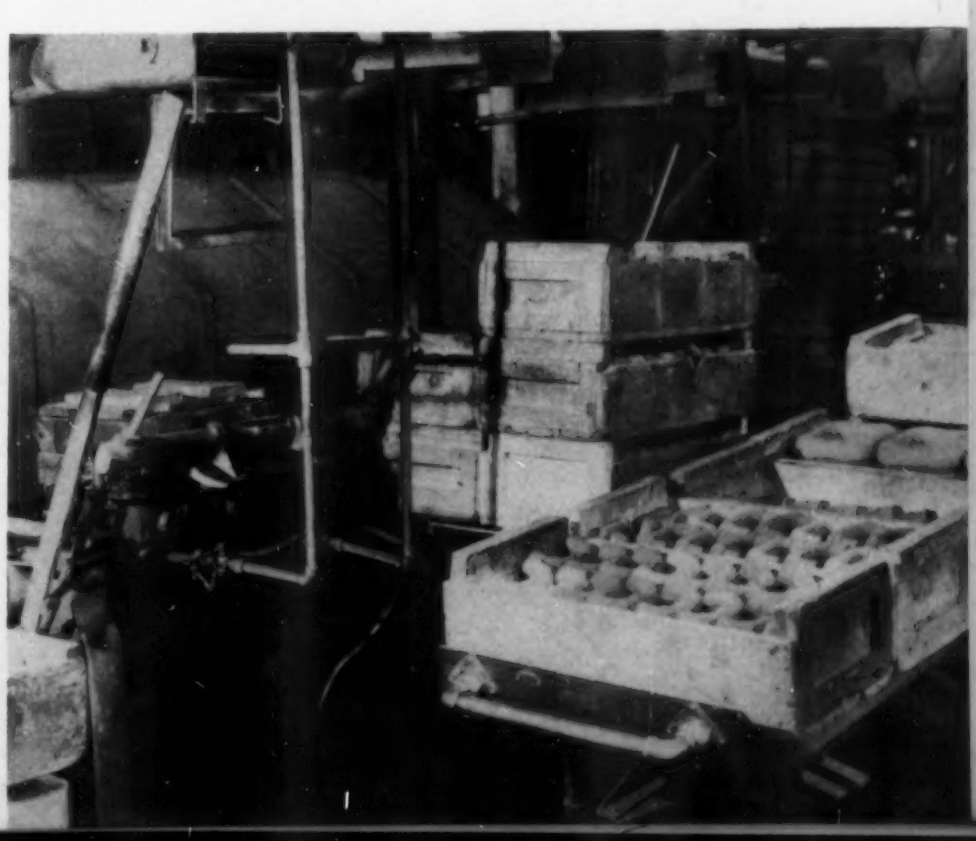


Fig. 7—Boxes of cores loaded on truck for delivery to molding bench.

consists of a carefully machined and vented split core box, and several hundred core dryers. This method and equipment is used when blowing fitting cores of over 2-in. diameter.

In the multiple system, the core dryers constitute one-half of the core box and are all accurately machined and pinned to fit the other half, which is known as the "mother." A set of equipment consists of one "mother" and several hundred dryers. As the sand is blown into the "mother" and one dryer, it is necessary only to remove the "mother," the core remaining in the dryer.

This system is much faster as the ore does not have to be carefully ransferred from the core box to the dryer. The cores are more accurate in dimensions. The necessary machining of the dryers, however, makes the original cost of the equipment much higher. Figure 2 shows the "mother" on the right and the dryer on the left.

Figure 3 shows the actual operation. To the right of the operator is a core blowing machine and to his left a rack on which the blown cores are transferred to the oven for baking. The operator shown is using the multiple system, blowing the cores directly into the dryers.

Core Shop

Figure 4 shows a more general view of the core shop. To the left is the belt elevator for filling the overhead sand hoppers of two blowing machines. In the center is an operator making cores by the unit system. However, the core has not been transferred from the core box to a dryer but is standing on a plate for drying.

Core Baking. The cores, stacked on racks as shown in Figure 3, are then transferred by a hand-jack truck to stationary, batch type, oil-fired core ovens, as shown in Fig. 5. These ovens are heated by oil burners located on the floor below, the heat entering from flues in the center of the oven floor and between the two racks of cores. The exhaust gas flues are located on the inside walls of the ovens about 1½ ft. above the floor level.

All cores over 1-in. thick are baked in these ovens. The time cycle varies from 2 hrs. at 450 to 475° F. for a batch of the smaller cores to as long as 12 hrs. or overnight for the heavier cores. The temperature on the overnight bake is reduced to 325 to 350° F.

The core shop foreman for several reasons prefers the long bake at a low temperature to a quick bake at temperatures of 475 to 500° F., even though linseed oil is capable of producing a good core under a wider variation of temperatures than many of the compounded core oils.

Baking Cores

By weight, the stationary ovens bake a little over half of the cores that are made. All of the smaller cores are baked in a continuous type, oil-fired oven, designed and installed in the author's plant over 30 years ago. It has been in almost continuous operation since that time, the only major change being the installation of oil burners to replace the original coke fires.

Figure 6 shows the outlet end of this oven. A rack of cores suspended from an overhead monorail conveyor can be seen. To the extreme left is a rack from which the baked cores have been removed. They have been placed on the single conveyor located at the extreme right side of the picture, where they proceed to the inspection, counting, and packing in boxes for transferal to the molders' work bench.

Between the empty and full trays can be seen the shaft with a sprocket wheel on the top, which carries the drive chain — moving the trays through the oven and back to the starting point in a continuous loop. The electric motor driving mechanism is on the opposite end.

The trays proceed through the oven at the rate of 1 ft. per min.

The baking portion of the oven, through which they travel, is 45 ft. long; therefore they are in the baking zone at 450 to 475° F. for 45 min. In the next 15 ft. most of the smoke and gases are exhausted. In the last 25 ft., the cores are cooled to handling temperature by air drawn from outside the building and blown across the slowly moving trays of cores.

Core Transportation. The present core shop building was centrally located between two foundry buildings which at one time contained a total of four batch type air furnaces and two cupolas. The present mechanized foundry was built in the farther end of the more modern of these two buildings. This necessitates the transportation of the cores from the core shop to the molders' benches in the foundry, a distance of several hundred feet.

Figure 7 shows the two-wheel, rubber-tired, spring-suspended type of truck, delivering a load of cores to the molders' bench. The raised ends on each box, in which the cores are packed, prevent crushing. Each molders' station has room for five or six boxes of cores. Delivery of cores to the molders and the return of the empty boxes continues during the entire working day.

Since the mechanized foundry occupies only a little over half of the building, plans are being considered for the construction of a new core shop in the foundry. The new shop, with the many improvements which are planned, may well be the subject of another paper on core shop practice.

DISCUSSION

Chairman: R. J. ANDERSON, Belle City Malleable Iron Co., Racine, Wis.

Co-Chairman: D. I. DOBSON, General Malleable Corp., Waukesha, Wis.

MEMBER: How do you mix the sawdust with the core sand? Is the core a dry sand core?

MR. SAWTELLE: We mix the sawdust right with the sand. These cores are baked cores. The first operation in making a batch is to mix 5 bu. of sand with 1 bu. of sawdust. Then add the oil and the dextrine to that mixture.

MEMBER: Did you say it was a dry sand core?

MR. SAWTELLE: It is a baked core, not a green sand core. These are baked core mixes.

MEMBER: Does the sawdust in the core burn out while baking the core?

MR. SAWTELLE: No, not at the baking temperatures we are using (400°-425° F.).

DAVID TAMOR[1]: Why do you use sawdust? Why do you not use wood flour?

MR. SAWTELLE: We have a wood flour which is frequently used. However, we can buy the sawdust cheaper than the wood flour, and it seems to do the job.

MR. TAMOR: In regard to the elbow cores, some were supported in driers and some were standing up. Why is that?

MR. SAWTELLE: The small cores can stand without sagging. However, the larger cores must be wired to prevent sagging when the core is standing. Therefore the cores are supported on driers.

O. J. MYERS[2]: I am interested in your core room foreman's suggestion that he would prefer to bake cores for a longer time at lower temperature rather than at a high temperature for a short time.

[1]American Chain & Cable Co., York, Pa.
[2]Werner G. Smith Co., Minneapolis.

What is the theory to support that practice?

MR. SAWTELLE: He may be thinking of temperature differentials in the oven. In an oven heated at 475° F., unless it is a circulating type of oven, there will be locations where the temperature may reach 500 or 525° F. That is too high. The average batch-type oven has a wide variation of temperatures, and while you say you are baking at 475° F. and are controlling the temperature in some locations of the oven by a thermo-couple, there may be locations that are much hotter. At the lower temperature, in an oven of the type where the temperature differences are great, you will not burn the cores.

MR. MYERS: Did you obtain higher tensile strengths in the cores from that practice?

MR. SAWTELLE: No.

MEMBER: Do you make cores larger than 6 in.?

MR. SAWTELLE: No. That is the largest core we make.

J. C. MCCARTNEY[3]: Are the core driers you use made of aluminum?

MR. SAWTELLE: We have both aluminum and gray iron driers. Those in the unit system are of aluminum, and those in the multiple system are of gray iron.

MEMBER: What would be the grain size difference in sand used for small cores (about ½ in.) and larger cores (about 4 in.)?

MR. SAWTELLE: On the handmade cores we are using all Marion sand, with an A.F.A. grain size of 119. In making small cores by the blowing machines, we use five parts of Marion sand to one part of the 45 A.F.A. grain size Jersey sand. The A.F.A. grain size of that mixture of sand would be approximately 90.

MEMBER: What bond strengths would be used for those cores?

MR. SAWTELLE: We have not made laboratory tests on the green strength of our core mixes. The Marion sand has 0.3 or 0.4 per cent clay-like material which will wash out on an A.F.A. test for clay. That percentage of clay imparts some green strength to the handmade core mix, and in the blown cores the larger amount of dextrine (7 qt. in 6 bu. of sand) gives a fair amount of green strength along with the small amount in the Marion sand itself.

MEMBER: Was the sawdust used for venting the core?

MR. SAWTELLE: No, it imparts collapsibility to the core.

MEMBER: Do you control the particle size of the sawdust?

MR. SAWTELLE: We have been getting the sawdust over a period of years from one source. They know that we want the finest they can supply. I have never made a grain size determination of the sawdust, but it is quite fine. It is not to be compared, however, with wood flour.

CHAIRMAN ANDERSON: Do you encounter much cutting of your core boxes by using 120 lb. pressure for blowing?

MR. SAWTELLE: I have no comparison because we have always used 120 lb. blowing pressure. However, there would be less cutting action from a lower air pressure. The speed at which we can blow a core, using that pressure, and the uniformity and the wide variation of sizes and shapes that we can blow at 120 lb. pressure, would probably overbalance the benefits of lower air pressure resulting in less core box wear.

[3] Union Malleable Iron Co., East Moline, Ill.

SEALING BRONZE PRESSURE CASTINGS THROUGH HEAT TREATMENT

By Lt. (j.g.) Fred L. Riddell, U.S.N.R.,
Formerly Division of Physical Metallurgy, Naval Research Laboratory,
Washington, D. C.

RECENTLY, the repairing of leaky castings by heat treatment has aroused considerable interest because it is considered by some foundrymen to be clean, rapid and fairly effective. The effects of such variables as proper temperature, time of treatment, degree of sealing and effect on mechanical properties have not been extensively reported.

The aim of the present investigation was to answer these questions for gun metal, valve bronze and hydraulic bronze by making porous castings and attempting to seal them by controlled heat treatment.

The mechanism of solidification of tin-bronze castings has been reported [1, 2] as one of the principal causes for porosity and leakage. In freezing, the bronzes with their long range of solidification pass through a mushy stage in which dendrites grow out into the melt and entrap pools of molten metal.

When this entrapped metal solidifies, the shrinkage is distributed throughout the casting wherever the interlocking dendrites prevent adequate feeding. Several investigators have explained the increase in pressure tightness following heat treatment as due to the closing of this microporosity.

In 1917, Elliott [3] reported that leaky bronze castings of complicated design could be made pressure tight by annealing for a short time at 1400° F. Nearly 75 per cent of the rejected castings were saved by this method, provided that the leakage was not too severe. Elliott found that the cored structure of the castings disappeared and that the elongation increased as a result of the heat treatment at 1400° F.

The observations of Elliott were in accord with the results of earlier investigators who found that tensile strength and ductility increased in bronzes held for 30 min. at 1300 to 1470° F.[6]

Jenkins,[4] in the investigation of a bronze of nominal composition 88 per cent copper, 8 per cent tin and

Fig. 1—Experimental bronze castings, bottom gated and cast at temperatures of 2280° F.



Presented at a Brass and Bronze Session of the Fiftieth Annual Meeting, American Foundrymen's Association, at Cleveland, May 9, 1946. Opinions expressed in this paper are those of the author and in no way reflect the official attitude of the U. S. Navy.

4 per cent zinc, concluded that annealing in the range 1300 to 1455° F. gave increased pressure tightness with perceptible increase in ductility and a definite increase in all mechanical properties. Jenkins preferred annealing at this range of temperature to annealing at 1000° F. Annealing for one hour per inch of thickness at a temperature of 1400° F. was recommended.

More recently, Fleck and Bunch[5] heat treated gun metal castings at 1200 to 1400° F. to stop leakage. The sealing was reported to take place throughout the casting, and therefore was not removed in machining. The heat treatments were given for 2 hours per inch of thickness in a laboratory electric furnace.

Complete absorption of the delta constituent took place with a consequent decrease in hardness, but both the tensile strength and elongation increased slightly. A bronze containing 3.39 per cent of zinc softened much less than a bronze containing 2.39 per cent of zinc. Fleck and Bunch[5] warned that sealing would be successful only with good foundry practice.

Microshrinkage in Bronze Castings

Smith, Wiegand and Bolton[6] were doubtful about salvaging castings by heat treatment. They stated that microshrinkage was not removed by fusion, but merely blocked by oxide scale. The original defect and weakness were believed to remain.

It was also stated that valve bronze annealed at 1600° F. became partially fused and had low mechanical properties. Quenching from 1000° F. proved to be harmful to the mechanical properties of this bronze. Slow furnace cooling from the annealing temperature was recommended if the mechanical properties were to be unimpaired.

Horner and Mason,[7] in a more recent publication, reported that 90 per cent of a group of 150 rejected castings was saved by annealing for 2 hours at 1200° F. The authors emphasized that leakage from foundry defects such as tears, cracks, cold shuts and direct inclusions could not be remedied. To be benefited by heat treatment, the casting must not show more than a slight "weeping" when under pressure. The proportion of castings salvaged by heat treatment depends to a considerable extent upon the skill of the foundryman in properly classifying the failure before attempting the cure.

There is little agreement among the investigators as to the cause of improved pressure tightness of bronze pressure castings. Jenkins[4] stated that sealing was dependent upon an increase in volume of the matrix resulting from absorption of the dendritic structure. Others have claimed that the segregated constituents melt and flow into the regions of the shrinkage, thus blocking the interlocking channels of porosity.

Smith, Weigand and Bolton[6] believed that the formation of oxides of larger specific volume than the base metal was responsible for blocking the shrinkage cavities through which leakage occurred. Smith, Weigand and Bolton did not favor heat treatment as a means for increasing pressure tightness because the original weakness of the metal was believed to remain although the porosity might no longer be sufficiently open for leakage.

In the present investigation, 34 bushings of gun metal, valve bronze and hydraulic bronze were cast in a manner intended to produce leakage in a hydraulic pressure test. The pressure at which the bushings began to leak was recorded after various sealing experiments. The changes in microstructure and mechanical properties resulting from heat treatment were also investigated.

Experimental Procedure

(a) Foundry Procedure. The melting charges consisted of either Grade A copper, zinc, Straits tin and lead of high purity or selected scrap of known composition from the foundry of the Naval Research Laboratory (Table 1). The charges were melted in a 210-lb clay-graphite crucible in a lift coil induction furnace. The virgin copper or the selected scrap was melted under a charcoal cover in order to prevent excessive oxidation of the metal.

When the charge had become molten, two ounces of phosphor-copper (15 per cent phosphorus) per 100 lb. of metal served for deoxidation before making the additions of alloying elements. The tin, lead and zinc, with necessary compensation for melting losses, were then added in that order, and the addition of two or more ounces of phosphor-copper was made before casting. The bronze was intentionally cast at a high temperature, close to 2280° F., in order to increase the likelihood of porosity.

The molds were made from Albany green sand (Grade 00) containing from 7 to 8 per cent moisture. The molds were allowed to air dry at least 24 hr. before casting. The bushings were cast with two in each mold and were bottom gated (Fig. 1).

Thirteen gun metal, 14 valve bronze and 12 hydraulic bronze bushings were poured from the same temperature (2280° F.) with the same gating and risering practice. The standard cast tensile test coupon (Type 10A of ND Specification 46 M6g) was used for determining the mechanical properties of the three compositions.

(b) Machining of Bushings. The "skin effect" of each alloy was in-

Table 1

COMPOSITION OF MELTING CHARGES

Heat No.	Elements, per cent				
	Cu	*Sn*	*Zn**	*Pb*	*Ni*
A1	87.78	8.95	3.17	0.10	—
A2	87.98	9.40	2.62	—	—
A3	88.40	9.10	2.36	0.14	—
A4	87.49	9.27	3.11	0.13	—
A5	88.02	8.32	2.85	0.08	0.78
B	87.36	4.80	2.00	0.95	4.89
C1	88.50	5.99	3.82	1.69	—
C2	88.41	5.28	3.83	2.48	—
C3	87.15	6.51	4.53	1.81	—
D1	85.65	5.11	3.76	5.84	—
D2	84.58	4.89	4.91	5.62	—
D3	84.76	5.18	4.82	5.24	—
E	88.05	6.09	2.81	—	3.05

*Zinc by difference.

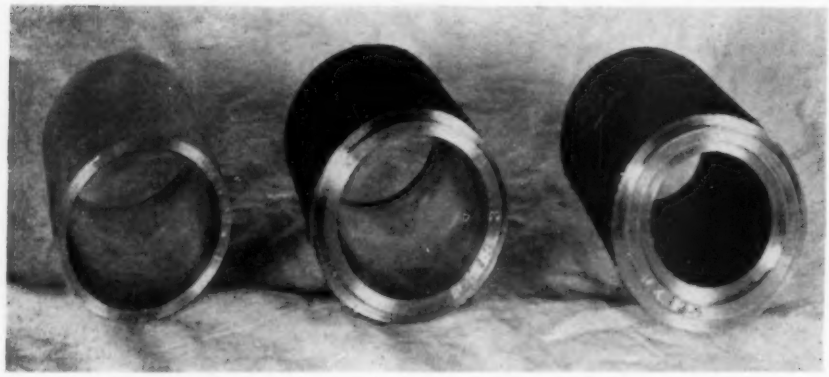


Fig. 2—Castings prepared for determining "skin effect" on pressure tightness, (left) inside and outside surfaces machined, (center) inside surface machined, (right) as-cast.

vestigated by (1) pressure testing in the as-cast condition, (2) with the inside surface machined, and (3) with the outside surface machined, approximately 1/16 in. being removed in each instance. A rim was machined on the ends of the bushings in order to fit the hydraulic tester.

The inside and outside diameters of the rim correspond to the diameters of the finished bushings. The inside diameter was 2-7/32 in., the outside diameter was 2-9/16 in., and the overall length of the bushings was 4-¼ in. (Fig. 2).

The highest test pressure for the bushings was 750 psi. This pressure was maintained for 5 min., and if no weeping occurred the bushings were considered pressure tight.

(c) Heat Treatment. The heat treatments were carried out in air, hydrogen, or oxygen-rich air atmospheres in a muffle furnace. The bushings with machined surfaces and the test bars with as-cast surfaces were placed in the furnace at room temperature, heated to the desired temperature, held for the desired length of time and then furnace cooled. Measurements of the diameters of the bushings were made following the heat treatments because of the possibility of dimensional changes.

In annealing the bushings in the hydrogen atmosphere, calcium chloride was used to absorb the water vapor and copper gauze at 842° F. (450° C.) was used to remove the oxygen in the hydrogen. In annealing the bushings in the oxygen-rich air, no attempt was made to remove the impurities in the oxygen.

Two manometers were placed in the system, one in the oxygen line, and one in the compressed air line in order to admit the respective gases at equal pressures. This resulted in a mixture of constant composition entering the box in the furnace. The gas entering and leaving the furnace was analyzed with an Orsat apparatus. The gas sample taken at the inlet contained 43.6 per cent oxygen and the sample taken at the outlet contained 39.0 per cent oxygen. The rate of flow was approximately 100 ml. per min.

(d) Micro-examination. The change in microstructure with heat treatment was investigated by heating samples of gun metal, valve bronze and hydraulic bronze at 1350° F. for 15 and 30 min., and at 1400° F. for one hour. The specimens were heated with the furnace, held at temperature and furnace cooled, then cut in half and polished on the freshly cut surface. They were etched with ammonium hydroxide and hydrogen peroxide followed by ferric chloride and alcohol. The photomicrographs were taken at 75 diameters.

Bushings failing at less than 25 lb. pressure were broken apart near the place of leakage. The inside surfaces of the shrinkage cracks were then examined at 75X magnification. Cross sections of bushing E-1 of Group 1 taken through the region of failure were projected on a ground glass screen for measurement of the widths of the shrinkage cracks and observation of the internal oxidation.

(e) Mechanical Tests. Unmachined coupons for tensile test bars were heat treated at the same time as the specimens for micro-examination and bushings for the sealing treatment. The tensile bars were of standard 0.505 in. diameter and

Table 2

CHANGES IN MECHANICAL PROPERTIES OF GUN METAL, VALVE BRONZE AND HYDRAULIC BRONZE WITH ANNEALING AT 1300 TO 1400° F.

Tensile Specimen	*Heat Treatment*	**Tensile Strength, psi.*	*Yield Strength, psi.*	*Elongation, per cent in 2 in.*	*Brinell Hardness, 500 kg. 30 sec.*
HYDRAULIC BRONZE					
D1	1	34,775	15,400	31.5	59
	2	29,400	15,100	18.0	52
	3	35,600	15,500	35.5	52
	4	34,100	15,300	29.5	50
	5	32,700	15,800	25.5	50
	6	34,225	15,425	28.5	51
VALVE BRONZE					
C2	1	36,600	18,100	30.55	60
	2	35,800	17,800	29.0	56
	3	36,600	18,200	36.5	56
	4	36,750	17,250	38.4	52
	5	35,850	17,000	31.7	51
	6	37,400	16,500	39.7	51
GUN METAL					
A3	1	44,000	21,250	34.5	73
	2	42,850	20,375	28.5	72
	3	46,650	22,150	46.0	72
	4	46,500	20,400	48.5	69
	5	44,400	19,000	52.5	57
	6	43,400	19,000	46.0	59

*Key to Heat Treatment—1. As Cast. 2. 1350° F. for 15 min. in air. 3. 1350° F. for 30 min. in air. 4. 1400° F. for 1 hr. in air. 5. 1400° F. for 3 hr. in H_2. 6. 1400° F. for 2 hr. in O_2 atmosphere.

Table 3

RELATIVE PRODUCTION OF POROUS CASTINGS FROM THREE ALLOYS

Alloy	*Bushings Cast*	*Leakers (Machined I.D. and O.D.)*	*Porous Castings, per cent*
Gun metal	13	12	92.5
Valve Bronze	14	11	78.5
Hydraulic Bronze	12	8	66.7

2-in. gauge length. The elongation, yield strength at 0.5 per cent offset, tensile strength and Brinell hardness are given in Table 2.

(a) Foundry Practice. To produce castings with the type of defect which would permit sealing by heat treatment, the foundry practice was chosen so as to give defective bushings. For the present purposes, perfect castings as well as those with gross defects were discarded. The high pouring temperature, bottom gating and 8 per cent of moisture in the molding sand generally gave the desired degree of porosity. The relative difficulty in making porous castings of the three alloys is shown in Table 3.

It is apparent that hydraulic bronze has the least tendency to leakage. Although 31 leaky castings were produced, less than half of these contained the "weeping type" of leakage suitable for sealing by heat treatment. Three nickel-tin bronze cylinders were also included in the present investigation in order to test the possibility of sealing this composition of bronze.

(b) Effects of Heat Treatment on Pressure Tightness. The available bushings were divided into two groups for testing. The first group, consisting of gun metal and the nickel-tin bronzes, was machined inside and out before testing.

The second group, made up of gun metal, valve bronze and hydraulic bronze, was tested first as cast in order to investigate the influence of the oxidized skin before machining the inside and then the outside diameters. The subdivision of the bushings of each group on a basis of composition was intended to reveal possible differences in the tendency to seal by heat treatment.

In Table 4 the bushings of Group 1 show that partial sealing by heat treatment has been confined largely to the outside layers of the bushings. Although the average pressure tightness has increased more than 100 per cent by annealing at 1350° F. and 1400° F., this means little from a practical point of view because a light polishing with emery cloth was sufficient to remove the thin layer of pressure-resistant oxide.

The reason for the average pressure of column 4, (Table 4) being below the average pressure of the machined bushings of column 1 is explained by the tendency of the machine tools used in preparing the specimens to drag metal into the openings of shrinkage cavities.

This mechanical obstruction was removed by the oxidation in treatments 2 and 3. Heat treatment in dried H_2 caused a further drop in pressure tightness because oxides occurring in shrinkage cavities were reduced to pure metals having considerably less volume than the oxides.

Reheating in an oxygen-enriched atmosphere for 3 hr. did not repair the damage done by hydrogen treatment. However, the longer time of treatment and higher concentration of oxygen in treatment 6 gave greater penetration, as shown by moderate decline in average pressure when the oxide scale had been removed (Table 4).

The bronzes of Table 5, in general, had a dense, pressure-tight skin which gave a fair degree of pressure tightness in the as-cast condition. The following variations in treatments gave results somewhat different from those in Table 4. Longer time at temperature gave higher pressure tightness. Annealing for 3 hr. at 1400° F. appeared to give no better results than annealing at 1300° F. for the same length of time.

Distortion of Thin Sections

In fact, annealing at the higher temperature may have been dangerous because of distortion of thin sections. The bushings annealed at 1400° F. had variations of diameter of approximately 0.005 in. The removal of the oxide scale did not result in any appreciable loss in pressure tightness following the longer annealings at 1300° F.

The hydrogen treatment also did not prove so detrimental. However, none of these results can be accepted as favorable, because the bronze is still mechanically very weak. The regions where leakage had previously taken place still possessed porosity easily visible at 75X magnification. The fracture of the porous bronze had a reddish discoloration.

The hardness remained lower in regions of leakage than in surround-

Table 4

HIGHEST PRESSURES FOR NO LEAKING OF BRONZE BUSHINGS
(PRESSURE HELD 5 MIN.)
Group 1.

	Pressures, psi.						
	1	2	3	4**	5	6	7**
Bushing Number	*As Machined*	*1350° F., 15 min.*	*1400° F., 1 hr.*	*Oxygen Scale Removed*	*1300° F., Hydrogen Atmosphere, 3 hr.*	*1200° F., Oxygen Atmosphere, 3 hr.*	*Oxygen Scale Removed*
A1-1	50	170	100	0	0	5	10
A1-2	100	200	300	25	5	*	25
A1-3	120	400	200	25	10	*	50
A1-4	300	400	600	40	5	100	75
A1-5	25	150	300	35	20	100	75
B1	300	750	750	150	25	400	350
B2	250	400	750	50	25	200	200
E-1	50	125	100	25	†	—	—
A5-1	100	300	320	25	10	25	25
Avge. Pressure	144	322	380	42	12	138	101

*Not heat treated further.
**Oxide scale removed by polishing with emery paper.
†Sectioned for photomicrographs.

ing regions of the surface which were pressure tight. At many places the difference in hardness was twenty to thirty Brinell (converted from Rockwell E), and with Brinell hardness of 60-70 numbers this means that a major percentage of hardness has been lost through the defectiveness of the material. Low hardness of bronze is an indication of unsoundness.

As further indication that sealing is a process of oxidation, two bushings, as-cast, were annealed in hydrogen without previous heat treatment in an air or oxygen-rich atmosphere. The pressure tightness was found to be the same or less when retested after this treatment.

Changes in Microstructure

(c) Microstructure. The disappearance of the cored structure and the delta constituent on annealing for various periods of time are the most significant changes in the microstructure. The grain size was unchanged by the heat treatments. In general, the delta constituent disappeared after one-quarter hour and the dendrites after one hour at temperature (plus slow heating and cooling), as shown in Fig. 3.

There was little difference in the appearance of the structure of gun metal, valve bronze, and hydraulic bronze after one hour of annealing. The dendrites and the delta phase had been dissolved in the matrix of alpha solid solution. The black areas in the gun metal are porosity and in the hydraulic bronze both porosity and lead particles.

The microstructure of a bronze which leaks under pressure is the same as that of any other bronze except for the presence of porosity. A photomicrograph of specimen E-1 of Group 1 (Table 4) which leaked at a pressure of 25 lb. showed that the microshrinkage cracks are from 0.01 to 0.05 in. wide (Fig. 4).

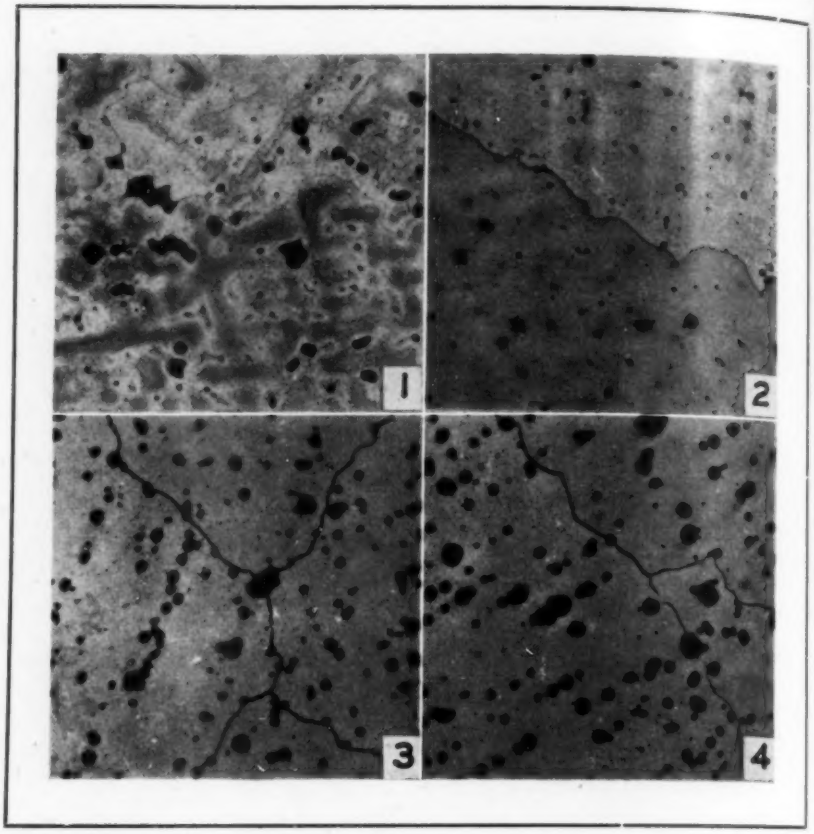


Fig. 3—Change of microstructure of composition "G" with heat treatment. (1) As-cast. (2) Heat treatment at 1350° F. for 15 min. (3) Heat treatment at 1350° F. for 30 min. (4) Heat treatment at 1400° F. for one hr. Magnification, 75X. Etchant, NH_4OH-H_2O_2 and alcoholic solution of $FeCl_3$. Specimens heated slowly to temperature and cooled slowly in furnace.

Since the sample was not etched, it is possible to identify the fringe along the edges of the cracks as an oxide layer. In this bushing the cracks were too wide for the oxide to fill the gap, and sealing by heat treatment was not effective.

(d) Mechanical Properties. The mechanical properties of sound tensile bars were little changed by the heat treatment, with the exception of the decrease in hardness which coincides with the disappearance of the delta phase after about 15 min. of annealing at 1350 to 1400° F. The hardness of gun metal, valve bronze and hydraulic bronze specimens generally decreased by 10 to 15 points Brinell (500-kg. load).

Gun metal specimens increased in tensile strength from 44,000 to 46,500 psi., and in elongation from 34.5 to 46.0 per cent after being annealed in air for one hour at 1400° F. Additional annealing of gun metal specimens produced slight decreases in tensile strength and yield strength. Annealing sound tensile bars of the three bronzes in hydrogen and oxy-

Table 5

HIGHEST PRESSURE FOR NO LEAKING OF GUN METAL, VALVE BRONZE AND HYDRAULIC BRONZE BUSHINGS (PRESSURE HELD 5 MIN.)

Group 2.

	Pressures, psi.							
	1	2	3	4	5	6	7	8
Bushing Number	*As Cast*	*Machined, Inside Diam.*	*Machined, Outside Diam.*	*1200° F., 3 hr.*	*1300° F., 3 hr.*	*1400° F., 3 hr.*	*Oxygen Scale Removed*	*1200° F., Hydrogen Atmosphere*
A2-1	50	50	300	620	700	700	700	—
A2-2	75	75	75	180	240	300	100	75
A2-3	125	750	300	700	700	750	750	450
C1-1	700	750	25	50	200	100	250	100
C1-2	750	150	25	50	200	200	220	120
C1-4	750	750	25	100	200	200	190	—
D1-1	750	750	100	750	750	750	750	300
D1-6	750	750	100	750	750	750	750	100
Average Pressure	494	503	144	400	468	469	464	277

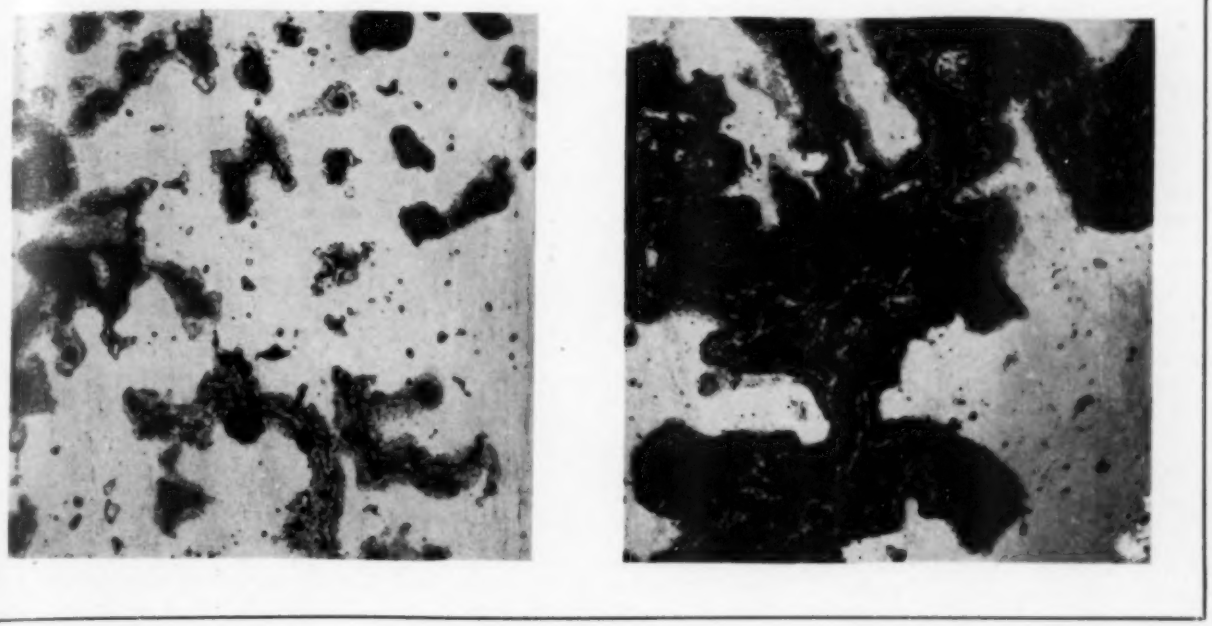


Fig. 4—Photomicrographs of bronze casting (Specimen E-1, Group 1, Table 4) showing microshrinkage. Heat treatment—1 hr. in air at 1400° F. Cross-sectioned and polished. Unetched. 75X. (From two different locations of same specimen.)

gen produced little effect on the mechanical properties (Table 2). Valve bronze and hydraulic bronze specimens were much less affected by the annealing treatment than was the gun metal.

Conclusions

The pressure tightness of leaky bushings of gun metal, valve bronze and hydraulic bronze was improved when they were annealed for 3 hr. at 1200 to 1300° F. in an air or oxygen-rich air atmosphere. No sealing whatever took place in a hydrogen atmosphere.

When bushings previously improved in pressure tightness by annealing in the oxygen-rich air were re-annealed in hydrogen, the pressure tightness decreased by 25 to 50 per cent. Removing the outer layer of oxide scale after sealing in air did not cause appreciable loss in pressure tightness provided that the bushings were annealed for 3 hr. at the necessary temperature (Group 2, Table 5).

The bronze specimens used in this investigation were sealed by the formation of oxides of larger specific volume than the base metal. The oxide layer on the inside of fine shrinkage cracks tended to close the openings provided that the cracks were of sufficient fineness. The tendency for leakage of unsound bronze may become serious after several years of service, and the wear resistance of moving parts is certainly lower with unsound metal.

The short-time pressure test at room temperature is not equivalent to conditions of service, which include the factors of corrosion, abrasion, application of stress, wear and changes in temperature. It is believed that the annealing of leaky bronze castings to improve pressure tightness should be exercised with care until such time as information may be obtained over a period of years on the performance of sealed castings in various services.

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DISCUSSION

Chairman: G. K. DREHER, Rogers Pattern & Foundry Co., Los Angeles.

CHAIRMAN DREHER: Do you know the ratio of depth of penetration of that oxide with time? Can you measure the depth of penetration? Can you increase it by extending the annealing treatment?

MR. RIDDELL: We tried taking photomicrographs. We did not go from the surface in. The photomicrographs were mostly in regard to porosity fissures. I could not say what the depth of penetration is and if it could be increased by extending the annealing treatment.

CHAIRMAN DREHER: Are you reasonably certain that was the effect?

MR. RIDDELL: Yes. We polished a bushing in an area that we knew leaked under a pressure of about 25 psi. Appearance in the etched and unetched conditions indicated, in my opinion, presence of oxides. Dr. Loring also looked at them and thought they were oxides.

CHAIRMAN DREHER: It seems like a rather strange procedure in a way.

J. A. DUMA[1]: I might mention an interesting example of a practical application of sealing bronze pressure castings through heat treatment in our foundry. It is our practice to radiograph all large

pressure bronze castings completely. When examination of the X-ray films indicates presence of microshrinkage the casting is seal-heat treated prior to any machining.

Castings containing no visible microshrinkage are not given this heat treatment. Nor do castings containing pronounced shrinkage areas, i.e., microshrinkage receive the treatment. Repair by welding followed by seal heat treating or scrapping of the casting are the only alternatives resorted to in the latter instance.

CHAIRMAN DREHER: Should we draw the conclusion from your statement that the procedure can be applied to what is known as microshrinkage areas?

MR. DUMA: That is right.

S. W. BRINSON[2]: We have to make the infamous silicon bronze. I know we had nothing but trouble with it. We are making Navy bells out of it now and we have to turn the pattern upside down and pour from many positions. In a silicon bronze elbow we had 14 leaks right through a constant section. We heat treated it and got it down to two leaks. We did cure some of it by heat treating.

T. C. BUNCH[3]: We did some work on this same subject about eight or nine years ago. The work was forced on us because we were having considerable trouble with leakage. Difficulties were encountered principally with high pressure castings manufactured from Navy Composition "G." We found that by annealing or heat treating in an oxidizing atmosphere an increase in physical properties was obtained, with the exception that hardness was reduced, and leakage due to microscopic shrinkage was sealed. Apparently oxidation played an important part in the sealing process since it was noted that in certain instances subsequent machine operations on the castings caused leakage to re-develop.

CHAIRMAN DREHER: This treatment should apply very close to the finished stages, then?

MR. BUNCH: If microshrinkage is extensive, heat treatment should be accomplished close to the finishing stages. Preliminary x-raying of rejected castings should indicate in most cases if heat treatment would be effective. We believe that proper melting practice, minimum machine finish allowances, judicious use of chills and engineered gating and risering will eliminate to a great extent the necessity of using heat treating salvage methods. We did not have to resort to heat treating during the past four years.

CHAIRMAN DREHER: Another conclusion has come out of this discussion that reinforces our previous opinions about the presence of hydrogen and the desirability of oxygen in the melting and handling of bronze. Has anybody carried on work with annealing in hydrogen? You have definitely reversed the procedure by using hydrogen instead of oxygen.

MR. RIDDELL: We heat treated in pure hydrogen and in every case the pressure tightness of the bushings decreased.

E. D. BOYLE[4]: There is possibly a related phenomenon in powdered metals.

MEMBER: In testing cylinder castings and valve castings, I believe the correct method would be to bring the pressure up until the elastic limit of the material is exceeded. In other words, when the casting is not pulled out of shape, release the pressure. That is the test, not merely an application of 200 lb. or 500 lb. pressure.

LT. COMDR. B. O. BROUK[5]: The Navy Department does not advocate the heat treatment of bronze castings even after they show a slight amount of microshrinkage. During the vast submarine building program, it was necessary to get a certain amount of G and M-metal castings out and get them out quickly, so they resorted to heat treatment at about 1350° F. and 1400° F. for about 2 to 3 hr. and I believe the figures will show they salvaged, without future welding and sending back to the foundry, about 46 per cent of their defective castings by this heat treatment method. It was through the Portsmouth (N. H.) Ship Yard's experience that the Naval Research Laboratory conducted this test and we went ahead and authorized it. We thought we could develop a test whereby it would even improve the method they used at the Portsmouth Navy Yard. However, after we found out that there were oxide inclusion formations we believe today it is not sound practice. We believe that it is just adding something into the metal as oxide inclusions. But I do mention that it was used in practice and as a method of salvage. To date we have not had any direct complaints and direct service rejects from this type of casting, although I could say the Navy Department does not advocate heat treatment and will not advocate compound impregnation.

CHAIRMAN DREHER: It is worth while to have that comment about the service life of those castings that were treated by this procedure.

[1]Norfolk Navy Yard, Portsmouth, Va.
[2]Norfolk Navy Yard, Portsmouth, Va.
[3]U. S. Navy Yard, Pearl Harbor, Hawaii.
[4]Puget Sound Navy Yard, Bremerton, Wash.
[5]Bureau of Ships, Washington, D. C.

RECOMMENDED PRACTICES FOR ALUMINUM AND MAGNESIUM PERMANENT MOLD CASTINGS

A.F.A. COMMITTEE REPORT

A.F.A.
Aluminum and Magnesium
Division
Committee on
Permanent Mold Casting

EVERY METHOD of metal forming has its own peculiar and characteristic set of design principles that should be followed for best results. Design engineers must therefore consider the process by which parts are to be made to obtain maximum properties for the completed assembly. A frank and open discussion of parts design by the design engineer and the permanent mold foundry engineer can substantially benefit both parties. Sometimes minor changes are enough to produce increases in production, manufacturing economics and/or improvements in physical properties.

Withdrawal of Cores

Provisions must be made in the casting design for the withdrawal of cores and for the removal of the casting from the mold. Where the design cannot be changed to permit the use of a solid core, split cores or multiple pieces of sand cores must be used. A minimum number of cores cuts down mold operating time and thus expedites production. Long thin cores should be avoided to keep maintenance costs at a minimum.

The exact jigging points should be marked on the blue print to aid the mold designer in placing parting lines, gating and vents. It is the practice to part the mold at the point that will permit easy removal of the casting. At the same time the parting lines formed by different parts of the mold are in mutual relationship as not to interfere with machine tool locating points. These are of real aid to both buyer and supplier in checking the accuracy of the casting from the same spots. One should also keep in mind the appearance of the joint marks on the finished part.

Metal Inserts

It is sometimes found desirable to cast inserts of other metals in light metal permanent mold castings. Since the light metal alloys have different coefficients of expansion from other metals, a definite foundry technique is required for the economical and proper use of inserts.

Wherever practicable, uniform sections should be maintained throughout a permanent mold casting. Section uniformity helps not only to simplify the problem of gating and feeding, but also to equalize the rate of solidification, a very important factor in controlling the quality of permanent mold castings. With any kind of cast products, when sections vary in thickness, there is a tendency to set up internal strains in the casting and cause non-uniform solidification, which may lead to shrinkage and cracking.

Gradual Section Increase

Therefore, in a design where light and heavy sections join, a gradual increase in thickness of the thinner section toward the point of junction will help considerably. In service, this blending assists in distributing loads over a larger area of the casting and thereby frequently eliminates undesirable stress concentrations.

The minimum section thickness possible in light metal permanent mold castings depends very materially on the size, pressure requirements and intricacy of the casting. In general, the minimum section thickness which can be cast in permanent molds is about 1/8 in. When extremely thin sections are required, however, it may be necessary to make an engineering compromise, since it is necessary to sacrifice smoothness of surface to attain the desired thinness.

The designer should provide generous fillets at all wall intersections in a permanent mold casting to help prevent shrinkage and cracking at these points. Sharp internal corners constitute a source of weakness in castings and should be avoided.

Multiple-Piece Cores

The use of multiple piece and dry sand cores broadens the scope of the permanent mold process and makes possible the production of some castings that could not otherwise be made by this method. If the "undercut" which necessitates multiple piece cores can be eliminated a savings in cost will result.

It is possible to hold dimensions on permanent mold castings within close limits. Very close tolerances may be reflected in increased product costs and in lower production rates. From the standpoint of economy, the designer should not specify tolerances closer than are absolutely necessary.

Presented at an Aluminum and Magnesium Round Table Luncheon at the Fiftieth Annual Convention of the American Foundrymen's Association in Cleveland, May 6, 1946, by A. Sugar, American Metal Co., Ltd., New York.

Mold Design. The production of sound castings by any method depends on the control of solidification. In permanent mold castings of regular section, it is not very difficult to attain progressive solidification. Castings for engineering use, however, are usually more intricate in shape and, therefore, solidification is not so readily controlled.

The general design of the mold, its wall thickness, the location of cores, of light and heavy sections in relation to mold partings, the position and size of vents and chills, consideration of gate and riser size and of gating and feeding position should all be approached with the object in mind of securing progressive solidification in the casting.

Types of Molds. The simplest type of permanent mold is a cavity in one piece of metal into which molten metal is poured and allowed to solidify. The cavity should have sufficient taper to permit the removal of the solidified casting. Uncored casting of simple shape may be made in two-piece molds, each mold half providing part of the mold cavity. A gate may be cut in each mold half and a riser may also be provided. Other simple uncored castings may require a three-piece mold consisting of two mold halves and a base. The more complicated the casting the more complex is the mold.

Permanent Mold Types

Some of the more common types of permanent molds may be roughly and briefly discussed. A type of hinged mold for making simple castings has one mold, half of which is stationary and mounted vertically on a base or some other means of support. The moving mold half is hinged to the stationary section by means of a pin. Cores and core pins may be pinched out by hand-operated pinch bar, pulled with hand-operated toggle, rack and pinion or an air cylinder, depending on the size.

If the core is large, it is often made an integral part of the stationary mold half. Often the cavity for the casting is all in the stationary part and the moving segment acts only as a cover. Taper on the finished casting is so placed as to make it stick to the stationary part of the mold.

If there is a large core, it may be made to serve as the means of ejecting the casting, or should this not be practical an ejector plate mounted on the stationary mold half and containing several equally spaced ejector pins operated simultaneously by the plate may be used for removing the casting adhering to the stationary mold side. Repeated accurate alignment of the sections of the mold cavity in relation to each other is insured by doweling the mold sections to each other. Tight closure is insured by the use of hand clamps.

Molds such as just discussed are generally used to make comparatively shallow castings of small to medium size. A modification of this type of mold used for making a rather large box with straight sides and many bosses consists of a core forming the stationary mold side to which an ejector plate is also mounted; hinged to each side of the stationary segment are two equal mold sections which make the outside of the box and which swing away from the casting after it has been poured.

Mold Segment Alignment

When deeper castings than considered above are to be made on the same type of mold, the movable half is supported independently. All of the mold segments are mounted on a base, with the stationary half bolted in position. The movable half is located by and moves back and forth guided by a keyway which is made an integral part of the supporting base. Again doweling should be used to insure the accurate alignment of the various mold segments.

To operate the movable mold segments a hand toggle, ram box, rack and pinion, air cylinder or hydraulic mechanism may be used. Cores and ejectors are again mounted and operated as previously discussed, generally by a rack and pinion. Hand clasps should be used when closure obtained with mechanical units is not positive enough.

Where core or casting is so massive that all mold segments must be moved to clear the finished castings, the mold assembly is somewhat different. When there are only two small halves, one end of each may be pinned to the base. Both mold halves swing over to form the mold cavity and both are withdrawn from the casting to permit its removal from the base. In order to prevent the shifting of mold halves in relation to the core, it is wise to key the mold halves along the line of their closure—dowel pins insure proper alignment of the mold halves in relation to each other.

Cores extending vertically into the mold cavity from above may be sand or metal, hung by core prints on top of the mold halves and keyed to assure correct positioning. Metal cores may be pinched out with a bar or removed by mechanical means.

Coring of Castings

When the design of the casting calls for coring from the base side of the casting, the core, if sand, may be located in the base by means of a print or, if iron, it may be made an integral part of the base and operated by a rack and pinion. Such a core may be used to eject the casting, or an independently operated ejector plate may be mounted to the base. Small cores or pins through the mold sections are generally actuated with hand toggles.

With a great many castings it is preferable to slide two or more mold segments in different directions to clear the casting rather than to swing them away by means of a simple hinge action. In such cases the various segments slide along the base guided and located by keys and operated by any convenient mechanism.

Where castings are large and heavy, mold segments may be hinged horizontally to the base. The casting cavity is partially cut in the base and the mold segments are pulled away by hand. Cores when large are mechanically operated and the casting freed by means of an elaborate ejector system. The casting may be lifted from the mold by hand or if very heavy by means of special tongs and a quick acting hoist.

The type of mold designed will depend on the particular casting and the specific practice of the foundry concerned.

Making Molds. Molds should be made under close and careful supervision. Patterns for castings for the various mold segments should be held to a tolerance of $\pm$ 1/32 of an inch. The casting cavity should be relieved to retain the advantage of the chilled iron in that part which

makes up the mold cavity. The contour of the back of the mold should follow the general contour of the casting cavity to help unify heat transfer through the body of the mold. The mold segments—bases—large cores, etc., are generally made of a good grade of close grained cast iron. Varying percentages of chromium, nickel and molybdenum may be added to improve the hot properties of the base material, but any well made iron casting—easy to machine—will serve the purpose. Uniformly good results have been obtained from special or high strength irons.

Machining tolerances on the mold cavity should allow no more deviation than ± .002 of an inch. Dimensional tolerances on permanent mold castings are dependent somewhat on size and location in respect to parting. See Table 1. The recommended allowance for shrinkage is .007 of an inch per inch.

Cores. Permanent mold castings often require many different types of cores in the same mold. In some cases a major portion of a casting is formed by a body core. Core pins may be incorporated into a mold to form cored holes or to lighten what would otherwise be a heavy section. Sections which form undercuts so that the mold cannot otherwise be opened away from the casting must be cored by means of sliding cores, collapsible cores, loose pieces, or sand cores.

Metal and Sand Cores

The choice as to whether a metal or sand core is used depends upon many factors and varies with each casting. Metal cores, in general, are cheaper and more accurate and are used whenever it is possible to do so. Sand cores are used when intricate interior construction of a casting makes the use of metal cores impractical.

Since molds are generally made of a high quality fine grained cast iron, the metal cores are usually made of the same material. It is a relatively cheap, readily available, and it stands up satisfactorily. In some applications, as in a long slender core pin, for instance, cold rolled steel is preferable to the cast iron because of its added toughness and ductibility. A non-deforming tool steel may be used if the service requirements warrant it.

Loose pieces and sliding metal cores are used to form undercuts. These are usually preferred over sand cores for such applications, providing they can be readily removed from the casting. The number of loose pieces in any one mold can be limited to the number which can be economically handled during the casting operations.

Collapsible sliding cores may be used in some applications in place of loose pieces. They eliminate the necessity for picking the loose piece out of the ejected casting, thus eliminating the manpower required for this step. The design of the casting, however, often requires a mold in which it is difficult to install collapsible cores. In these cases loose pieces or sand cores must be used.

Metal cores may be removed by means of a pry bar, a rack and pinion, or by some mechanical means such as air or hydraulic cylinder. In other instances, especially in the case of body cores, the mold may be removed first and then the casting is removed from the core by means of ejector pins.

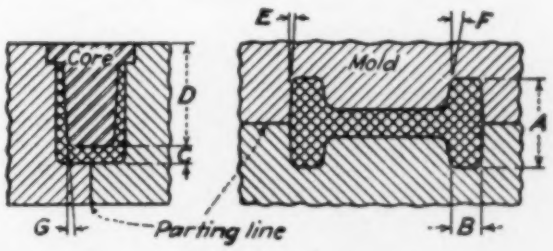
Generally, it is possible to locate metal cores in a mold without much difficulty. A stationary core is, of course, the easiest type to locate. A sliding core such as one which is removed from the casting before opening the mold requires a sufficiently large and accurately machined bearing surface, so that the core is always properly aligned. A loose piece must be made to fit tightly enough so that no molten metal can flow into the crack between the loose piece and the mold and thus increase the difficulties of removing both the casting from the mold and the loose piece from the casting.

Locating Prints

When the design of the casting restricts the use of metal cores, the points to be decided in using sand cores, are locating prints to give suitable support to core, easy escape of gases that may be developed from core binder, removal of chills and wire used to support the core, rapid shaking out of the sand after the alloy has solidified.

Dry sand cores should be made of a highly permeable sand and so made with a surface that permits for

Permanent Mold Tolerances



(A) ACROSS PARTING LINE: ± 1/64 for 1 in. or less—above 1 in. add 0.002 to tolerance per inch of length.
(B) BETWEEN POINTS PRODUCED BY ONE PART OF THE MOLD: ± 1/64 for 1 in. or less—above 1 in. add 0.001 to tolerance per inch of length.
(C) BETWEEN POINTS PRODUCED BY THE CORE AND MOLD: ± 1/64 for 1 in. or less—above 1 in. add 0.002 to tolerance per inch of length.
(D) MAXIMUM LENGTH OF CORE SUPPORTED TO ONE END: Diameter of core x 10.
(E) OUTSIDE DRAFT: 1° minimum—3° desirable.
(F) DRAFT IN RECESSES: 2° minmum—5° desirable.
(G) DRAFT ON CORES: ½° limited—2° desirable.

Note:
(1) Minimum diameter of cores = 1/4 in.
(2) Minimum web thicknesses for distance of:
3 in. or less = 1/8 in.
3 to 6 in. = 5/32 in.
Over 6 in. = 3/16 in.
(3) Machine tool locating points should be indicated.
(4) Allowance for machining:
Castings up to 10 in. long = 1/32 in. minimum, 3/64 in. desirable.
Castings over 10 in. long = 3/64 in. minimum, 1/16 in. desirable.
Surfaces formed by sand cores = 1/16 in. minimum.
(5) Minimum radius of fillet:
Radius should equal average thickness of joining walls.

Table 1—Properties and Applications of Permanent Mold Casting Alloys

Composition Limits

Aluminum-Base Alloys

		Element, per cent								
Alloy	*Aluminum*	*Copper*	*Iron*	*Silicon*	*Manganese*	*Manganese*	*Zinc*	*Nickel*	—Other Elements— *Ea.*	*Total*
C-3	Remainder	7.0±1.0	1.4	4.0	0.07	0.5	2.5	0.3	—	0.5
S-1	Remainder	0.1	0.8	5.0+1.0 −0.5	0.05	0.3	0.3	—	0.05	—
S-2	Remainder	0.4	0.8	5.0+1.0 −0.5	0.05	0.3	0.3	—	—	0.3
CG-1	Remainder	10.0±0.8	1.5	1.0	0.25±0.10	0.3	0.4	0.3	—	0.3
CN-21	Remainder	4.0±0.5	1.0	0.7	1.5±0.3	0.3	0.1	2.0±0.3	0.05	0.15[2]
CS-4	Remainder	4.5±0.5	1.2	2.5±0.5	0.05	0.3	0.3	0.3	—	0.3
SC-1	Remainder	4.5±0.5	1.0	5.5±0.5	0.1	0.3	0.5	—	—	0.5
SC-41	Remainder	1.5±0.5	0.9	12.25±0.75	0.55±0.15	0.75±0.15	0.4	0.05	—	0.5
SG-1	Remainder	0.2	0.6	7.0±0.5	0.3±0.1	0.3	0.3	—	0.05	0.15
SN-41	Remainder	1.0±0.5	1.3	12.0±1.0	1.0±0.3	0.1	0.1	2.5±0.5	0.05	—

Magnesium-Base Alloys

		Element, per cent							
Alloy	*Magnesium*	*Aluminum*	*Manganese (Min.)*	*Zinc*	*Silicon*	*Copper*	*Nickel*	*Iron*	*Other Elements*
A-10	Remainder	10.0±1.0	0.10	0.3	0.3	0.05	0.03	—	0.3
AZ-63	Remainder	6.0±0.7	0.15	3.0±0.5	0.3	0.05	0.03	—	0.3
AZ-92	Remainder	9.0±0.7	0.10	2.0±0.4	0.3	0.05	0.03	—	0.3

[1]See Note 4. [2]Chromium, 0.2 per cent.

NOTE 1—Where single units are shown, these indicate the maximum amounts permitted.

NOTE 2—Analysis shall regularly be made only for the elements specifically mentioned in this table. If, however, the presence of other elements is suspected, or indicated in the course of routing analysis, further analysis shall be made to determine that the total of these other elements is not present in excess of limits specified in the last column of the table.

NOTE 3—The following applies to all specified limits in this table: For purposes of acceptance and rejection, an observed value or a calculated value obtained from analysis should be rounded off to the nearest unit in the last right-hand place of figures used in expressing the specified limit.

NOTE 4—Titanium is permissible in all alloys, but the maximum shall not exceed 0.2 per cent.

Physical Properties

Aluminum-Base Alloys

		—Tensile Strength—		Yield Strength	—Elongation—		Brinell*	Yield Strength	Endurance	Electrical
Alloy	*Temper*	*Min.*	*Typical*	*Typical*	*Min.*	*Typical*	*Hardness*	*Compress*	*Limit*	*Conductivity*
C-3	Ast Cast	25,000	30,000	28,000	†	1.0	80	24,000	9,500	—
S-1	Ast Cast	21,000	24,000	9,000	2.5	9.0	45	9,000	—	—
S-2	Ast Cast	21,000	24,000	9,000	2.0	7.0	45	9,000	—	—
C6-1	Aged	30,000	35,000	31,000	†	1.0	100	31,000	8,500	34
	Solution Quenched Aged	40,000	48,000	36,000	†	0.5	140	36,000	9,000	—
CN-21	Aged	34,000	40,000	34,000	†	1.0	105	34,000	10,500	34
	Solution Quenched Aged	40,000	47,000	42,000	†	0.5	110	46,000	9,500	32
CS-4	Solution Quenched	33,000	40,000	22,000	4.5	10.0	75	22,000	9,500	35
	Solution Quenched Aged	35,000	45,000	33,000	2.0	5.0	90	33,000	10,000	50
SC-1	As Cast	24,000	28,000	16,000	†	2.0	70	16,000	—	37
SC-41	Aged	32,000	36,000	28,000	†	0.5	105	30,000	—	29
	Solution Quenched Aged	36,000	40,000	—	†	0.5	125	—	—	—
S6-1	Solution Quenched Aged	33,000	40,000	27,000	3.0	5.0	90	24,000	—	41
SN-41	Aged	31,000	36,000	28,000	†	0.5	105	30,000	—	29
	Solution Quenched Aged	40,000	47,000	—	†	0.5	125	—	—	—

*500 kg. lead, 10 mm. ball; †not required.

Magnesium-Base Alloys

		—Tensile Strength—		Yield Strength	—Elongation—		Brinell*	Yield Strength	Endurance	Electrical
Alloy	*Temper*	*Min.*	*Typical*	*Typical*	*Min.*	*Typical*	*Hardness*	*Compress*	*Limit*	*Conductivity*
A-10	As Cast	18,000	22,000	13,000	1.0	2.0	54	13,000	10,000	—
	Heat Treated	29,000	33,000	12,000	5.0	8.0	52	12,000	12,000	—
	Heat Treated Aged	29,000	34,000	14,000	†	2.0	69	19,000	10,000	—
AZ-63	As Cast	24,000	29,000	14,000	4.0	6.0	50	14,000	11,000	—
	Heat Treated	34,000	40,000	14,000	7.0	12.0	55	14,000	14,000	—
	Heat Treated Aged	34,000	40,000	19,000	3.0	5.0	73	19,000	13,000	—
AZ-92	As Cast	20,000	24,000	16,000	1.0	2.0	65	16,000	11,000	—
	Heat Treated	34,000	40,000	16,000	6.0	10.0	63	16,000	14,000	—
	Heat Treated Aged	34,000	40,000	23,000	1.0	2.0	84	23,000	13,000	—

*500 kg. lead, 10 mm. ball; †not required.

Supplementary Data

Aluminum-Base Alloys

Alloy	*Specific Gravity*	*Density Lb. Cu. In.*	*Solidification Range, °F.*	*Coefficient** of Expansion 68-212° F.*	*Coefficient** of Expansion 68-595° F.*	*Corrosion Resistance*	*Casting Properties*	*Response to Heat Treatment*	*Machineability*	*Principal Uses*
C-3	2.83	0.102	1165-975	12.2	13.3	Fair	Good	Yes	Excellent	General, automotive cylinder heads, wash machine agitators-vacuum cleaners, etc.
S-1	2.66	0.097	1165-1070	12.2	13.3	Excellent	Good	No	Fair	General with maximum corrosion resistance.
S-2	2.66	0.097	1165-1070	12.2	13.3	Good	Good	No	Fair	General cooking utensils.
CG-1	2.90	0.105	1160-1005	12.2	13.0	Fair	Good	Yes	Excellent	High hardness castings.*
CN-21	2.78	0.101	1165-995	12.5	13.6	Good	Fair	Yes	Good	Aircraft cylinders and pistons. Diesel pistons.
CS-4	2.76	0.100	1160-980	12.2	13.3	Good	Good	Yes	Good	General, where high strength and ductility are required.
SC-1	2.79	0.101	1140-960	11.9	12.7	Fair	Excellent	No	Good	General.
SC-41	2.67	0.097	1095-1000	10.5	11.6	Good	Fair	Yes	Fair	Automotive Pistons.
SG-1	2.65	0.097	1130-1075	11.9	13.0	Excellent	Excellent	Yes	Good	General, where high strength and corrosion resistance are required.
SN-41	2.68	0.097	1095-1000	10.5	11.6	Good	Fair	Yes	Fair	Pistons.

*Pistons, Iron Sole Plates.

Magnesium-Base Alloys

A-10	1.81	0.066	1100	—	16.0	Poor	Good	Yes	Excellent	General.
AZ-63	1.83	0.066	1135	—	16.0	Fair	Fair	Yes	Excellent	General.
AZ-92	1.82	0.066	1110	—	16.0	Fair	Fair	Yes	Excellent	General, where maximum strength and pressure tightness are required.

**Multiply by 10^{-6}.

reasonable handling. The core binder must be of low gas forming material and of such a nature that it will soften when the hot alloy comes in contact with cores and will never set so firmly that it will cause trouble in removing the core from the casting.

Cores for magnesium base—alloys should contain sufficient preventive agent to keep the alloy from burning or becoming pitted due to formation of oxide.

A rub jib is often required to obtain an accurate location surface on a sand core in order to correctly locate the core in the mold.

Pin vents, such as a small round hole containing a square pin, are convenient to use on a metal core. This also must be vented through the back of the mold. Porous emery stone is sometimes used for vents, although the disadvantages of this material are that the pores eventually become plugged with mold wash, and the contour of the mold must be ground into the surface of the emery. Ejector pins located in stationary cores provide more venting than is ofentimes accredited to them.

Venting. Unfortunately iron molds do not have the permeability of sand molds. Therefore, venting is ever a problem with a permanent mold foundry. Vents are used as a means of exhausting air from mold cavities which in turn tend to sharpen detail in castings.

Types of Vents

There are two specific types of vents, one made by the parting of the mold and by core parts, which are natural vents and cause no additional marks on the castings; and plug vents which are placed on surfaces remote from parting lines and are so placed that air will escape through the back of a mold.

Holes, too small to allow metal to flow through but large enough to allow gases to escape, are filed or drilled in mold sections where they will not affect the casting contour or surface. Vent lines not over 0.010 of an inch deep connect the mold cavity to holes which lead into the open. Risers are cut open to furnish exits for air in the mold cavity as well as a guide for stopping the pour. Molds in operation are tilted and so poured as to provide maximum venting. The angle of tilt will provide a natural vent at certain points by reducing the distance of an overhanging shoulder or boss. Tilting should be tried before cutting up a mold with vent plugs.

When castings contain large flat areas, generally very difficult to pour in a permanent mold, extra venting must be supplied for such sections. This is done by judiciously placing plugs over the area in question and filing numerous vent lines on the circumference of each plug to permit the escape of air from these sections. The sizes of vent plugs are determined by the area showing mis-runs. Remote heavy sections are also vent plugged or chilled and vent plugged to relieve shrinkage and to allow the voiding of air which might otherwise be trapped and create enough back pressure to prevent filling of the mold cavity. Ejectors and core pins are also used to give advantageous venting.

The sand cores of a semi-permanent mold casting are themselves elaborately cored and criss-crossed with vent lines to permit the free passage of air and core gases. No permanent or semi-permanent mold can be vented too much.

Gates, Runners and Risers

Provisions of runners and gates for a casting may be compared to an assembly system. In the construction of a component, the various parts are brought together at the right time and joined carefully. Passing on, stage by stage, for each successive addition, until after a series of intricate and well planned operations, the fully assembled unit is produced. Everyone with a background of planning is aware that "bottle-necks" in an assembly line must be avoided. On the same basis, a skilled permanent mold designer avoids "bottle-necks" in running and gating.

In designing for light metals, he recognizes the fact that light metals are lazy and have to be helped to flow into the mold cavity. Hence light metal molding is characterized by the provision of numerous wide feeders and comparatively heavy risers, with a very definite tendency to counteract defects produced by incorrect feeding and gating with heavy risers.

The permanent mold designer's goal is to make certain that gates and feeders are large enough and so placed as to avoid cold shuts and undue agitation and to supply ample flow of metal to fill the mold cavity at a satisfactory rate; that no one runner or section of a runner will be made to serve the purpose of two or alternatively will be made to contribute less than its share. The design, size and position of gates and runners must be such as to achieve maximum uniformity of heat distribution so that the solidification in the poured casting will be progressive.

The mold cavity should be so laid out that the heavy sections of the casting are at the parting, whenever possible, so that heavy sections may be fed directly by gate and feeder, or if their positioning be remote from the gate that proper risering be resorted to. It is preferable to have the casting in a permanent mold in a vertical position except for the tilt of the mold. However, should another pouring position offer operating or other advantages, the designer will not hesitate to lay out the mold so as to gain the benefits so derived.

In the design of the mold segments sufficient room should be provided around the cavity for alterations in size and position of gates, runners and risers. There must be sufficient height of gate and feeder so that enough liquid head is provided to insure filling all parts of the mold cavity. Should the mold as designed provide too much head, it may be cut away; if the head be insufficient, pouring blocks must be attached to the mold parts.

Preventing Metal Leakage

There should also be enough room on the sides of the mold faces to allow for gates and feeders and sufficient flat surface to make a good seal to prevent metal leakage at the time of pouring. Small and medium sized castings are filled through one gate from a single ladle. Larger castings may require as many as four gates, each fed simultaneously from a separate ladle.

Pour Control. Efforts have long been made to eliminate the human element factor in the pouring of a permanent mold casting. When the pour is too hard, the metal is agitated and likely to trap and carry air and dross over into the mold

cavity; when the pour is too easy, misruns are prevalent (Fig. 1). Control may be partially attained as follows:

1. By putting a sump at the base of the gate. Such a device will not in itself be fully effective but will help decrease possible turbulence;

2. By making the volume of the feeder about one-third greater than the volume of the gate, thus slowing the rate of flow into the cavity;

3. By drastically choking the stream of metal from gate to runner. Either by means of multiple fingers, at various levels, tapering from runner to gate and so placed as to flow metal upward from gate to feeder, or by means of a fan slit arrangement tapering from the gate to an opening slightly more than the thickness of the casting at the feeder;

4. Or a gooseneck curve in the gate, supplemented by a fan, may be used as a method to control pour. The tapered fan slit from gate to runner contributes to positive pour control but places almost the entire burden of making sound castings on the shoulders of the designer, for the gate, riser, feeder, etc., must be right to produce sound castings.

This is a difficult problem, especially in castings where the sections vary drastically in mass. In castings with changing sections, it would be ideal to vary the rate of flow of metal into the mold cavity with the changing mass of the casting. When metal feeds through a tapered fan slit no changes can be made in the rate of flow into the mold cavity because the slit affords a positive choke and no casting can be poured too "hard" unless the slit is too open.

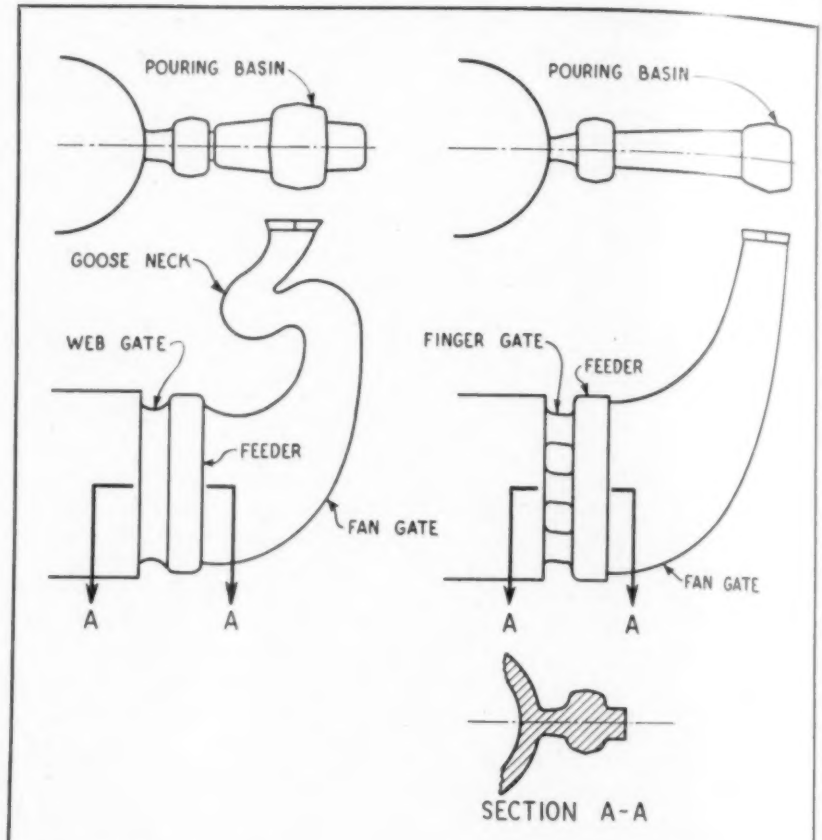


Fig. 1—Drawing illustrating nomenclature used to describe pour control gating of passages leading to or from a permanent mold.

Rate of Pour

Thus scrap due to blow holes and castings containing hard spots are kept at a minimum. With a more open connection between gate and runner a good mold operator can determine at what rate to pour different sections of any castings after he has made a few trial pours.

When multiple tapered fingers are used to connect gate to runner, the operator has partial control over the pour and may vary it to suit the varying mold sections. The connecting fingers are also so located as to feed massive sections more readily, and eliminate any necessity for change in the rate of pour. With this set-up, fewer obviously scrap castings will be made but it must be remembered that the pourer has only partial control over possible turbulence and its resultant defects.

From the gate, the metal filling the mold cavity flowers into a side runner or riser, or, in some cases of horizontally placed cavities, into a riser directly over the casting. The side feeder or runner should be about one-third larger in cross-sectional area than the gate.

The purpose of this is, (a) to slow down the rate of metal flow and thereby reduce possible turbulence, (b) to make the flow of metal into that part of the mold cavity which forms the casting easy and uniform, and (c) to feed the casting, offsetting crystallization and contraction shrinkage. The feeder should be shorter on both bottom and top than the gate, but provisions should be made at the bottom for a small sump and at the top for a small pressure ball.

A slit extending almost the length of the runner acts as the inlet for metal from runner to casting cavity. Only the sump and pressure ball are not tied into the casting. The width of this slit inlet should approach the thickness of the casting at the point it ties in. About 0.010 in., generally less, should be left to serve as a witness mark for use as a guide when trimming the gate off the casting.

When the casting is filled directly from the riser, the same general procedure is followed, the inlet increasing or decreasing in cross-section with changes in thickness of sections of casting being fed. Slit inlets are used in order that hot metal will be supplied to solidifying zones to decrease the tendency for porosity due to too long solidification periods which would be produced with the hotter metal or locally over-heated mold sections.

There is generally the tendency to carry the too large solid risers used in sand casting over into permanent mold designs. Such large risers, with their long solidification time, slow down production, cause

local over-heating of the mold and are responsible for draws and shrinks in adjacent areas. In theory, little more metal is required in a riser for feeding, than that necessary to compensate for contraction in the volume of the metal in the section being fed by it.

Incidence of unwanted draws and porous areas due to too long solidification time can be greatly reduced if design is such as to give even solidification from the bottom upward. Risers are best used and considered as devices to supply metal to make up crystallization and contraction shrinkage, rather than as arrangements to force metal down into the body of the casting in order to hold up a thick section by sheer fluid head. As stated before, riser sections should follow the general contour of the casting sections which they feed with the inlets opened as much as possible, leaving only a witness mark for trimming.

Wall Thickness. It was formerly considered essential to make mold sections as heavy as possible because increased sturdiness in construction supposedly minimized warpage and cracking in mold segments. This meant increased operating difficulties due to exceedingly cumbersome molds. Another notion was that even though a long period of time was required to bring the heavy molds up to operating temperatures, once reached this temperature would stay more or less constant.

Thermal Conditions

Present experience contradicts those old ideas. Uniform temperatures throughout the mold are not always necessary. More often a temperature gradient across certain sections of the mold gives the correct thermal conditions for the desired rate of solidification of the casting.

It is common practice to apply external heat to that part of the mold which forms the minimum section of the casting and may consequently receive too little heat to maintain the desired temperature in that part. It is more feasible, and from an operating standpoint preferable, to operate a lighter weight mold without the necessity for using external heat, and to depend on the heat given up by the metal during solidification to make up the losses due to convection and conduction. In extreme cases, it may even be desirable to cool very heavy casting sections which may be in too light mold sections by impinging a stream of air at the affected spot or by holding waste periodically wetted with water at such a section.

There is no hard and fast rule for determining proper mold thickness. Some advocates of thin mold sections have suggested wall thickness of about three times the thickness of the casting. Permanent mold castings are commonly made in thickness of 0.080-0.090 in. This would indicate a mold wall thickness of 0.300 in. While such a light mold may be extremely fast, it probably would not have a very long life.

Molds of about twice that thickness, ⅝ in., have been built and have proven to be mechanically weak. Cracking at changes in mold section was quite prevalent even when all possible precautions for its preventions were taken. When used for making castings 0.125 in. thick, such molds were also found to be very sensitive to temperature fluctuations of either metal or mold.

Mold Cracking

When new molds with a wall thickness of ¾ to ⅞ in. were made for the above parts, the tendency to crack and warp was found to decrease. The sensitivity to temperature changes had also decreased but the molds were somewhat slower than those with the thinner wall section.

Nor can a mold thickness of three times the casting section be accepted for heavy castings, say ¾ to 1 in. in cross section. The resultant molds would be very massive, difficult to operate from a mechanical standpoint and difficult to control mold operating temperature. It must be remembered that in all molds, the supports and various operating sections in contact with the mold all carry away their share of the heat from the mold segments.

Therefore, mold segments need be no heavier than required to maintain the required temperature gradients. In general, mold sections need be no heavier than 1½ to 2 in. or slightly more for extremely heavy castings. It is good practice to have the mold contour follow roughly that of the casting to be made in it, so that the transfer of heat from metal to mold be on the basis of equivalent sections.

As in sand castings, but far more sparingly, chills find valuable application in securing an even rate of solidification and in influencing its direction. In permanent molds, chills are also pressed into service as vents as well as acting the conventional role of substitute for risers at inaccessible heavy casting sections.

The degree of chilling is controlled by the material and the length of the insert used for this purpose. Chills are generally copper, however, steel may be used when only small increases in chilling are desired. With venting always a problem in permanent molding, advantage is taken of the presence of all chills, and vent lines are invariably filed on the sides of the chill inserts. The application of chills can only be stated in general terms as the foundryman alone, with his skill and experience, can decide on the precise application of chills for each particular job.

Mold Preparation and Painting. A phase of permanent molding procedure passed over rather lightly by the uninitiated is the part of mold painting. This phase of the process is really an art. Skill at the application and trimming of mold paint requires long practice before satisfactory results can be assured.

There are several proprietary paints on the market and every permanent mold foundry has its own preferred formula. Basically all mold paints contain a refractory such as whiting, French chalk or a similar material, and a binder such as water glass, both suspended in water for easy application with a spray gun. A lubricant such as rouge or graphite may be added to aid the parting of mold segment and casting.

Mold Paint Formula

One paint which has been used successfully in production over a period of years is made up as follows: About 30 gallons of water in an appropriate container is brought to a boil. Seven quarts of 40 per cent water glass are added and stirred into the boiling water. Then 10 lb. of whiting are added and thoroughly dispersed through the boiling mixture and finally one pound of rouge or graphite is also

added to the boiling material. The suspension is boiled for a few minutes longer, allowed to cool to room temperature and is then ready for use. The solid ingredients tend to settle out on standing. Therefore the paint should be well stirred before it is ladled into a spray gun prior to application.

A paint of the above type is used on mold faces, cores, gates, runners, risers, etc.—all sections that go to make up the mold cavity. Some cores and core pins that have a long stroke, in contact with the newly set casting, generally require more lubrication. This is supplied by water suspension of graphite applied periodically, as required, with a spray gun or after each cast by dipping the part into the graphite water suspension. It must be noted that such a coating will differ in thermal characteristics from the one produced by the above mentioned paint.

Prior to painting, molds are cleaned and heated to about operation temperature which is in the neighborhood of 700° F. When a new mold comes out of the machine shop, an individual pipe burner should be made for it. The burner should be so constructed that the entire mold is uniformly heated and distortion due to unequal expansion is kept at a minimum. The various mold segments, which have been previously stress relieved after rough machining, are assembled.

Pre-Heating of Molds

Using the newly constructed burner, another long, rough normalizing treatment should be given to the assembled mold. During this treatment all mold faces should be tightly clamped together—the cores are left loose and the heat applied from the outside is gradually increased to the desired temperature and held at temperature for a few hours or days depending on the size of the mold. By such treatment possibilities of cracking and warpage of the mold are kept at a minimum. Should any distortion or warpage take place during the treatment the mold parts are refitted.

The hot mold is then disassembled for painting. The sections to be painted are slightly roughened by a light sand blast in order to make the paint stick better. A light coat of paint is sprayed over all parts of the mold which are to be covered. This coat is rubbed down with steel wool.

Painting and rubbing down are continued until a tightly adhering layer of the proper thickness has been applied. The average thickness is around 0.002 in.—lighter or heavier in different sections depending on the chilling effect desired.

A skilled painter will finish a job with two or three coats and finish all the gates, runners and risers which require a much heavier coat by applying additional paint to these areas with a hand brush. The thickness of the paint in critical sections is trimmed by scraping to the proper thickness and is finally smoothed over to leave no marks on the casting. The mold sections are then reassembled, the mold closed and reheated to operating temperature with its burner.

A good paint job should last through 24 hr., more or less, of continuous production, depending somewhat on the number of pours, the size of the casting, the rate of operations, etc. It is generally necessary to make minor repairs and trims on the paint intermittently through the production period.

Mold Paint Property

The painted surface should not be too smooth. Among the properties of the paint which are taken advantage of are its ability to enable the metal to lie quietly against the mold face, to help control the solidification of the casting, and to protect mold sections against erosion by molten metal.

The modification of paint in thickness or type will influence the rate of solidification of the casting. The thickening of the refractory paint in certain areas will delay solidification at that point appreciably. Hence, gates, runners, risers, thin areas of the casting are among the sections which are given a heavier than normal coat of paint so that heat is retained and thus other sections of the casting fed more effectively. At sharp changes in section, it is customary to remove the paint altogether and so provide chilling.

Elaborately trimmed paint jobs should be avoided because of difficulties in exact duplication. It is wiser to study possibilities thoroughly and obtain control of progressive solidification by other means, such as gate or riser changes, by application of heating or cooling to the affected area, or whatever other steps may be effective, rather than to try to obtain this control only through paint.

On a mold in production, the paint wears progressively smoother and becomes less and less effective, until misrun castings become quite common. When this point is reached, it is advisable to remove the old paint with a wire brush or a light sand blast, and then reheat and repaint the mold as previously outlined.

It is not good practice to bring a cold mold previously painted, from room temperature up to operating temperature—unless only sample castings are to be made. The difference in the expansions of the paint and the mold will cause the paint to crack or pull away from the mold in spots as operating temperature is being reached. Results under such conditions will not be satisfactory.

Melting Practice. A sound melting practice consistent with the economics of the production of good quality aluminum alloy and magnesium alloy permanent mold castings cannot be stressed too highly. A poor or haphazard melting practice can defeat the best of permanent mold design.

In considering furnaces for melting aluminum and magnesium alloys, it is necessary to have in mind the nature of the alloy; make up of the melting charge; rate of output required; operating conditions; dross loss and fuel consumption. Where the output of light alloys is only a small part of the total production, the melting charge may be made directly in the casting room holding furnace.

Melting of Alloys

In foundries where a large rate of output of molten alloy is required the alloys are melted in a battery of tilting furnaces with capacities of 250 lb. or more. This is done to cut the time period of melting as short as possible; to prevent excessive oxidation, gas absorption, iron pick up and waste of fuel; also to insure a continuous supply of molten alloy to the casting room.

Where large tonnages of one aluminum alloy are consumed, melt-

ing may also be done in a large reverberatory furnace. The molten alloy is transferred from the tilting furnaces to a bull ladle or crucible with suitable capacity and carried in buggies or tram rail for larger amounts to stationary holding furnaces which may range in capacity from 250 to 600 lb. in the casting room.

Aluminum Alloys. Iron pots are generally used for melting aluminum alloys. A suitable coating or wash is applied to the inside of the pot so as to prevent iron pick up from the molten aluminum alloys. For melting the aluminum-magnesium and aluminum-silicon alloys which have a marked tendency to pick up iron from the melt, it is advisable to use crucibles.

Small reverberatory furnaces 2,000-3,000 lb. in capacity are another commonly used melting medium. These units may be charged through a door or well on the back of the furnace and molten metal may be ladled out of one or two wells on the front end. Ingot and clean scrap are charged at about the same rate as molten metal is being ladled out. This rate should be such that the automatically controlled casting temperature does not fluctuate during the time the unit is in production.

Furnace Fuels

Furnaces may be heated with fuel oil, gas or electricity. The choice of which will be dictated by economic reasons rather than by any marked technical superiority of one or the other for the purpose of melting.

Light alloys should be heated only to the minimum temperature which will permit of its going through the necessary handling, without getting below the proper pouring temperature. The proper pouring temperature for casting in any alloy is the lowest temperature consistent with running any particular casting when the mold is properly gated, vented, chilled and heated. It is also possible to go to the other extreme of pouring the alloy too cold, causing cold shuts, trapped air and lack of sharpness in resultant casting. Pouring temperatures most generally range between 1250° to 1500° F. depending on type of casting poured.

In modern foundries, aluminum and magnesium alloys are melted under automatic temperature control. Satisfactory control can be obtained with a portable pyrometer and thermocouple by checking the temperature several times during a heat. Another and more consistent means of temperature control may consist of a thermocouple suspended over each furnace and connected to a main pyrometer control board.

Fluxes are not generally used when the alloy is properly handled. Only clean alloys should be used and they should be kept clean throughout the process of handling. Fluxing is not a substitute for proper remelting conditions but if alloys have been melted under inadequate control or a large amount of dirty foundry scrap has been added to the melt, fluxing is helpful. There are a number of fluxes to assist in deoxidizing and degassing aluminum alloys. In general, the powdered fluxes are employed to clean or deoxidize the melt, whereas the gaseous fluxes are used not only to clean but also to degas the molten alloy.

Flux Constituents

The fluxes may consist of aluminum chloride alone or a mixture of calcium fluoride and sodium chloride. These salt fluxes are stirred dry into the molten charge. A few minutes is allowed for the slag and oxide to appear on the surface. The bath is then skimmed. Gas fluxes such as nitrogen and chlorine, if used are bubbled into the molten alloy through a refractory or a coated iron pipe nearly reaching the bottom of the melt. Solid or gaseous fluxes must be introduced into the molten alloy absolutely dry.

During the casting operation the molten aluminum alloys are uncovered and subjected to continual oxidation by the atmosphere, hence the caster should sweep aside with the bottom of the hand ladle the oxide film formed, and very carefully insert the hand ladle for quiet removal of the alloy from the pot, taking every precaution that dross will not enter the mold cavity.

The pouring should be rapid enough to prevent cold shuts and misruns and at the same time slow enough to allow the casting to solidify progressively. The stream between ladle and sprue should be short as possible, and at no time should the stream be broken. This will avoid entrapment of air, agitation of the molten alloy and inclusion of oxide film. In order to aid in preventing cracks, the casting should be removed from the mold as soon as it has set.

At the end of the day or after each shift, pots, ladles, skimmers and other equipment coming in contact with molten alloys should be thoroughly cleaned and a protective coating applied. A mixture of four parts whiting, two parts sodium silicate and ten parts water is satisfactory when a thin coat is applied. Iron pots should be thoroughly scraped to remove any loosely adhering pot scale. This will minimize the occurrence of so-called hard spots in the resultant casting.

Melting Pot Washes

The wash may be applied to pots at such a temperature that the wash neither boils off nor remains wet on the pot surface. The use of wash should, of course, be applied to all iron equipment coming in contact with the molten alloys as a protection against iron pick up. Pots should be inspected at least once a day for cracks or breaks beginning to appear which might lead to quick failure. The damage done by a break out may be extremely serious, since it will not only cause the loss of a heat but it may also injure the furnace and the furnace tender or caster. It is not common practice to permit the molten alloy to freeze in the bottom of the pot because of possible damage to the pot.

Another point that cannot be stressed too strongly is the establishment of a system to segregate scrap. This applies especially to jobbing foundries where a variety of alloys are used. It is an easy matter to contaminate an alloy either through mixing scrap or by melting or pouring from equipment which had been used previously to melt another alloy and had not been thoroughly cleaned prior to the change in alloy.

Magnesium Alloys. Magnesium permanent mold casting alloys are most generally melted in steel pots free from metals for which magnesium has an affinity. Chromium, vanadium and related metals are practically insoluble in magnesium, and hence alloy steels containing these elements may be used safely

for pots to melt magnesium alloys.

Because magnesium has a tendency to burn and absorb nitrogen from the atmosphere rapidly when molten, it is necessary to keep the molten charge covered at all times with a flux that has good covering power. Each melting method requires a particular type of flux which may be procured from the producer of the alloy with which the flux is to be used.

When melting in a tilting furnace some of the fluxes lack the necessary covering properties, thus in order to insure exclusion of the air from the surface of the metal, a flow of dry sulphur dioxide gas is used in combination with the flux to prevent oxidation of the magnesium alloy.

Magnesium Melting Furnaces

Tilting furnaces for melting magnesium alloys are designed with hinged covers that can be tilted back to permit cleaning and changing of the pots. The pots differ from those used for aluminum in that a baffle or pipe is welded along the side of the pot at the spout. The well thus formed extends to within a few inches of the bottom.

When metal is poured from such a pot, care should be exercised that the molten metal is not drained below the bottom of the well. The above arrangement is used to maintain the flux cover over the molten metal. The remaining molten heel will also facilitate melting down the next charge.

In filling the hand ladle from this type of furnace, the tilting may be accomplished by a pneumatic motor actuated by a foot control pedal. This leaves the caster's hands free to powder the stream with sulphur in order to produce an atmosphere of sulphur dioxide gas to prevent oxidation.

After each heat, while the pot is still hot, it is thoroughly scraped to remove any loosely adhering material. During the cleaning operation flux is sprinkled on the molten heel to prevent burning. After the settlings and dress are removed from the surface of the molten heel, the pot is ready for the next charge. In magnesium permanent mold practice the charges are most generally made directly in the casting room holding furnaces.

In the stationary open pot furnace method of melting, the most widely used flux for cleansing and protecting the magnesium alloy from oxidation consists of 60 per cent anhydrous magnesium chloride and 40 per cent sodium chloride. The ratio of flux used is 1 to 1¼ lb. of flux to 30 lb. of alloy depending on the amount of foundry scrap charged with the ingot. When the flux is melted it forms a liquid bath for the molten alloy. After the alloy has melted, the charge is thoroughly stirred to put the flux in action to wash the alloy free of impurities and allowed to stand and cool till the desired casting temperature is reached.

To take metal for pouring, the liquid flux is pushed aside with the bottom of the hand ladle. As soon as the ladle full of metal is withdrawn, the flux re-forms a cover over the area from which the ladle was withdrawn. Immediately on removal of the alloy from the pot, the surface of the alloy in the ladle is powdered with sulphur to prevent oxidation.

It is not necessary to clean the melting pot in a stationary furnace after every heat providing the charge is kept at a level that will permit removal of magnesium alloy without disturbing the settlings. At intervals the settlings in the bottom of the pot are removed with a strain spoon.

Any flux or alloy added to the molten heel or charge must be perfectly dry. The same applies to all equipment coming in contact with molten metal. Ladles are most generally preheated by placing same on the projecting part of the furnace to drive off moisture.

Grain Refinement

To refine the structure, the molten alloy is subjected to a temperature between 1600° F. to 1650° F. for 10 min. and allowed to stand and cool to the desired casting temperature is reached. Casting temperatures range from 1240° to 1400° F. depending on type of casting poured.

No washes are used in magnesium melting practice as with aluminum.

At all times the molten magnesium alloy should be under strict temperature control and sufficient flux on the surface leaving no part of the molten alloy exposed to the air or atmosphere.

Temperature Control of Molds. In order that the permanent mold casting process be economically feasible, it must be capable of consistently producing castings of a uniform and satisfactory quality. An important step in this process is the control of the mold temperature within a definite range during casting operation. It is unlikely that all sections of a permanent mold would operate at the same temperature, and it is also unlikely that different molds would operate within the same temperature range.

There are probably limiting maximum and minimum temperatures between which a mold will operate successfully. It may be necessary to apply external heating or cooling to maintain the mold at equilibrium within the limiting temperature range. Each mold has its own temperature range/time cycle and it is necessary to determine experimentally the correct operating range for each mold and to hold all the conditions within a definite range so that the castings will be of the desired quality.

Mold Temperature

Mold temperature is controlled through the accurate control of the pouring temperature of the metal being cast, by the casting cycle, by the thickness and trim of the coat of refractory paint, by the thickness of the mold sections, and by the application of external heating or cooling.

In order to maintain the mold temperature within the correct range, the casting cycle and pouring temperature should be held constant. The pouring temperature can be controlled either manually or automatically. It is possible to control the cycle time either by means of timers or by removing the casting at the same temperature or degree of solidification each time. Since the operator usually goes through the same mold preparation procedure each time before casting, it is relatively easy to maintain a uniform casting cycle.

A contact or surface pyrometer provides a rapid and easy method for measuring the temperature of the actual casting surface of the mold. Its disadvantage is that the temperature can only be determined

while the casting is removed from the mold. A thermo-couple inserted into a hole in the back of the mold can be used to indicate temperature variations during the casting cycle. Care must be taken, however, to prevent drilling the hole too close to the casting surface, as this may cause local overheating of the mold.

Molds will vary in temperature during operation between those areas adjacent to thin sections of the casting. Variations in mold temperature and, hence, in the casting solidification rate must be properly controlled in order that progressive solidification may take place. Five things are commonly done to effect mold temperature differentials. In order of decreasing desirability, they are:

1. Proper design of gating and risering to place the hot metal in the most advantageous positions.
2. Mold section thickness may be varied in order to provide more or less mold volume for proper dissipation of heat.
3. Mold wash thickness may be varied in order to provide more or less insulation in various sections of the mold.
4. Application of additional heating or cooling to the mold to correct for conditions unattainable by the other three methods.
5. Use of chilling fins; chilling of the mold may be accomplished in several ways. Fins can be attached to the back or sides of the mold of uniform or varying lengths to provide various degrees of chill.

Mold Chilling Mediums

Air can be used as a chilling medium by blowing it across fins, by directing the flow against a certain section of the mold, or by drilling holes into the mold for the entrance and exit of a stream of forced air.

Water may be used to obtain extreme degrees of chilling. It can be fed through a channel in the mold similar to the method used for air chilling. Another method would be to use the heat absorbed by boiling the water as a chill, condensing the steam and re-using the same water indefinitely.

Finishing Aluminum. Gates and risers are usually removed from the castings by means of a band saw of suitable tooth design and operated at speeds of from 500 to 1200 or more ft. per min., or by sprue cutters or abrasive cut-off wheels. It is often feasible to make a fixture for a particular casting and remove gate and riser and also finish in one operation.

The final finishing of castings is generally performed by grinding wheels, disk grinders, and files. Sand blasting by air or by water and sand is also employed both as a means of cleaning and as a final finish, particularly for architectural castings. The degree of finish is largely determined by the use to which the casting is put.

Welding of Castings

Welding by approved methods may be done with the oxy-acetylene torch or electric arc. Welding of other than slight surface defects such as sand holes and misruns is not recommended. The welding of heat treated castings is also not recommended unless they are subsequently reheat-treated.

Large castings should preferably be preheated in a furnace before welding to minimize the danger of cracking. They should likewise be cooled slowly to room temperature after welding. Small castings can usually be preheated satisfactorily with a torch. The aluminum-silicon alloys are relatively easy to weld, being practically free from the tendency to crack except when the casting is very intricate or extreme variations in section thickness exist.

The aluminum-copper alloys are more susceptible to cracking when welded owing to "hot shortness." The aluminum-magnesium alloys are extremely difficult to weld satisfactorily because of dross formed by oxidation of the molten alloy. Radiographic inspection may be used to check the quality of welds, as well as to check gating and risering.

Ornamental Finishing

Aluminum alloy castings lend themselves to practically all methods of ornamental finishing. Such mechanical finishes are produced by polishing, buffing, or satin finishing are readily applied with polishing equipment of the same type as used with other metals. Abrasive wheels made up of sewed muslin buffs 6- to 12-in. in diameter to which emery of the desired size is glued are generally employed, operating at speeds of 5000 to 7000 ft. per min. For higher polishes, muslin or felt wheels and a polishing material such as tripoli or greaseless compounds may be used. Other desirable finishes or combinations of finishes may be applied with circular wire or fiber brushes or hand rubbing.

For certain applications, electroplated finishes are desired. Such plated finishes as nickel, copper, zinc, and chromium are as readily applied to aluminum alloys, providing certain simple precautions are observed. Care should be taken to select the type of alloy and plate best suited to the particular job. Generally speaking, electroplated coatings find their best possibilities for parts not subject to continuous outdoor exposure.

The electrolytic process of oxidation or anodic treatment is often used as a means of finishing certain aluminum alloy castings. Several methods for producing such electrochemical coatings are in use, being differentiated by the electrolyte used. Otherwise, the processes are quite similar, in that the article to be coated is made the anode in a bath of specified composition. The oxide coatings so produced possess, in addition to their protective value, remarkable abrasion resistance, dielectric characteristics, and electrical insulating properties. Certain types of coatings are very absorbent and can be readily impregnated by dies and mineral pigments both for color effects and to increase the corrosion resisting properties.

Rough Trimming

Finishing Magnesium. Castings are rough-trimmed to remove sprues, gates, and risers; sand blasted; and given a first inspection prior to heat treatment. Final clean-up generally follows the heat treating operations. Band saws, made from 20-gauge spring temper steel with 4 teeth per inch, are operated at a speed of approximately 4000 ft. per min. Power hack saws should have 10 teeth per in. and hand hack saws 14 teeth per in.

Single-cut files are recommended and the type known as "body worker's" files are quite satisfactory for heavy work. Portable electric or pneumatic chippers, rotary files, and grinders are used to good advantage. Medium-hard grinding wheels (K to

M for alundum) are recommended. A grain size of approximately 20 is satisfactory for foundry snagging operations, while a grain size of 30 to 46 is best for more accurate work.

Tools used on magnesium alloys should not be used for other metals to avoid the danger of sparks. This is particularly true with grinders due to the inflammability of the powder. Special precautions should, therefore, be taken to see that all skim gates and core wires are removed from the castings before grinding and that the wheels do not touch steel inserts. Operators' clothing should be of smooth cloth with no pockets or cuffs in which dust can collect.

Fire Risk

The grinding wheels and the floor around them should likewise be kept free from dust, preferably through the use of a suction exhaust system. If proper precautions are observed, the fire risk is small and if ignition should occur, the flame will be confined to a minimum of material. Regular molding sand will quickly extinguish any fire that does start. Graphite, table salt, or cast iron borings should be used in case of fire in the machine shop or around equipment that might be damaged by the abrasiveness of sand.

Welding and soldering of castings are not as common practices in the magnesium foundry as with some other metals, although these operations may be used to repair small surface defects in unstressed sections of certain types of castings. In acetylene welding, the casting should be heated to 600 to 750° F. The filler rod should be of the same general composition as the casting and a flux should be used. The completed welds should be thoroughly cleaned in the chrome-pickle solution (see below) to remove any adhering flux.

Soldering Castings

Castings to be soldered should be thoroughly cleaned by wire-brushing, chipping, or filing so as to present a clean surface to which the solder can bond. They are then preheated to 300 to 400° F. and the solder vigorously puddled with a sharp-pointed steel tool until it has alloyed with the magnesium. Suitable soldering materials include 90 per cent cadmium-10 per cent zinc or 60 per cent cadmium-30 per cent zinc-10 per cent tin. The former is less brittle but requires a somewhat higher temperature.

Painting is the most practical surface finish for magnesium castings where appearance and protection are important. Chemical finishes have been developed but they are more important as preliminary treatments for painting than as decorative coatings.

The successful painting of magnesium alloy castings offers no unusual difficulties where approved materials are used and good painting practice is observed. Oxide films and dirt are generally removed mechanically by sandblasting or wire brushing. Oil and grease are removed by washing with a solvent, such as carbon tetrachloride or gasoline, or are removed by treatment in hot alkaline cleaners. Cleaned castings are then given the chrome pickle treatment. This consists of a 1/12 to 12-min. dip in an aqueous solution containing 1.5 lb. of sodium bichromate and 1.5 pints of concentrated nitric acid per gallon of solution.

Chrome Pickling Castings

This bath becomes depleted on use but its efficiency can be restored by proper additions of bichromate and acid. The bath should be discarded after it has been rejuvenated four or five times. Upon removal from this bath, the castings are allowed to drain for a few seconds and then thoroughly washed in hot water. This treatment improves the corrosion resistance of the metal and gives an excellent "tooth" which increases the adhesion of subsequent paint coatings.

The selection of the primer is very important in view of the fact that all representatives of a given class of paints do not show the same service performance. Vehicles with maximum adhesion and maximum imperviousness to water should be selected where service conditions are severe.

Baking primers and finishes have excellent adhesion and are, therefore, recommended for small castings or wherever they can be applied. Zinc chromate primers meeting Navy Specification P-27 should be used on large castings in service near salt water or continuous moist air. The latter constitute the best all around primers and should be used where the type of service is unknown.

The choice of surface and finish coats is likewise important although perhaps not so exacting. A good quality oil paint or enamel is satisfactory for interior work while full synthetic and oil-base synthetic enamels represent the better class of products for outdoor service or near the sea coast.

Heat Treatment of Aluminum. The mechanical properties of many aluminum alloys may be materially improved by a suitable heat treatment. The type of heat treatment employed is determined in most cases by the properties desired in the castings.

Solution Heat Treatment

As a general rule, the heat treatment consists of two steps; namely, the solution treatment, which is carried out in the range of 900 to 1000° F., followed by a quench, with an accelerated age or precipitation heat treatment at from 300 to 600° F. The solution treatment is preferably done in electric- or radiant-tube-heated furnaces, since it has been found that slight traces of sulphur and moisture in the furnace atmosphere will materially affect castings of some of the alloys, causing intergranular embrittlement.

Circulatory fans are desirable in the furnaces to provide a uniform distribution of the heated air. The quench which follows the solution heat treatment is usually in boiling water in order to minimize internal stresses. An air-blast quench may be used under certain conditions, but will usually result in a lowering of the mechanical properties. In the case of large intricate castings having thick and thin sections, it may be necessary to eliminate the solution treatment entirely and use only a precipitation treatment.

Aging Heat Treatment

Castings of most heat treated alloys, after solution treatment, undergo a room temperature aging reaction, resulting in an increase in tensile strength, yield strength, and hardness, with a decrease in ductility. This aging proceeds rapidly at first, and, for all practical purposes, has sufficiently progressed after 3 or 4 days to give expected solution treated properties.

Complete aging may be accelerated by a precipitation heat treatment. This may be done in either an electric or gas-fired oven. By proper selection of temperature and time, various degrees of hardness can be obtained which pass through a maximum and, if carried sufficiently far, approach the hardness of the annealed material.

Material which has been carried past the maximum hardness is frequently referred to as "over-aged." This condition is frequently used where maximum stability is desired. In cases where stability of the material is desired and where maximum physical properties are not necessary, a simple precipitation treatment is sometimes employed.

Heat Treatment of Magnesium

Most magnesium sand casting alloys now being used commercially are amenable to heat treatment with marked improvements in mechanical properties. Both the solution and the precipitation methods are used, depending on properties required. The solution heat treatment (H.T.) of the Mg-Al-Mn alloys consists of 16 hr. heating at 775° F. in an non-oxidizing atmosphere followed by a quench. It increases the tensile strength, toughness, and percentage elongation of the alloys. The smaller percentage aluminum in the Mg-Al-Mn-Zn alloys permits the use of a lower temperature and this is desirable to avoid the hot-short temperature zone occasioned by the presence of zinc.

Precipitation Treatment

The aging or precipitation heat treatment (H.T.A.) is given to the solution heat treated material and is substantially complete after 16 hr. at 350° F. This treatment raises the yield strength and hardness of the alloys with a lowering of ductibility and toughness. Properties intermediate between those of the solution heat treated and the fully aged conditions can be obtained by giving the alloy a partial or short-time precipitation heat treatment. This partial aging is not used extensively, however, because of the practical difficulties encountered with furnace charges consisting of various sized castings with both thick and thin sections.

Accurate temperature control is very important, particularly with the solution heat treatment which is carried out at a temperature close to the softening point of the alloy. Electrically heated ovens equipped with automatic temperature control and fans that give reversals in direction of the circulating atmosphere insure the proper temperature regulation.

Defects—Their Causes and Prevention. The principal defects encountered in a light metal permanent mold foundry are dross and inclusions, blow holes, shrinkage, porosity, cracks and misruns. Such defects are largely a function of melting, pouring, gating, venting and painting.

Dross and inclusions are usually due to dirty metal; improper or insufficient skimming of the metal in the melting furnace and/or in the ladle; to improper pouring and gating which permits too great agitation of the metal as it is poured and as it enters the mold cavity.

Blow Holes

Blow holes are usually due to dirty metal whose increased viscosity makes difficult the escape of entrapped air or gas; to improper pouring and gating where overly turbulent metal will carry with it air which cannot be given off before solidification; to improper venting which prevents the escape of air or gas in either the metal or the mold cavity.

Coarse grain size may not always be due to high metal temperature, but may be an inherent characteristic of the particular metal being used. Such coarse grains can cause shrinkage especially of the surface variety. In such a case, the addition of a grain refiner to the holding pot will eliminate or minimize the shrinkage defect.

Shrinks may appear either as surface depressions or as porous spots. They indicate either lack of feeding or lack of progressive solidification. Shrinkage may be eliminated by changing the trim or amount of paint in the shrink area; by the use of adequate riser and chills; or by changing the gating so that cooler metal is supplied to the areas in which shrinkage occurs. When shrinkage is accompanied by large grain size it is generally an indication of a high pouring temperature. By increasing the area of the inlet to the mold cavity, it may be possible to pour the casting at a lower temperature, thus decreasing the amount of hot metal at any one point and overcoming the tendency for shrinkage to occur. Lack of venting may also be the cause of shrinkage in permanent mold castings.

Porosity Causes

Pinhole porosity in aluminum is due to gas, generally hydrogen, absorbed by molten metal, usually from contact with moisture, and which has been released as the metal solidified. The principal factors which influence the occurrence of pinhole porosity are: High melting and pouring temperature—absorption of gas in melting; use of oil scrap which has not been properly cleaned and refined; excessive agitation of metal during melting, transferring and pouring; rate of solidification too slow; insufficient venting.

Pinhole porosity is not nearly as prevalent in magnesium castings as in aluminum. It generally occurs with other defects such as dross pockets, blows, etc., and is probably due to similar conditions. To aid in its elimination clean metal should be poured from a ladle held close to the mouth of the sprue. Proper gating for maximum skimming and minimum agitation of the metal is also a factor. Tilting and venting of molds, particularly for large flat areas of tin aids in the elimination of this particular defect.

Hot Shortness

Cracking in castings is the result of "hot shortness" in the alloy at temperatures just below the freezing point. To minimize cracking, sources of resistance to contraction as the casting cools through the hot short range should be eliminated. Weak parts may be strengthened by the use of ribs, fillets or fins, or with gating technique which has for its goal the promotion of progressive solidification.

Misruns may be due to metal poured at too low a temperature; the mold being too cold or its temperature improperly controlled; the mold paint being improperly applied and trimmed; insufficient venting may build up enough back pressure to hold back the air in the mold cavity, large areas of such air may then be surrounded by molten metal and

cause misruns to occur in casting.

Development and Field of Use. A.S.T.M. specifications are the bases for Table 1. The alloys listed by A.S.T.M. were in use before and during the war and are now somewhat dated. War experience necessitates the review of limits recommended for materials for use in permanent mold casting. Based on this experience, A.S.T.M.'s specifications are being drastically revised at present. Changes in alloys and composition limits may, therefore, be expected in the very near future. For this reason, detailed discussion of developments and uses will be held in abeyance until ideas on post-war specifications crystallize.

Committee Personnel

Personnel serving on the Permanent Mold Castings Committee of the Aluminum and Magnesium Division were as follows: Alfred Sugar, *Chairman,* The American Metal Co., Ltd., New York; L. G. Howlett, Bernett Castings Co., Hartford, Mich.; A. W. Stolzenburg, Aluminum Company of America, Detroit, Michigan; Major J. W. Wheeler, Consultant, Springfield, Mass.; W. E. Carver, Amory Permanent Mold Co., Jamaica Plains, Mass.; W. J. Klayer, Aluminum Industries, Inc., Cincinnati, Ohio; and R. E. Ward, Eclipse-Pioneer Division, Bendix Aviation Corp., Teterboro, N. J.

DISCUSSION

Chairman: R. E. WARD: Eclipse Pioneer Div., Bendix Aviation Corp., Teterboro, N. J.

Co-Chairman: A. T. RUPPE, Bendix Products Corp., South Bend, Ind.

HIRAM BROWN[1]: You spoke about running permanent mold castings of about ⅛-in. in thickness. Would it be practicable to have ⅛-in. thickness on the wall and bottom of a frying pan, for example.

MR. SUGAR: You can pour a casting such as a frying pan ⅛-in. thickness all over. It can be done. In fact, you can pour irregular surfaces such as a washing machine agitator or a waffle grid down to 0.090-in. all over thickness. However, in an article like a frying pan, you would prefer to have a bottom that is a little thicker than the sidewall to get better and more uniform heat dispersion, i.e., uniform heating all over the pan. You can run them as thin as 0.125-in. and do it successfully.

MR. BROWN: We were running that thickness before and we had trouble. I was wondering if there was some particular reason to expect trouble on a job of that type.

MR. SUGAR: The thinner the section is the more difficult it is to run. You might change your pouring temperature or your mold painting. Very often you will find by roughening your mold paint you can pour a much thinner casting than you can with a normal mold paint job.

J. G. MEZOFF[2]: I would like to ask Mr. Sugar about the use of his term "even solidification." Is that contrasted with progressive solidification?

MR. SUGAR: I meant progressive solidification, with the casting solidifying towards the gate at all times.

MR. MEZOFF: I understood you to read in your report that in a permanent mold casting, you wished to get more or less uniform sections throughout, in order to obtain even solidification.

MR. SUGAR: Progressive solidification is what I had in mind.

M. G. CORSON[3]: I would like to ask Mr. Sugar whether he recommends degasifying the metal before pouring, and secondly whether he has ever treated the castings for density. I believe that there are two causes for trouble. First, if your metal has gas, that gas will develop gas pockets, and if you have a bubble of gas which did not show up as cast, it may show up by heat treatment by forming a blister.

MR. SUGAR: You speak of gas as being detrimental. That will depend on the type of casting that you are about to make. If you are making an item for direct consumption, that is, certain items like agitators, you will find gas is very helpful in filling your casting. On the other hand, if you are looking for the maximum physical properties, then gas will have a very detrimental effect on it.

I think you will find that your melting practice has a very important bearing on the amount of gas that is put into the metal. If you get a good grade of ingot and your melting practice is also good, there will be little or no need for you to degasify your metal.

MR. CORSON: The ingot metal usually comes with quite an amount of gas in it and it has a very low density to start with. That is why I recommend, in the case of Alloy 220, making a small heat and checking the density. If the density checks 2.565, the metal is all right; otherwise, it must be cleaned. Secondly, if you have gas in your metal, blistering is practically unavoidable.

MR. SUGAR: You mean in heat treatment.

MR. CORSON: Yes, and the main trouble, particularly in the use of Alloy 220, when you try to die cast it, it is susceptible to intergranular shrinkage.

MR. SUGAR: Does that have anything to do with the amount of gas?

MR. CORSON: No, it may be absolutely clean; they are entirely different things. Intergranular shrinkage means that during solidification, the grains pull apart.

MR. SUGAR: I would not associate gas with intergranular shrinkage. For shrinkage, you will feed and for gas blistering, you can degasify with boron trichloride, nitrogen or chlorine gas.

MR. CORSON: The metal should be melted by electrical resistance at the outset, so that no gas will touch it at all.

MR. SUGAR: That will depend entirely on the quality of the ingot you get and the quality of your melting. If the ingot is good and your melting practice is good, then there should be no necessity for degasing. On the other hand, if either the ingot is poor or your melting practice is poor, then you should degasify.

G. G. GLIDER[4]: In respect to degasifying or cleaning the metal, have you found any correlation between cleaning of the metal and the grain size of the metal? Do you not find that the cleaner you get the metal, the coarser the grain?

MR. CORSON: Yes, and I like it, because I get higher tensile strength and higher ductility. Of course, if it is a question of surface finish, the large grains are not desirable, but otherwise, it does not make much difference whether the grain is 3 or 4 mm. or 0.1 mm. in size.

CHAIRMAN RUPPE: Do you recommend a binder for sand cores?

MR. SUGAR: No. Any good binder that is used normally in making sand cores is satisfactory.

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[1]Solar Precision Castings, Des Moines, Ia.
[2]Dow Chemical Co., Midland, Mich.
[3]Consulting Metallurgist, New York.
[4]Ebaloy Foundries, Inc., Rockford, Ill.

VETERAN TRAINING

This paper interprets provisions of the G.I. Bill in relation to educational benefits to qualified veterans of World War II. These provisions are interpreted in relation to the Four-Year Foundry Course and the Two-Year Machine Shop Training Course given by the author's firm under the G. I. Bill to returning veteran employees.

W. M. Owen
Assistant Director of Training
Caterpillar Tractor Co.
Peoria, Ill.

ONE OF THE PROVISIONS of Public Law No. 346, better known as the "G. I." Bill, is "On-the-Job Training," a program subsidized by the Veterans Administration to further the education and training for veterans in industrial trades, crafts, and similar constructive learning. The Bill provides federal government aid for readjustment to civilian life of the returning World War II veteran. Any veteran who has served 90 days or more, who has other than a dishonorable discharge, who served after September 16, 1940, and before the termination of the war is entitled to benefits.

Training Institution

A veteran who is eligible for educational benefits under the "G. I." Bill may elect to take his training or education at any approved educational or training institution (including training courses conducted by industrial establishments). It does not matter whether or not the institution which will accept him as a student or trainee is located in the state in which he resides. The author will confine his remarks to how the bill applies to industrial training and more particularly how this program has been operating at Caterpillar Tractor Co.

For many years prior to the war, Caterpillar Tractor Co. had conducted a number of full-time training courses. Prior to the war this company had eight full-time courses in its plant at Peoria, Ill. These courses included apprentice training in the machine shop, foundry, and pattern shop, and employee training courses in machine shop, sheet metal work and arc welding.

Training Courses Offered

In addition to these, a College Graduate Training Course, Business Training Course and a Cooperative Training Course were offered. In November, 1944, these courses were approved by the Veterans Administration for benefits under the "G. I." Bill. The author's company was among the first to have training courses approved by the Veterans Administration.

At the present time this company has approximately 300 veterans enrolled in their various training courses, all of whom were either receiving subsistence allowance or have their papers in process. The author will point out what the benefits are for a trainee or apprentice who is a veteran and who qualifies for educational benefits by citing some wage rates under the "G. I." Bill.

The veteran who enrolls full-time in a school, college or university has the cost of schooling (tuition, books, fees, etc.) up to $500 per year paid by the Veterans Administration. In addition to this, the veteran may receive $65 per month if single, or $90 per month if married, as a subsistence allowance.

The veteran who is taking "on-the-job" training and receiving compensation for his work, will be entitled to the difference between the beginning rate of pay for journeymen in the field for which he is being trained and his training rate of pay, so long as the difference does not exceed either $65 per month or $90 per month, depending upon whether he is single or married, and so long as this difference added to his regular training pay does not exceed $175 per month for single men and $200 per month for married men.

Foundrymen Course

Four-Year Foundrymen Course. At Caterpillar Tractor Co., Peoria, Ill., there is a gray iron foundry where for a number of years a Four-Year Founderymen Apprentice Course was conducted. Table 1 shows that the wage rates for this particular course commence at seventy-one cents for the first six months, seventy-six cents for the next six months, eighty-one cents for the next six months, etc. Below these figures are these same rates converted to equivalent monthly rates, thus seventy-one cents per hour for 40 hours per week is equal to $123.07 per month. The graduating monthly rate of $199.33 is also indicated. This is the rate of $1.15 per hour converted to a monthly figure.

Now if a qualified veteran enters this training course he will receive seventy-one cents per hour or $123.07 per month for the first six months. He is being trained for a job which will pay him $1.15 per hour or $199.33 per month. The difference between these two figures

Presented at an Apprentice Training Session during the Fiftieth Annual Meeting of the American Foundrymen's Association, May 9, 1946, in Cleveland.

Table 1

TRAINING PAY AND SUBSISTENCE ALLOWANCE FOR QUALIFIED VETERANS IN TRAINING AT CATERPILLAR TRACTOR CO.[1]

1040-hr. periods[2] (6 mon.)	*Four-Year Foundryman Course*								*Two-Year Machine Shop Training*			
	1	2	3	4	5	6	7	8[3]	1	2	3	4[4]
Hourly rates	$ 0.71	$ 0.76	$ 0.81	$ 0.86	$ 0.92	$ 0.97	$ 1.02	$ 1.07	$ 0.91	$ 0.95	$ 0.98	$ 1.01
Equivalent monthly rates	123.07	131.73	140.40	149.07	159.47	168.13	176.80	185.47	157.73	164.67	169.87	175.07
Graduate monthly rates	199.33	199.33	199.33	199.33	199.33	199.33	199.33	199.33	195.87	195.87	195.87	195.87
Difference	$76.26	$67.60	$58.93	$50.26	$39.86	$31.20	$22.53	$13.86	$38.14	$31.20	$26.00	$20.80
Subsistence allowance per mon. (single)	51.93	43.27	34.60	25.93	15.53	6.87	—	—	17.27	10.33	5.13	—
Subsistence allowance per mon. (married)	76.26	67.60	58.93	50.26	39.86	31.20	22.53	13.86	38.14	31.20	26.00	20.80

[1] Showing increasing training pay and decreasing subsistence allowance under the G. I. Bill. Figures are based on 40 hr. per week.
[2] Numerals indicate successive 1040-hr. periods. Wage increases every 1040-hr. period, credit given for overtime.
[3] Graduating rate, $1.15. [4] Graduating rate, $1.13.

is $76.26 per month. The Veterans Administration will pay this veteran $51.93 per month, if he is single, or $76.26 per month if he is married.

Table 1 indicates the subsistence allowance per month for a married veteran and indicates the decreasing amounts that will be received by the veteran from the Veterans Administration. The top line indicates the increasing compensation he receives from the company. Both of these figures added together will equal the graduating rate of $199.33 per month. The subsistence allowance per month for single men added to the training pay per month totals $175, which is the top limit established by the Veterans Administration for single men.

The intention of the "G. I." Bill is to compensate the veteran for the time spent in military service by making available to him the difference between the training rate of pay and the graduating rate of pay, if he is enrolled in a bona fide training course. Recent changes in the bill placed a top limit on total earnings.

Two-Year Training Course

Two-Year Machine Shop Training Course. Table 1 shows the training rates for a two-year machine shop training course conducted by the author's company. The training rates of pay are ninety-one cents, ninety-five cents, ninety-eight cents, and $1.01 per hour for each six months period. The graduating rate is $1.13 per hour. Indicated in Table 1 are the differences between the graduating rate and the training rate.

In this particular course a veteran will receive $38.14 per month from the Veterans Administration for the first period of training if he is married, which together with the pay he receives from the company for training will equal the graduating rate. If the veteran is single he will receive $17.27 per month for the first period of training. This same pattern is followed in all training courses at the author's plant. The training rates and course length varies in the different courses.

"On-the-Job" Training

"On-the-Job" Training. A training course that is set up to meet the requirements of the approving agency for the Veterans Administration is more attractive to the veteran than one that is not. A veteran returning from service qualified and interested in the foundrymen course would naturally rather go to work for $1.15 per hour than seventy-one cents per hour. And that is the difference of having the training course approved by the subsistence allowance under the "G. I." Bill and not having it approved.

In most states the approving agency for the Veterans Administration is the State Board for Vocational Education. These people are guided by certain requirements that must be met, and the State Board for Vocational Education will help set up training in your business establishment. The requirements are fairly simple and include,

First, a list of the job processes upon which you agree to give the veteran experience. These job processes or job types should be comprehensive enough to develop a well-rounded mechanic in the field in which he is being trained.

Second, classroom instruction must be given to enable the veteran to learn related information pertaining to his craft or trade. This related information is to be given at the rate of 144 hours per year and the specific subjects listed in the schedule. The department of Vocational Education will assist by offering the required courses in your local high school.

Third, training rates of pay on a gradually increasing scale averaging no less than 50 per cent of the beginning journeyman's rate for the period of training. These three requirements must be in writing and submitted to the approving agency.

Administration Requirements

It is not the intention of the Veterans Administration, contrary to what many people think, to pay the difference between what a veteran may be receiving on a job and the top rate for that job.

For example, a veteran may be working on, let us say, a sand slinger. Let us say he is making $1.00 per hour and that is the starting rate paid on this job. Let us assume further that the top rate for the job of a sand slinger operator is $1.40 per hour. Some people think that by merely asking the Veterans

Administration the veteran will receive the difference of forty cents per hour or $69.00 per month from the Veterans Administration.

This could only be true if you set up a bona fide training program wherein you agree to train the veteran in all phases of the operation, that you give related instruction at the rate of 144 hours per year and that you have a schedule of increasing rates leading to the beginning rate for journeyman, $1.00 per hour, and further that the trade was an apprenticeable trade or one that normally required training to adequately handle the work. Training for just a sand slinger operator could not meet the requirements.

The Veterans Administration and the State Board for Vocational Education in the State of Illinois are not only willing, but are anxious to help in setting up training courses so that veterans may participate in the benefits under the "G. I." Bill. The Veterans Administration does not impose any restrictions whatever as to whom you employ or in any way interfere with your normal way of doing business.

Their only concern is that through the approving agency (the State Board for Vocational Education) that they have assurance that the training course meets their (The State Board) standard. Should you not meet the standards, at some later date the Veterans Administration will merely cut off the subsistence allowance to the veteran and interrupt his training under the "G. I." Bill.

DISCUSSION

Chairman — F. C. CECH, Cleveland Trade School, Cleveland.

Co-Chairman—W. J. HEBARD, Continental Foundry & Machine Co., East Chicago, Ind.

R. W. GARDNER[1]: I was quite interested in your program of apprenticeship, particularly where you indicated that you had a two-year program for machinists who had at the end of this time received a rate of $1.13 an hour, and you had a four-year program for apprentices who at the end of that time received $1.15.

What influence does the time factor have on your ability to get men to follow the foundry apprenticeship? There are many comments on the difficulty of getting foundry apprentices. It seems the program should be reversed. It looks like an inverted process.

MR. OWEN: We do have a four-year machinists apprentice course for young men 16 to 25 years of age. Formerly the age limits were 16 through 18, but we broadened the age requirement after the war. The two-year course in the machine shop is for the older employee. An applicant has to be between 20 and 32.

MR. GARDNER: How do your apprentice to machinist and foundrymen totals compare as far as your wishes are concerned?

MR. OWEN: At the moment we have 176 four-year apprentices in our machinist program, and a full quota would be about 275. Presently, in the foundry we have 12 apprentices, and a full quota would be 35 or 40.

Most of these 176 men are returned veterans, that have been on leave from our courses. We have the problem of attracting the right type of young man into our foundry. I think that people have a preconceived idea of what a foundry is. When a young man comes to the plant and asks for training, we discuss our various courses with him. We point out to him some things about the foundrymen apprentice course. Usually there is no waiting list and a boy can start right in if he is interested and qualifies. He will go back and talk with his parents, who most always discourage the boy.

We will ask a boy who shows a spark of interest, "Will you take a little time and let us take you through our foundry so that you can see through your own eyes what foundry work is?" We get the boy there and he changes his mind considerably about the foundry, even if he does not decide to take apprenticeship training in the foundry.

We do a lot of missionary work in that respect and it helps. All of us are going to have to do a lot more to help break down this barrier in the prospect's mind about the foundry. He hears it from all sides, "For heaven's sake, do not go into a foundry!"

Our foundry apprenticeship course, which is four years in length, takes the apprentice through all phases of foundry work, including some time in the pattern shop. He is given machine shop instruction. He will get more variety of work in his four years than any other four-year apprenticeship we have. We have a difficult time getting the number of people and the types of people we like to have in our foundry course.

T. W. GALLAGHER[2]: I would like to comment on Mr. Owen's statement where he said $65 a month is allotted to unmarried and $90 to married veterans. The law stipulates $65 without dependents.

I understood him to say that 144 hours of classroom instruction would be compulsory. The 144 hours is interpreted by the Veterans Administration for a registered course, "on the job" training, as not compulsory. You can go to any night school under this training program. Under this training program in the foundry, we find that this particular course of apprenticeship, compared to "on the job" training, gives us a better working man, who makes more money and gets along a lot better on the "on the job" program than the apprenticeship program in the foundry.

Also, the man is allowed remuneration for tools or equipment he needs for training, such as molder's shoes or anything else, up to and including $100.

MR. OWEN: I am glad you brought that up. I did not mention about the tools and incidental expenses. The Veterans Administration will pay the veteran for tools if the company customarily requires the non-veteran apprentice or trainee to provide himself with tools.

In regard to 144 hours per year related instruction, that applies to both apprentice and "on the job" training. The Veterans Administration says that any training course that is set up in industry has to be set up with 144 hours per year of related classroom work. The required related instruction can be had through night school courses, company training, or a home study course. Related instruction has to be supplied through some means, even though the training is not apprentice training but is "on the job" training.

If the company provides related instruction and it meets the requirements of the approving agency, the requirements are fulfilled. The Veterans Administration states that if we fall short of 144 hours of related instruction they will take the veteran out of the training course.

[1] Ford Motor Co., Dearborn, Mich.
[2] Lake City Malleable Co., Cleveland.

FANS AND EXHAUSTERS

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B. F. Sturtevant Division
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The author defines fans and blowers. He discusses the centrifugal fan, squirrel cage type fan, radial blade or paddle wheel type fan and the non-overloading type fans. Design and selection of hoods and piping, dust collectors, and fans are interdependent.

RESEARCH and development which have brought fume and dust hoods, piping systems, dust collection and disposal, important component parts of an air handling system to their present high efficiencies involve much in the way of time and expense. Full benefit of these developments cannot be realized without the use along with them, of an equally efficient device which, figuratively speaking, gives them life—just as the heart does to the human machine with *its* veins or ducts and *its* dust and waste collectors. A well built human machine is helpless without a good heart or pump. Similarly, a well designed air handling system cannot produce its potential benefits without it too having a good "heart" which we have come to know in the foundry industry as a fan or blower.

The first use of a fan for collecting and conveying dust was in 1866. The writer quotes from an article by Mr. F. Burnham McLeary in the March, 1914, issue of the *World Work*, referring to this application as follows:

"Now this man Sturtevant was a Yankee and brimful of Yankee brains. One industry, it appears, did not satisfy him. And so when his workmen complained that a shoe buffer he had made, stirred up a fearful cloud of dust, Mr. Sturtevant made a study of the blower problem. The result was a little exhaust fan which seated itself close to the buffing wheel and sucked away every particle of leather. The air was left clean as a whistle. Thus came into being the first satisfactory exhaust fan, and the father of all modern fans and blowers in use today."

Fans Defined

The NAFM (National Association of Fan Manufacturers) has recently issued Bulletin No. 105 entitled "Standards, Definitions and Terms in use by the Fan and Blower Industry." The author quotes from it as follows:

"The term *fans and blowers* has come into general use by custom. Within the industry, the terms *fans, blowers* and *exhausters* convey the following ideas:

1. Fans refer to the complete assembly of a machine comprising the rotor and housing and includes both blowers and exhausters.

2. Blowers are fans used to *force* air under pressure.

3. Exhausters are fans used to *withdraw* air under suction. A centrifugal fan consists of a fan rotor or wheel within a scroll (or spiral) type of housing, including driving mechanism supports either for belt drive or direct connection."

A centrifugal fan is a centrifugal pump. It imparts energy to the air passing through it by the action of centrifugal force, drawing air in at the center and discharging it at the periphery of the wheel into the housing from which it passes out through the discharge opening or outlet.

Despite being invisible when clean, air has definite weight or mass; 13.33 cu. ft. of air measured at 70° F. and 29.92 in. Hg. (normal atmospheric pressure at sea level) weigh 1 lb. This means that a fan handling 13,333 cfm. is actually working with a mass of 1000 lb. per min. or one ton every 2 min. A mass equal to 1 ton of sand or 1 ton of iron passes through 30 times an hour in that particular size of fan. It is obvious, therefore, that a fan for use in a foundry must be a rugged piece of equipment.

The measure of the total energy imparted to a quantity of air flowing through a fan, is the total pressure which it develops. When that quantity of air, or volume in cfm., passes through a fixed fan outlet opening, it must do so at a definite velocity since Velocity is $\frac{Q}{A}$, Quantity divided by Area. This velocity represents kinetic energy, the energy of motion, and is measured by the velocity pressure which creates that motion.

The difference between this velocity pressure and the total pressure generated by the fan is known as the static pressure. Velocity pressure acts only in the direction of air flow. Static pressure acts in all directions: at right angles to, and in the direction of, the air flow. It is the pressure which tends to burst or collapse a pipe or chamber. It represents potential energy.

The writer likes to think of velocity and static pressures in a less technical sense. He likes to think of velocity pressure as the vehicle which speeds the air along. He likes to think of static pressure, because it also acts in the direction of air flow, as the advance guard, the motorcycle squad, which moves ahead of the speeding vehicle, brushing aside all obstructions to give clear passage to the vehicle.

Presented at a Session sponsored by the Safety and Hygiene and the Plant and Plant Equipment Committees at the Fiftieth Annual Meeting of the American Foundrymen's Association at Cleveland, May 6, 1946.

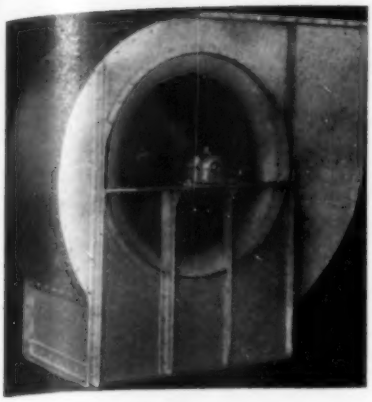


Fig. 1—Squirrel cage or multiblade type fan.

Imagine, if you will, that the inside surfaces of your air piping are not smooth but instead covered with bumps and projections, bristles and obstructions. Imagine further that our invisible motorcycle squad moves ahead smoothing out these obstructions. The measure of the energy with which this is done is the static pressure. *In short, it is the force which overcomes resistance to the flow of air.*

It takes horsepower to develop pressure. Obviously, the slower the air has to move, the less velocity pressure is required. The less resistance there is to overcome, the less static pressure is required. Keeping these component parts of the total pressure to a minimum, keeps the horsepower required to a minimum. Low air velocity also means less abrasion or erosion on elbows and ducts. Lower total pressure means lower fan speeds and longer life.

If the true resistance of a system is no more than the static pressure for which a fan connected to it is selected, the cfm. rating of that fan at that static pressure will actually flow through the system provided:

1. That the manufacturer's tables are rated on NAFM and ASHVE Standards.

2. That restrictions or bends at the inlet or outlet, especially the former, do not change the fan characteristic, which is determined by test with 10 diameters of straight pipe on the inlet or outlet or both.

3. That additional connections from additional grinding wheels, fume hoods and dust collectors over and above the original design are not added ad infinitum without increasing the original speed and possibly the size of motor.

The writer cannot emphasize too strongly that a definite relationship exists between the fan and the system to which it is connected. Many engineers have the impression that when a fan is selected from the fan tables for a certain cfm. and static pressure and the fan is operated at the speed indicated, the fan will automatically deliver that amount of air. That is not necessarily so.

A fan has an infinite number of capacities at a given speed. Imagine a fan in operation with no duct work on it and with merely a solid plate fastened to a short outlet extension. This is one extreme. We call this "Blocked Tight." The cfm. delivery is zero. The pressure developed is slightly less than the maximum pressure the fan can develop at that speed. It is called the Static No Delivery. It has no practical value in the field.

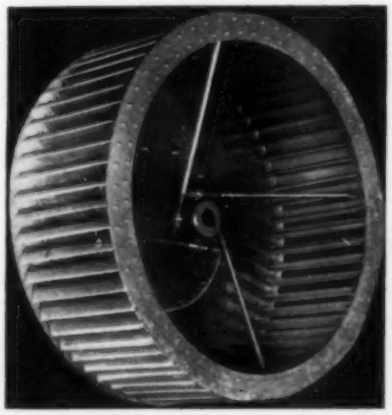


Fig. 2—Multiblade forward curved blade rotor for fan in Fig. 1.

The other extreme is to remove the solid plate and then we have the maximum capacity of the fan at that speed. It is called the Wide Open Volume. Between these two extremes there is an infinite number of capacities, and which one you obtain is wholly dependent upon the nature and magnitude of the restrictions placed on either the inlet side or discharge side of the fan, or both, in the form of hoods, piping, and collectors, rain caps, etc.

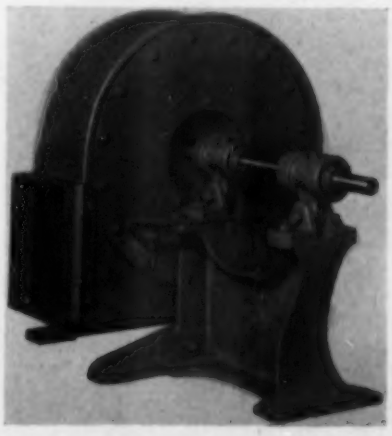
A fan will do as much as it is permitted to do. It will operate anywhere in its capacity range subject only and completely to the resistance of the system, in other words, the actual static pressure loss. A fan table merely says in effect—"*If* the static pressure loss is no more than *x* inches, then the fan will deliver the specified air quantity and only then."

If the static pressure loss is more, the capacity falls off. If it is less, the capacity increases. In the former case you find insufficient air movement at your hoods and a settling of dust in the ducts due to too low a velocity. In the second case, increased velocity is the result and usually the horsepower increases. With excessive velocity, erosion of elbows and piping is more rapid and collector maintenance may be increased.

Squirrel cage-type fan. There are on the market three distinct types of centrifugal fan. One of the oldest is the multiblade forward curved blade fan, sometimes known as the squirrel cage type (Fig. 1 and 2). It usually has 60-64 blades for single width single inlet, with or without the cup shaped depressions in the blades. It is primarily a ventilating fan for low static pressures, not a fan for dust collecting systems. It occupies a minimum of space for a given duty and runs at the slowest speed.

On the other hand, it has a rapidly rising horsepower characteristic and can easily overload a motor by a reduction in static pressure or by adding additional connections. The curved dished blades make good receptacles for dirt and muck. The

Fig. 3 — Radial blade or paddle wheel type of fan.



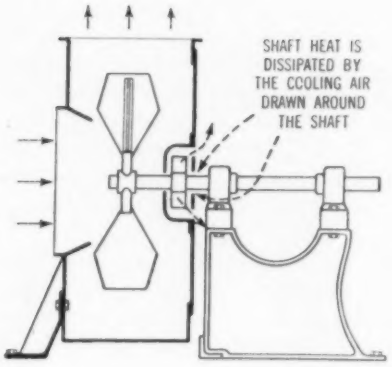
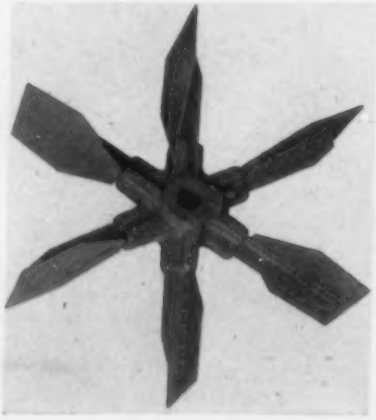


Fig. 5—Diagrammatic sketch of fan in Fig. 3, with provision for cooling shaft.

writer feels this fan has no place in a foundry with the possible exception of the ventilating or air conditioning system for the office section where the pressure requirements are low and the air relatively clean.

Radial blade or paddle wheel type fan. The second type is the radial blade or paddle wheel type (Fig. 3). This is the old reliable of the industry. It is truly an industrial fan. The wheel is made with 6 or 8 simple broad flat blades with or without side plates, depending on the nature of the material in the air stream (Fig. 4). It has been most commonly used for dust collecting work handling exhaust from grinding, buffing and polishing wheels, with or without dust collectors, and before and after dust collectors, usually in capacities up to 25,000 cfm. These fans are obtainable for larger capacities but they become bulky and run at low speeds, making for costly

Fig. 4—Radial blade rotor for fan in Fig. 3.



driving equipment.

These fans are also commonly used in core oven work and heat treating for recirculation and exhaust, handling high temperature air. They can be provided with heat radiating devices to protect the bearings without the use of water cooling, (Fig. 5).

The paddle wheel design lends itself very well to fabrication from heat resisting alloy sheet steel or to one piece castings of heat resisting alloy for that application.

The pressure characteristic is more stable than that of the squirrel cage type and the horsepower does not increase so rapidly with a decrease in system resistance. The paddle wheel fan can well be called the



Fig. 6—Non-overloading type fan.

old reliable of the industry.

Non-overloading type fan. The third type is the non-overloading type fan having a wheel with about 12-16 broad blades for single width, single inlet, backwardly inclined away from the direction of rotation (Fig. 6). While this fan is the newest of these three types, the first commercially successful design of backwardly inclined blades was actually made in 1913.

This fan has many desirable features. It has a steep stable pressure characteristic over a wide range of capacity. It is the closest approach to a constant volume centrifugal fan. An increase in system resistance causes less capacity reduction than in other types.

It has a self-limiting horsepower

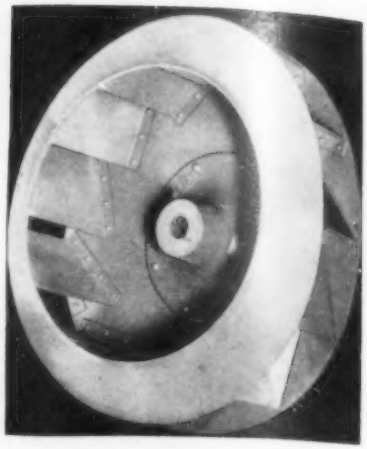


Fig. 7—Rotor with 12 broad backwardly inclined blades for fan in Fig. 6.

characteristic. A decrease in system resistance will not overload the motor since the fan has a predetermined maximum horsepower for every speed. Of the three types, this one has the highest efficiency.

It runs at high rotative speeds which permits the use of smaller diameter V-belt sheaves with a high horsepower rating per belt. It lends to direct connection to standard A. C. motors.

If the non-overloading fan has a wheel with flat, rather than curved, backwardly inclined blades (Fig. 7), it can be used successfully for handling dust laden air, particularly in capacities above 25,000 cfm. where it can show an appreciable savings in horsepower.

The flat blade is self-cleaning and the wheel will remain comparatively free from any deposits on the blades. There is no question about its use on the leaving side of an efficient dust collector. Its use without dust collectors for dry shakeout exhaust and fume exhaust cannot be given blanket approval as it is wholly dependent on the possibilities present for erosion and corrosion. However, its favorable characteristics definitely warrant first consideration.

All centrifugal fans can be furnished in NAFM Arrangement No. 1 and No. 2 where the wheel is overhung, with both bearings on the drive side and with a clear inlet. They can also be furnished in NAFM Arrangement No. 3 which has the wheel supported between

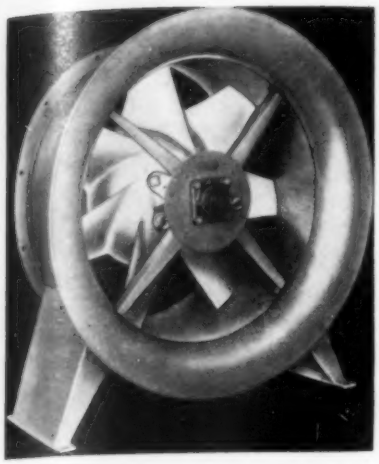


Fig. 8—Axial flow type fan.

bearings with one bearing in the fan inlet on a single inlet fan. Generally speaking, ball bearings with dust seals are best suitable for foundry installations.

More thought should be given to the use of Arrangement No. 3 except in the case of heavy dust loadings, dirty or corrosive fumes, and hot gases. In these cases the cantilever Arrangement No. 1 is unavoidable, but careful attention must be given to foundations and setting of the equipment. It is not reasonable to expect that overhung wheel fans and direct connected fans particularly, can be taken from a freight car and placed on a foundation without further examination. This apparatus must be checked for level and alignment.

Since the war, there has been a marked interest in the axial flow fan. During the war 45,000 of these fans were furnished for Navy combat vessels alone and used for hull ventilation, replacing earlier designs of noisier, heavier centrifugal fans. In addition, the Navy procured many more for forced draft boiler applications developing pressure up to 60 in. This fan will definitely find a place in foundry work. It can easily be built for the capacities and pressures you require in dust collecting work, although at present commercial offerings are limited largely to 1½ to 3 in. pressure. It is most practical when direct connected to the motor.

The problem of keeping bearings and V-belt drives out of the air stream in belt driven units is not serious and has been worked out. The axial flow fan should not be used indiscriminately, although it is suitable for certain fume exhaust work involving large volume of air. The author used the axial flow fan for that purpose for the past five years and believes it is the fan of the future.

Poor maintenance, addition of more connections without changing speeds and motors, are responsible only in a *small* measure for unsatisfactory performance of fume and dust collecting systems. By far and large, it is because of divided responsibility for the installation itself.

The writer tried to point out how interdependent the design and selection of hoods and piping, dust collectors, and fans are. The practice of purchasing these items individually at minimum cost, together with the unwillingness of the purchaser to pay for an extremely important intangible, namely, *Engineering Responsibility*, is the reason for unsatisfactory installations even when utilizing the best equipment available.

Do not think in terms of dust and fume collecting *equipment*. Think in terms of dust and fume collecting *systems*. Place the entire responsibility at one source. Contract for the entire system with one party on an installed and satisfactory operating basis. Whether this contractor be a fabricator, dust collector manufacturer or fan manufacturer, whether he manufactures 10 per cent or 90 per cent of the contract material in his own shops, is not of prime importance. The important thing is that this contractor be financially responsible and, above all, possess an engineering staff with the engineering "know how."

For every cu. ft. of air exhausted from a building another cu. ft. will leak in to replace it. No building can be made airtight. In the winter time this can create a hazardous, cold, drafty condition. Do not install a dust or fume exhaust system without giving equal consideration to the replacement of that air through a properly designed tempered air supply system.

Granted these investments in capital equipment are costly, but the day is gone when they can be avoided. State laws have seen to that. But, in addition, efficient fume and dust collecting systems contribute definitely to increasing morale and production. Obviously, therefore, they merit your most careful consideration.

Fig. 9—Three-blade light duty rotor for fan in Fig. 8.

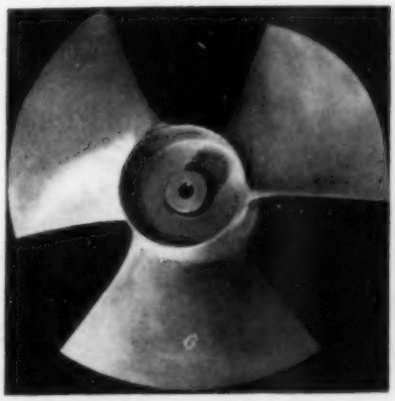
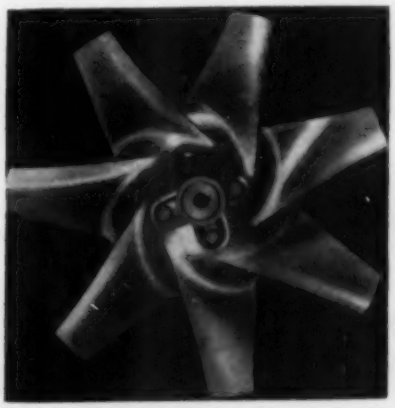


Fig. 10—Multiblade heavy duty rotor for fan in Fig. 8.



STEEL SUSCEPTIBILITY TO HOT-TEAR FORMATION IN CASTINGS

A contribution to the subject of hot-tear formation in low carbon steel castings. One possible cause may be the allotropic transformation of delta phase into gamma phase in steel.

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State Institute
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In any discussion of hot-tear formation in steel castings, the question of the susceptibility of various steels to hot tearing always arises, as pointed out by Briggs* (1943).

Original investigation and observation data indicate that:

(a) Various steels under the same conditions and depending upon these conditions are quite different in their susceptibility to hot-tear formation.

(b) The tendency to hot-tear formation is essentially influenced by the allotropic transformation of delta phase into gamma phase, that takes place in most kinds of steel during the cooling of castings.

The influence of this transformation will be clear if the following points are considered as the basis of the theory of hot-tear formation:

Hot tears are formed in most cases due to the hindered contraction of the metal during cooling.

Factors which are responsible for hindered contraction may be divided, in most cases, into two groups—internal and external.

(a) The hindrance caused by the mold, i.e., by the material of the etc. For identification, this type of mold, cores, cross-bars of the flasks, hindrances will be termed "external hindrances."

(b) The hindrances which appear due to the temperature gradient produced between the adjoining parts of a casting of various cross sections. In many cases, these temperature gradients reach a high degree in the cooling process. This type of hindrances will be termed "internal hindrances."

Both types of hindrances mentioned above can exist in a casting, either separately or together. To avoid confusion, other factors which may play a part in hot-tear formation have not been discussed in this paper.

Carbon Steel

Plain carbon steel immediately after complete solidification (just below the solidus) produces either:

(a) Solid solution of carbon in delta iron (in carbon steel, which contains carbon up to the limit of point *H* of the peritectic line *HIB* in the iron-carbon constitutional diagram) or:

(b) A mixture of solid solution of carbon in gamma iron and of solid solution of carbon in delta iron (in carbon steel, which contains carbon within the limits of points *H* and *I* of the peritectic line, called low carbon steel), or:

(c) Solid solution of carbon in gamma iron (in carbon steel with the carbon content above the limit of the peritectic point *I*).

Out of the latter kinds of steel must be distinguished the types in which paritectic reactions exist (in carbon steel with carbon content within the limits of points *I* and *B* of the peritectic line, further called by agreement medium carbon steel).

In this connection, the usual division of contraction into an "above pearlite" point and "below pearlite" point must be considered insufficient. The contraction must be divided into three stages:

(a) Contraction in the region of delta phase.

(b) Contraction in the region of gamma phase.

(c) Contraction in the region of alpha phase.

Hot-tear formation is greatly affected by a sharp change of the coefficient and the rate of contraction at the moment of the allotropic transformation of the first two phases (the transformation of delta phase into gamma phase) inasmuch as the mechanical properties of steel are known to be quite low at these temperatures.

It must be considered that this transformation accompanies a contraction in volume. Therefore, the temperature of this transformation can be considered "the main critical temperature" of hot-tear formation.

The transformation of gamma phase into alpha phase plays a secondary role (especially for carbon steel) because by that time the mechanical properties of steel significantly improve.

External Hindrances. If only external hindrances to normal contraction of the metal are present, hot tears appear as a result of failure of the metal at such high temperatures at which the cohesion between crystal grains became weaker than the grains themselves.

In such cases, the low carbon steel

*C. W. Briggs, "Hot-Tear Formation in Steel Castings," Transactions, American Foundrymen's Association, vol. 51, pp. 57-85 (1943).

This paper was presented and discussed at a Steel Castings Session of the Fiftieth Annual Meeting, American Foundrymen's Association, at Cleveland, May 10, 1946.

is less susceptible to hot-tear formation than the medium carbon steel. This is explained by the fact that the improvement of the mechanical properties of the crystals and of the cohesive force between them at the falling temperature near the solidification temperature is at a greater rate in the low carbon steel than in the medium carbon steel.

It is important to note that, in the low carbon steel, with the increase of carbon content up to the limit of the peritectic point better results are obtained. This can be explained by the fact that with the increase of carbon content in this kind of steel the temperature of the allotropic transformation of delta phase into gamma phase is comparatively higher.

The total contraction of metal in this case in the region of delta phase (when the mechanical properties of the metal are very low) is insignificant and, consequently, the danger of metal failure at the temperatures of this region is smaller.

Coefficient Change

Change of the coefficient and of the rate of contraction produced at the transformation of delta phase into the gamma phase is also not so harmful due to the fact that the improvement of the mechanical properties of low carbon steel after complete solidification begins sooner with the increase of carbon content. This refers especially to the toughness of metal due to the presence of gamma crystals.

Furthermore, by the time transformation occurs mold material has not yet been heated to such high temperatures and the hindrances to normal contraction reach dangerous magnitudes (in the cases when such types of hindrances exist).

Alloying elements eliminate the tendency of steel to hot-tear formation caused by external hindrances to normal contraction of metal if they effect the following:

(a) Increase at high temperatures the cohesive force between the crystal grains of the metal.

(b) Produce immediately after complete solidification solid solutions by means of which the strength and plasticity of the crystals improve.

(c) Do not produce new structural constituents, brittle at high temperatures, deposited at the grain boundaries.

More favorable alloying elements of those mentioned in the foregoing are those which widen the gamma region at the expense of narrowing the delta region, both in temperature and concentration directions.

Alloying elements narrowing the gamma region are less favorable than those widening the gamma region, even if they are added in such quantities that the gamma region entirely disappears and, in this case, the polymorphic change is absent.

Crystals Lower

This is explained by the fact that the toughness of crystals in alpha state is lower than that in the gamma state, and that here the cohesive force between the grains is realized more clearly.

Internal Hindrances. If only internal hindrances to normal contraction of metal are present, hot tears appear because thin parts of a casting tear off from the thick parts. In the case of plain low carbon steel, the temperature of the thin parts of the casting may drop during the cooling cycle to such values that the allotropic transformation of delta-gamma phases within them will end, while in the thick parts of the casting this transformation has not yet started. In consequence, the coefficient and the rate of contraction of the thin parts sharply increase, and the tendency to "tear off" from the thicker parts increases.

As at the same time the mechanical properties of metal of the thin parts of a casting which are affected by the allotropic transformation in the region of gamma phase are better than those of the thicker parts which are still in the region of delta phase, hot tears appear in the thick parts, i.e., in the weakest parts of a casting, as previously mentioned.

Gamma Phase

In the case of the medium carbon steel, all parts of a casting (thin as well as thick) just below the solidification point are in the region of gamma phase. In this case, the tendency of thin parts to tear off from the thicker parts is eliminated because both parts have the same structural constituent, and the sharp changes of physical properties which exist at the temperatures of the polymorphic transformation are absent.

Consequently, if only internal hindrances to the normal contraction of metal are present, the medium carbon steel is less susceptible to hot-tear formation than is the low carbon steel. The influence of carbon content in both kinds of steel upon hot-tear formation affected by internal hindrances to the normal contraction of metal has not been completely investigated as yet.

However, it may be supposed that the increase of carbon content in the low carbon steel will play a favorable part due to the increasing temperature of the allotropic transformation of delta-gamma phases.

Therefore, with the increase of carbon content in the low carbon steel the temperature difference between the thin and the thick parts of a casting at the beginning of the allotropic transformation of delta-gamma phases is quite small and, consequently, the tendency of thin parts to tear off from thicker parts is inconsiderable.

Alloying Elements

Proceeding from incomplete data on the influence of alloying elements upon hot-tear formation when only internal hindrances to normal contraction of metal are present, a conclusion can be reached that the alloying elements which widen the region of the gamma phase at the expense of the region of the delta phase, both in the temperature and in the concentration directions, are favorable, especially if they improve the mechanical properties of metal.

Alloying elements which narrow the region of the gamma phase have, in this case, a smaller but still favorable influence upon the elimination of the tendency of steel to hot-tear formation. But this can occur when these elements are added in such quantities that the gamma region vanishes completely, and thus the steel is transferred into a range where the allotropic transformation at high temperatures is absent.

It may be supposed that the elements which raise the temperature of the allotropic transformation of delta-gamma phase will also be favorable.

Internal and External Hindrances, and Other Factors. If both types of hindrances to normal contraction of the metal exist, the influence of carbon content and of alloying elements upon the susceptibility of steel to hot-tear formation depends upon the combination of these hindrances. However, under all of these circum-

stances the better steel is that which:

(a) Has no transformation of delta-gamma phases.

(b) Improves the mechanical properties of steel soon after complete solidification.

(c) Produces gamma solid solution immediately after complete solidification.

(d) Has no chemical compounds, deposited at the grain boundaries, which are brittle at high temperatures.

(e) Has no low fusible chemical compounds, at grain boundaries.

Little doubt exists that alloying elements, with regard to susceptibility of steel to hot-tear formations, in all cases have considerable influence upon heat conductivity and fluidity of the metal, diffusion processes existing during solidification, coefficient of contraction, and other properties. This influence is a subject for further research.

Discussion

C. H. LORIG[1] *(written discussion):* Mr. Gelperin, in suggesting as one possible cause for hot tears in low-carbon steel castings the allotropic transformation of delta phase into gamma phase, has made an interesting contribution to the subject of hot-tear formation. According to the iron-carbon constitution diagram, those steels with less than about 0.18 per cent of carbon will contain delta iron immediately after solidification and would, by the reasoning of Mr. Gelperin, be susceptible to hot tearing as a result of allotropic transformation. The steels of higher carbon content would contain no delta phase by the time they have completely solidified, and hence other causes for hot tearing would necessarily be operative on such steels.

Because the delta to gamma transformation results in a contraction in volume, it is not difficult to visualize its importance in promoting hot tears in the low-carbon steel castings, especially since such steels contain phosphorus and sulphur which may form constituents that exist in the vicinity of grain boundaries of lower melting temperature than the bulk of the steel.

On the basis of Mr. Gelperin's suggestion, it is reasonable to expect that the hot-tear propensity diminishes, perhaps sharply, in going from compositions in which the delta phase exists in the solid to those in which the phase is nonexistent. This is provided that the longer freezing range for the latter compositions, or some other factor, does not compensate for the disappearance of the transformation effect. Foundries may have noted such a change in hot-tear propensities with change in carbon content, but have not attempted an explanation. On the other hand, the matter may not have had much consideration, since the bulk of the steel cast is of a carbon content outside the range for the existence of the delta phase in the solid, and few foundries have had experience with low-carbon melts. In view of the absence of any supporting data in Mr. Gelperin's paper, in the interest of a fuller understanding of causes for hot tears, it would seem desirable that the importance of the delta-gamma allotropic transformation with respect to hot-tear formation be appraised by some one.

C. W. BRIGGS[2] *(written discussion):* Mr. Gelperin presents an interesting discussion on hot-tear formation, especially from the point of view that the allotropic transformation of delta phase to gamma phase is important in effecting sharp changes in the coefficient and rate of contraction of steel, and thereby influencing the hot-tear formation in steel castings. The author, however, does not justify his position with the presentation of data. It is understood by the writer that the collection of such data is decidedly not an easy matter. This is most certainly true of any study involving temperatures close to the solidification temperature of steel. However, such quantitative data are necessary to evaluate completely the suggestions which the author has put forth in his paper.

Studies by Driesen[1] for ingot iron indicate that there is a change in the coefficient and rate of contraction at a temperature of about 2555° F. This is because of the pronounced change from the delta phase to the gamma phase. For low carbon steels of 0.10 to 0.18 per cent carbon, the transformation that takes place is between a mixture of delta plus gamma phase and gamma phase. This transformation apparently does not produce a pronounced change in contraction; at least Briggs and Gezelius,[2] in studies of the free and hindered contraction of carbon steel, were unable to record any contraction differences between the two steels (0.08 and 0.15 per cent carbon), which solidified as a mixture of delta plus gamma, and the other four steels which solidified as the gamma phase.

In view of the fact that the author stated that the temperature of the transformation of delta plus gamma to gamma phase was the "main critical temperature" of hot-tear formation of low carbon steels, the writer wondered what the author would consider as the critical temperature of hot-tear formation of medium carbon steels? It is believed that the manner of selecting the critical hot-tear temperatures as set forth by Hall[3] and Briggs[4] is the more preferable.

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[2] Technical Research Director, Steel Founders' Society of America, Cleveland.

Briggs' References

1. J. Driesen, *Ferrum*, 1913-14, vol. 11, p. 129.

2. C. W. Briggs and R. Gezelius, TRANSACTIONS, American Foundrymen's Association, vol. 42, pp. 449-470 (1935).

3. H. Hall, Iron and Steel Institute Special Report 15, 1935, pp. 63-94.

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A.F.A. CLAY AND FINENESS DETERMINATIONS CAN BE QUANTITATIVE

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MOST RECENT published information on the reproducibility of test results of the fineness test for sands is that of Stanton Walker[1]. That report indicated that improvements in test performance could be made. The present study was made to determine whether reproducibility could be achieved for A.F.A. Clay and the Fineness Test.

The expenditure of time and money in the performance of these tests is substantial, hence they should be performed only after a *satisfactory gross sample* is obtained from the railroad car, heap, or system. Accuracy depends largely upon the care taken in sampling. It would be inadvisable to make these determinations where the practice is to take a small sample in a haphazard manner.

Determination of A.F.A. Clay Content of Foundry Sands. Before the fineness of the foundry sand can be determined any A.F.A. clay content must be separated from the sand grains. A.F.A. clay separation is a sedimentation procedure[2]. Essentially this procedure requires the dispersal of the A.F.A. clay in a dilute electrolyte solution and decantation or washing of the suspended clay particles from the sand grains.

Distilled water is not available in many foundries, hence tap water is used. The writer found that almost immediately after the sand sample is agitated with sodium hydroxide, as electrolyte, and tap water the flocculated clay settles upon the sand grains and is difficult to remove by decantation. A continuous washing apparatus was found to remove more flocculated clay than could be accomplished by the Standard A.F.A. procedure.

Hills[3] demonstrated that temperature of the water used was a factor determining the percentage of clay obtained; in the sands he used, the colder the water used the higher the percentage of clay obtained.

Water Temperature a Factor

Tap water "as received" and distilled water at room temperature were used by the writer in all work reported. With increasingly colder water, more water of hydration surrounds each suspended clay particle thereby preventing the particles from flocculating and settling out.

Sodium oxalate was tried as an electrolyte with tap water. The slight solubility of this salt required a large volume of water to dissolve sufficient material. Potassium oxalate has a greater solubility and can be made up in stock solutions of various strengths.

A series of pilot tests were conducted to determine the strength of potassium oxalate solution which would give the best results with the *particular tap water* used. The proper amount of any electrolyte may vary with tap water in various communities and sections of the country. It was found in this instance that 25 ml. of a one per cent solution of potassium oxalate gave the best results. Ten grams of this oxalate per liter of water was made up for the stock solution.

Pilot Tests for Clay Content

Pilot tests were made according to the following procedure. A 50 gram sample of foundry sand containing clay was oven dried at 105-110° C. and then agitated in a rapid stirrer for five min. with various amounts of the electrolyte and sufficient tap water to make a liquid volume of 500 ml. The liquid portion was then poured into a slender tall cylinder of 350 ml. capacity and set aside for at least six hrs. At the end of that time the concentration which produced the least settling-out of the clay (denoted by least depth of clear supernatant liquid) was chosen for use in the A.F.A. clay determination.

A procedure is presented for selecting the proper electrolyte to use with tap water for the A.F.A. clay determination. In the locality where the tests were made, 25 ml. of a one per cent potassium oxalate solution added to 475 ml. of water was found to be satisfactory. Wet mixing and riffle splitting of the sand sample will give quantitative duplicate determinations for the A.F.A. clay content and Fineness Number of foundry sands.

Presented at a Sand Research Session of the Fiftieth Annual Meeting of the American Foundrymen's Association at Cleveland, May 8, 1946.

Table 1

Duplicate Determinations for A.F.A. Clay Content from 18 Carloads of Sand Using 3 Types of Clay Dispersing Solutions

Car No.	Dispersing Solutions: Standard A.F.A. Procedure; Distilled Water			500 cc. of Saturated Sodium Oxalate Solution with Tap Water			25 cc. of 1% Potassium Oxalate Solution and 475 ml. Tap Water		
	A	B	Aver.	A	B	Aver.	A	B	Aver.
1	1.61	1.45	1.53	1.69	1.49	1.59	1.47	1.40	1.44
2	1.03	1.49	1.26	1.25	1.08	1.17	1.36	1.32	1.34
3	1.10	1.15	1.13	0 96	1.06	1.01	1.74	1.79	1.77
4	1.78	1.71	1.75	1.57	1.44	1.51	1.61	1.72	1.67
5	1.68	1.23	1.46	1.67	1.57	1.62	1.76	1.76	1.76
6	1.61	1.69	1.65	1.58	1.45	1.52	1.77	1.84	1.81
7	3.54	3.75	3.66	4.41	4.26	4.33	3.89	3.68	3.78
8	1.51	1.58	1.55	1.80	1.73	1.77	1.97	2.17	2.07
9	1.79	1.95	1.87	1.73	1.83	1.78	1.84	1.94	1.89
10	1.87	1.97	1.92	1.96	1.58	1.77	1.87	1.90	1.89
11	1.68	1.74	1.71	1.65	1.68	1.67	1.55	1.64	1.60
12	1.56	1.52	1.54	1.62	1.64	1.63	1.63	1.65	1.64
13	1.88	1.57	1.73	1.82	1.56	1.69	1.77	1.84	1.81
14	1.50	1.55	1.53	1.51	1.42	1.47	1.74	1.68	1.71
15	1.58	1.82	1.70	1.62	1.62	1.62	1.60	1.64	1.62
16	1.65	1.54	1.59	1.69	1.73	1.71	1.78	1.94	1.86
17	2.03	1.90	1.97	1.97	1.97	1.97	1.87	2.08	1.98
18	1.62	1.68	1.65	1.58	1.54	1.56	1.67	1.67	1.67
Aver.			1.75			1.74			1.85

Table 2

Difference Between Duplicate Determinations in Per Cent A.F.A. Clay for Each Electrolyte Used for the Eighteen Cars Reported in Table 1

Car No.	Electrolyte: Sodium Hydroxide and Distilled Water	Sodium Oxalate and Tap Water	Potassium Oxalate and Tap Water
1	.16	.20	.07
2	.46	.17	.04
3	.05	.10	.05
4	.07	.13	.11
5	.45	.10	.00
6	.08	.13	.07
7	.21	.15	.21
8	.07	.07	.10
9	.16	.10	.10
10	.06	.03	.09
11	.04	.02	.02
12	.31	.26	.07
13	.05	.09	.06
14	.24	.00	.04
15	.11	.04	.16
16	.13	.00	.21
17	.06	.04	.00
18	.10	.38	.03
Aver.	.165	.112	.08

Duplicate determinations were made for the A.F.A. clay content from eighteen car-loads of sand. Table 1 gives the results of these duplicate determinations using three electrolytes.

All results below the heavy line (Table 1) were obtained after the weighed oven-dried samples had stood in 50 ml. of water for 18 hr. before making the clay determination. In a few instances this soaking increased the A.F.A. clay percentage obtained. For control work this soaking procedure may not be desirable because of the length of time involved. Only when control tests are questioned would a soaking period be resorted to. Much time is required in any event to conduct accurately the clay and fineness test and short cuts will only lead to unreliability and uncertainty.

Table 2 shows the difference between the percentages of the duplicate A.F.A. clay determination for the 18 carloads of Table 1. From a quantitative and a statistical viewpoint, the potassium oxalate and tap water was superior to sodium hydroxide and distilled water as an electrolyte. For the 18 carloads of sand tested the difference between duplicate tests in per cent A.F.A. clay, using potassium oxalate and tap water was 0.08 per cent whereas for sodium hydroxide and distilled water it was 0.165 per cent.

Removal of the suspended clay using an automatic clay washer required from 40 min. to 2 hr. when sodium hydroxide and tap water was used. The potassium oxalate with tap water appeared to keep the clay in better suspension and comparative tests for washing times were made.

Following are results for two or more tests upon each sand washed using the continuous washer.

Comparative Sand Washing Test Results

A.F.A. Content, per cent	Potassium oxalate with tap water	Sodium hydroxide with tap water
4.0	25	over 40
1.5	15	over 40
2.4	20	over 40
5.0	25	over 120

In addition to saving time a more thorough sand washing job was obtained using potassium oxalate and the particular tap water used.

Reproducibility of fineness test results. The first investigation was made by placing 250 pounds of a washed and dried sand in a mixer and dry mixing for 20 min. Then 10-15 lb. increments were removed and stored in paper bags for future testing. When an increment was used it was placed in a laboratory mixer and mixed for two min. Then a 50 gram test sample was removed by spoon for the fineness determination. Table 3 gives the results for tests upon nine of the increments.

Visual observation of the dry mixing action of the grains indicated

Table 3

Sieve Analysis of Nine Increments from One 250-Lb. Sample of Washed and Dried Sand, Mixed Dry for 20 Minutes. Amounts Given Are the Percentages Retained on Various Screens

U.S. Sieve No.	20	30	40	50	70	100	140	200	270	Pan	A.F.A. Fineness No.
1	....	1.60	44.16	39.02	11.10	3.42	.56	.08	.04	.04	38.
2	....	1.62	44.54	39.22	10.84	3.24	.52	.06	.02	.02	37.8
3	....	1.52	43.10	38.90	11.32	3.92	.88	.14	.04	.06	38.6
4	....	1.54	44.32	38.62	10.88	3.58	.72	.14	.04	.04	38.2
5	....	1.70	46.82	38.96	9.56	2.42	.34	.06	.02	.02	36.7
6	....	1.54	43.62	38.40	11.22	4.12	1.00	.06	.08	.08	38.8
7	....	1.60	41.74	38.10	11.92	4.48	1.38	.38	.16	.28	40.4
8	....	1.56	41.64	38.66	12.00	4.28	1.10	.26	.12	.20	38.7
9	....	1.40	39.30	38.12	13.38	5.32	1.68	.44	.18	.26	41.30

Table 5

COMPARISON BETWEEN WET AND DRY MIXING METHODS LISTING RANGE BETWEEN HIGH AND LOW VALUES FOR EACH SCREEN FOR ALL INCREMENTS

U. S. Sieve No.	Dry Mixing	Wet Mixing
30	.30	0.34
40	7.52	2.15
50	1.12	0.75
70	3.82	1.58
100	2.90	0.71
140	1.34	0.14
200	0.38	.03
270	0.16	.01
Pan	0.26	.02
A.F.A. Fineness No.	4.6	0.5

Table 6

SIEVE ANALYSIS MADE IN DUPLICATE FOR EACH OF SIX CARS. FIGURES GIVEN REPRESENT PER CENT RETAINED ON VARIOUS SCREENS

U. S. Sieve No.	20	30	40	50	70	100	140	200	270	Pan	A.F.A. Fineness No.
Car 1	.08	2.66	30.08	33.56	13.76	8.32	5.96	1.94	.74	.86	49.5
	.06	2.50	28.92	32.72	14.32	9.36	6.76	2.36	.86	1.00	51.8
Car 2	.10	3.05	30.90	32.95	13.92	8.57	5.32	2.01	.80	.76	48.7
	.10	3.24	33.41	33.28	13.00	7.74	4.54	1.80	.69	.54	45.7
Car 3	.09	3.32	32.00	30.75	13.70	8.61	5.58	2.16	.70	.49	48.2
	.09	3.28	31.40	31.10	14.05	8.75	5.31	2.15	.70	.40	48.1
Car 4	.04	3.22	30.18	33.88	14.15	8.64	4.92	1.73	.64	.67	47.9
	.09	3.49	32.10	31.75	14.34	8.42	4.38	1.63	.56	.40	47.4
Car 5	.10	3.13	30.03	32.59	14.00	8.30	5.68	2.27	.75	.63	49.
	.14	3.62	33.50	30.45	13.47	8.09	5.02	2.24	.72	.57	47.7
Car 6	.06	3.04	29.54	32.82	14.42	9.32	5.76	1.94	.66	.58	49.6
	.06	2.76	27.04	32.00	14.94	10.04	6.74	2.42	.92	.66	52.2

the mixing action was not as thorough as desired and the differences between increments as shown in Table 3 were not surprising.

In an attempt to improve the procedure another 250 lb. of the same sand was placed in a large mixer along with sufficient water to give a moisture content of 0.75 per cent. Mixing time was ten min. Again ten increments were removed and each increment was quartered down by a riffle type splitter to determinate size after mixing. The determination sample size was whatever remained in the splitter box after the last quartering. The sample weights varied from 40 to 70 grams. Table 4 gives the results for this set of tests.

Wet Mixing Aids Reproducibility

Comparison of the results of Table 3 with Table 4 will show that the reproducibility was decidedly improved when the sands were wet mixed. Table 5 shows a comparison between the wet and dry mixing methods by listing the range between the high and low values for each sieve. For example, from Table 3 for the number 40 sieve the data shows for increment 5, that 46.82 per cent sand was retained, this being the high value, and for increment 9, 39.30 per cent sand was retained, this being the low value. The range is then 7.52 per cent. Table 5 is constructed to show the values of range are much more quantitative for the wet mixing than for dry mixing.

Six carloads of sand were sampled and each gross sample was mixed and quartered down to determination size in accordance with the Standard A.F.A. procedure. The results of duplicate determination are given in Table 6.

The data of Table 6 was not satisfactory from the quantitative view point when compared with previous data from the wet mixing of clay free sand.

Table 4

SIEVE ANALYSIS OF NINE INCREMENTS FROM ONE 250-LB. SAMPLE OF WASHED AND DRIED SAND MIXED WITH WATER FOR TEN MINUTES. AMOUNTS GIVEN ARE PERCENTAGES RETAINED ON THE VARIOUS SCREENS

U.S. Sieve No.	20	30	40	50	70	100	140	200	270	Pan	A.F.A. Fineness No.
1	....	2.32	41.10	36.40	12.80	5.24	1.40	.22	.05	.03	39.5
2	....	2.17	40.30	36.00	13.60	5.69	1.54	.23	.06	.03	39.9
3	....	2.19	40.80	36.40	13.05	5.61	1.54	.24	.06	.03	39.8
4	....	2.02	41.15	36.65	12.75	5.27	1.47	.22	.05	.03	39.7
5	....	2.08	41.95	36.40	12.25	5.03	1.49	.21	.05	.05	39.5
6	....	2.18	42.45	35.90	12.02	5.00	1.44	.23	.05	.03	39.4
7	....	2.13	41.60	36.40	12.25	5.37	1.44	.23	.05	.03	39.5
8	....	2.28	41.05	36.40	12.80	5.32	1.52	.23	.05	.03	39.4
9	....	2.36	41.98	36.61	12.30	4.98	1.48	.22	.05	.03	39.4

Better quantitative duplicate determinations were obtained from 18 carloads of sand when the following procedure was followed. The gross samples weighing 80 to 90 lb. were placed in a mixer of sufficient capacity after taking the sample. The sample was mixed for ten min. and then as the mixing continued small increments of sand were removed until 15 lb. were accumulated. The sand remaining in the mixer was discarded. The 15 lb. of sand was immediately quartered down to determination size by means of a riffle-type splitter. Sands containing up to four per cent moisture presented no problem in quartering. After the last splitting all sand remaining in the splitter box was used for the determination. Table 7 gives the results of duplicate fineness determinations for the 18 carloads of sand.

Conclusion

A satisfactory electrolyte for use in performing the A.F.A. Clay and Fineness Tests is one consisting of 25 ml. of one per cent potassium oxalate and 475 ml. of tap water. Pilot tests using various electrolytes and various concentrations were made before selecting the proper electrolyte. When using an automatic clay washing apparatus the chosen electrolyte produced more effective sand washing in shorter time with the materials used.

Quantitative duplicate determinations were achieved for the fineness test by a simple procedure of mixing the gross sample before drying, immediately quartering to determination size and using all the sand re-

Table 7

SIEVE ANALYSIS MADE IN DUPLICATE FOR EACH OF 18 CARS OF SAND. FIGURES GIVEN REPRESENT PER CENT OF SAND RETAINED ON VARIOUS SCREENS

U. S. Sieve No.	*20*	*30*	*40*	*50*	*70*	*100*	*140*	*200*	*270*	*Pan*	*A.F.A. Fineness No.*
Car 1	.14	6.57	38.78	26.40	11.97	6.96	4.41	1.80	.61	.59	45.1
	.20	6.67	38.72	26.40	11.92	6.98	4.45	1.82	.63	.61	45.2
Car 2	.23	7.09	37.58	27.58	13.50	7.43	3.17	1.17	.45	.52	43.5
	.17	6.76	37.85	27.95	13.10	7.22	3.16	1.19	.49	.53	43.6
Car 3	.12	5.44	33.70	28.92	15.40	9.26	3.86	1.30	.58	.58	45.9
	.13	5.86	34.15	28.95	14.91	8.73	3.62	1.37	.49	.61	45.5
Car 4	.17	5.25	34.78	27.90	14.21	8.57	4.12	1.61	.63	.80	48.0
	.17	5.56	34.00	28.35	14.00	8.47	4.05	1.66	.63	.76	47.9
Car 5	.12	6.09	35.49	28.25	13.95	8.71	3.40	1.18	.59	.53	44.8
	.10	5.80	35.21	28.45	14.33	8.77	3.33	1.18	.54	.53	44.9
Car 6	.12	5.37	35.62	30.05	14.20	7.96	3.07	1.02	.52	.49	44.2
	.17	5.58	36.05	29.80	13.72	8.09	3.00	1.02	.51	.50	43.8
Car 7	.08	2.84	24.39	30.55	19.09	12.00	4.17	1.09	.47	.58	48.7
	.09	2.84	24.60	31.20	18.88	11.68	4.16	1.08	.48	.60	48.5
Car 8	.12	4.56	30.30	29.75	16.40	10.25	4.03	1.23	.53	.56	46.8
	.17	4.60	30.30	29.98	16.35	9.85	3.94	1.20	.55	.53	46.5
Car 9	.10	4.36	30.90	32.40	16.55	8.82	2.92	.89	.50	.50	45.1
	.10	4.37	30.50	32.85	16.48	8.87	2.94	.87	.49	.51	35.1
Car 10	.18	5.18	34.05	30.15	14.55	8.46	3.62	1.33	.52	.56	45.4
	.20	5.62	33.90	29.60	14.20	8.22	3.60	1.32	.55	.58	45.3
Car 11	.09	5.33	34.95	28.45	14.40	8.40	3.61	1.51	.63	.66	45.6
	.14	5.33	34.70	28.60	14.32	8.44	3.58	1.21	.66	.63	45.5
Car 12	.11	4.97	32.90	30.48	15.45	8.60	3.16	1.06	.57	.54	45.3
	.08	5.12	33.35	30.50	15.13	8.38	3.12	1.06	.56	.51	45.0
Car 13	.14	6.18	33.50	27.75	15.24	9.18	3.72	1.21	.55	.54	45.4
	.12	6.14	33.35	27.90	15.10	9.38	3.74	1.19	.56	.53	45.3
Car 14	.16	5.68	33.20	29.00	15.70	8.83	3.26	1.01	.55	.51	45.1
	.17	5.55	33.25	29.00	15.58	8.80	3.24	1.04	.53	.53	44.9
Car 15	.13	5.33	34.35	28.80	14.41	8.61	3.61	1.28	.59	.60	45.6
	.16	5.34	34.35	28.80	14.50	8.59	3.54	1.24	.59	.62	45.6
Car 16	.15	4.97	33.45	29.40	14.90	8.57	3.72	1.33	.60	.60	45.9
	.10	5.26	33.40	29.80	14.40	8.45	3.70	1.34	.57	.57	45.9
Car 17	.13	5.95	37.35	28.60	12.95	7.14	3.15	1.26	.51	.45	43.9
	.12	6.04	37.50	28.90	12.95	7.09	3.18	1.26	.51	.45	43.8
Car 18	.11	5.01	33.40	27.75	14.48	8.77	4.72	1.94	.73	.78	48.0
	.13	5.13	33.35	28.20	13.95	8.65	4.75	1.81	.76	.80	48.0

maining in the splitter box after the last splitting.

Acknowledgment

The writer wishes to express his appreciation to the American Steel Foundries for permission to publish these data. This study was made while the writer was employed in their research laboratory. In particular, Mr. John A. Rassenfoss, Supervising Engineer, supervised and encouraged this study.

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DISCUSSION

Chairman: B. H. BOOTH, Carpenter Bros., Inc., Milwaukee, Wis.

Co-Chairman: R. E. MOREY, Naval Research Laboratory, Washington, D. C.

J. L. HALL[1]: Did Table 1 show that with sodium hydroxide and distilled water you got approximately twice the amount of clay?

MR. WILLIAMS: No, it showed the difference between duplicate determinations, not the total percentage of clay.

DR. HALL: What was the percentage of clay in the samples?

MR. WILLIAMS: It was between 1 and 4 per cent. The paper itself gives the actual percentages of clay. I just picked out the difference between the duplicate determinations.

DR. HALL: Then, in your case, in your locality, you feel that the test using potassium oxalate and tap water was better than the standard A.F.A. test with sodium hydroxide?

MR. WILLIAMS: Yes.

O. J. MYERS[2]: How much sodium oxalate did you use?

MR. WILLIAMS: I used 25 cc. of one per cent solution. I tried several other concentrations before I arrived at that one. If you use too much or too little, it does not work as well. There seems to be a point at which you get best results.

MR. MYERS: Would not the variation in Table 7 be a function of the amount of sand obtained in the screen as well as the method by which you arrive at the results?

MR. WILLIAMS: The amount retained on the screen in this case varied quite a bit over the 18 carloads. The actual fineness number ran between 52 to 46. Duplicate checks obtained after removing the clay by the wet mixing method are close.

MR. MYERS: The 70 mesh screen in one case showed over 70 per cent variation and in another case 8 per cent. Did you have more sand on the 70 mesh screen in all cases and does that account for the larger variation?

MR. WILLIAMS: No. In many cases the 40 and 50 mesh screens had more sand on them than the 70 mesh screen.

G. F. WATSON[3]: Why did you use the potassium oxalate?

MR. WILLIAMS: It was necessary to find something that would work with the tap water in the particular locality where I was making these tests.

MR. WATSON: We used sodium pyrophosphate, which we found very effective, with some of the clays we tested. Sodium hydroxide was ineffective, but sodium pyrophosphate worked with all clays. Did you try that?

MR. WILLIAMS: No. There may have been something else that worked as well, but sodium hydroxide would not work with tap water. There is a host of electrolytes you could try. In addition to the oxalates, you could try the silicates or the phosphates. Even though all of these electrolytes have the sodium ion, you might find that potassium or magnesium salts would give much better results, depending entirely upon the tap water in your locality.

DR. H. RIES[4]: I would like to ask Mr. Watson if he is using tap water?

MR. WATSON: We are using tap water for part of our work.

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[1]American Manganese Steel Div., Chicago Heights, Ill.
[2]Werner G. Smith Co., Minneapolis.
[3]American Brake Shoe Co., Mahwah, N. J.
[4]Cornell University, Ithaca, N. Y.
[5]Harry W. Dietert Co., Detroit.
[6]American Steel Foundries, Indiana Harbor, Ind.

DR. [illegible]: Are you using distilled water too?

MR. WATSON: That is right.

H. W. DIETERT[5]: Do I understand Mr. Williams correctly that an automatic sand washer was used in this test?

MR. WILLIAMS: Yes.

MR. DIETERT: Was it satisfactory?

MR. WILLIAMS: It was in this particular case. There are three prongs that go down into the jar of the rapid stirrer. Solder on three brass strips, about 3½ to 4 in. in length, down to the bottom of those three rods. You will find ability to remove clay from sand grains is markedly improved. I had that done in two or three cases, and it works satisfactorily. I know one case where the clay content was doubled by just putting the baffles on.

MR. DIETERT: How wide and how long were these strips?

MR. WILLIAMS: Wide enough so you can still get the stirrer down there and leave a little room for the water going around. The length depends on how much work you want to put into it. I would say around 3½ to 4 in.

MEMBER: You mean the bottom of each leg was attached.

MR. WILLIAMS: It is a long strip. You just soldier or braze onto the vertical rods. This rod bends at the bottom slightly. You can follow the bend.

MR. WATSON: We have had those strips on for about six years now. When we first started out, we could not get along without them at all. We used a strip about ½ in. wide. I imagine it is about 4½ to 5 in. long. It prevents that funnel-like effect. The strips give better washing and reduce the time of washing immeasurably.

J. A. RASSENFOSS[6]: Specifications for baffles are described completely in a report by the Bureau of Standards in their Journal of Research. The procedure for the use of rapid washer of the type similar to a malted milk mixer, as first developed by the Bureau of Standards, is completely reported in this literature. You can purchase on the market cups for a malted milk stirrer, equipped with soldered-in baffles that meet the specifications given in this paper. Of course, you can use the same idea for dimensions and materials for other types of washers in which the stirring unit is used.

MR. WILLIAMS: Mr. Rassenfoss is familiar with the work that I have just described. We used the type of stir and baffles that he has just described to you.

CONDITIONS OF FLOW IN BRONZE CASTINGS

J. T. Robertson
and
R. G. Hardy
Naval Research Laboratory
Washington, D. C.

IT IS GENERALLY UNDERSTOOD that the design of the gate exerts a fundamental influence on the quality of the cast metal. The cleanness of the metal, proper directional solidification in the casting, and economical removal of the gate necessitate careful design of the gate and the location of the runners leading to the casting.

Much has been published in an effort to establish the general principles of good gating practice for various metals. Certain techniques have come to be accepted as being generally applicable to a particular type of casting or to a specific alloy but, in general, each casting is an individual problem.

This investigation was a study of the physical condition of some bronze alloys flowing in commonly used gating systems. The purpose of the work was to find the effect of various foundry conditions on the behavior of flowing bronze, and to establish some fundamental principles which could be applied to copper-base alloys in any type of gating system.

Review of Previous Work. A properly designed gate should introduce metal into the mold in such a way that: (a) proper directional solidification may be obtained, (b) the gate may be easily removed, and (c) the casting will be free from dross. This investigation is mainly concerned with the third requirement, that is, gating to prevent dross in castings. Dross arises from two sources; dross on surface of the metal in the crucible carried through to the casting, and dross formed by turbulent flow of the metal within the mold.

Several methods have been recommended for removal of dross before the metal reaches the mold cavity. Higgins[1], Lehman[2], and Crown[3], have discussed the importance of pouring basins. Dross from the crucible should float on the surface of a reservoir of the metal in a properly designed pouring basin, and only clean metal should enter the gate below. The pouring basin must be kept full, and the pouring stream directed so that metal does not flow directly down the sprue, but remains in the pouring basin long enough for the dross to rise to the surface.

Various types of gates have been designed to trap dross in the gating system and thereby prevent it from entering the casting. Higgins[1], Crown[4], Hensel[5], and Roberts[6] recommended that a strainer gate consisting of a sand wafer core containing numerous small holes be placed across the ingate or downgate to hold back the dross.

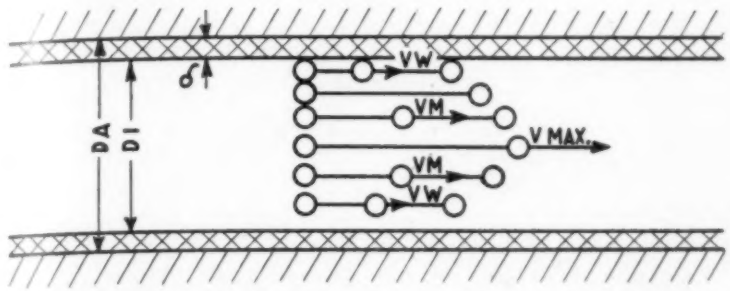
Other types of gates for retaining dross include one with a dead end to receive the first dirty metal[6], a small dummy skim riser before a constriction so designed that dross will float up into it[6], and a "swirl" gate which utilizes the centrifugal force of swirling metal to permit only clean metal to continue to the casting.

Even though the metal flowing through a gate is clean, dross may be produced when the metal leaves the gate and enters the mold cavity. This is especially true of aluminum bronze.

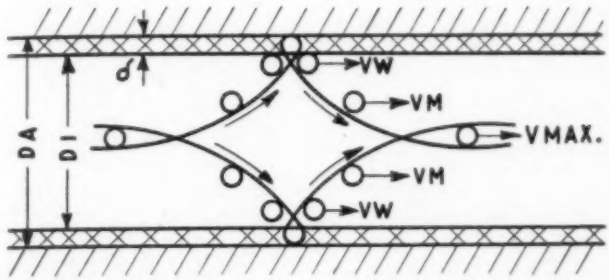
In the casting of some alloys the metal is poured directly into the riser in order to help promote directional solidification [1, 8, 9, 10, 11]. However, most proponents of direct riser pouring agree that this cannot be done with a drossy metal such as aluminum bronze because of the creation of dross by the splashing of the metal as it reaches the bottom of the mold. For this reason, most foundrymen recom-

Some factors influencing laminar and turbulent flow in bronze castings have been investigated with gun metal, manganese bronze and aluminum bronze. The diameter of the gate was one of the most important factors governing the linear velocity of the molten metal. Laminar flow produced the best mechanical properties, surface conditions and fracture of aluminum bronze. Laminar flow also was beneficial to manganese bronze, but had little influence on mechanical properties, surface or fracture of gun metal.

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STREAMLINE MOTION OF LIQUID PARTICLES



TURBULENT MOTION OF LIQUID PARTICLES

VMAX. = VELOCITY OF FLOW AT CENTER.
VM = AVERAGE VELOCITY.
VW = VELOCITY OF FLOW NEAR MOLD WALL.
DA = DIAMETER OF FLOW CHANNEL.
DI = INNER FLOW DIAMETER.
δ = THICKNESS OF STATIONARY ZONE NEAR MOLD WALL.

Fig. 1—Diagram showing streamline and turbulent flow conditions.

mend bottom gating for alloys which form dross [3, 4, 8, 9, 11, 12].

Even when a casting is bottom gated and the metal does not splash upon cores or projections of the mold, dross still may be produced if the metal is in a turbulent condition. Reversed horn gates are useful for preventing such turbulence [4, 12]. The horn gates should be used in the "reversed" position, that is, the large end of the horn gates should be adjacent to the casting. If the gate is used with the small end next to the casting, the jet caused by this constriction produces severe turbulence.

Lipps[13] and Ruff[14] explain how turbulence produces dross (Fig. 1). Streamline flow is characterized by all of the particles of the fluid moving in a straight line in the direction of flow. Turbulent flow is characterized by motion back and forth transverse to the axis of the stream. This turbulent flow results in new metal always coming to the surface of the stream, oxidizing and forming dross. In streamline flow, the same metal is always at the surface so that only a thin oxide layer is formed and relatively little dross is produced.

It is sometimes possible to calculate whether turbulent flow or streamline flow is taking place as a fluid flows through a circular channel when the velocity of the fluid, its viscosity and the diameter of the channel are known. According to Reynolds' formula:

$$Re = \frac{DVP}{u}$$

where

D = diameter of channel in ft.
V = velocity of fluid in ft./sec.
P = density of fluid, lb./ft.^{3}
u = viscosity of fluid in lb./sec. ft.
Re = Reynolds' Number.

A fundamental relationship has been established between the Reynolds' number and the friction factor (calculated from pressure drop due to friction) for fluids flowing through pipes[17]. The abrupt change in the slope of the curve at Reynolds' numbers of approximately 2,000 to 4,000 is the result of fundamental differences in the nature of flow above and below this region. Applying this relationship to the present problem, it may be surmised that if Re is less than 2,000 the flow should be streamlined, assuming that molten metals behave as do other fluids[17].

If conditions are such that Re is greater than 2,000, then flow may be turbulent. The viscosities of some copper-tin alloys have been measured[15] and found to be within the range for which the foregoing data were obtained. If the velocity of the copper-tin alloy in the gate is known, then the diameter of gate which will produce streamline flow can be calculated.

Velocity Measurement

Although the importance of the rate of pouring has been recognized, few investigators actually have made measurements of the velocity of molten metal in sand molds. It usually has been assumed that formulas for flow of water in pipes may be applied to the flow of metals in molds. The following formula is often quoted:

$$V = K\sqrt{2gh}$$

in which

V = velocity of efflux in ft./sec.
g = acceleration of gravity 32 ft./sec./sec.
h = height of reservoir in ft.
K = a constant.

Dwyer[10] offers a formula for calculating the length of time necessary to fill a mold. It is not evident whether this formula was derived from theoretical considerations or empirical observations. Benkoe[16] presents a theoretical formula for the velocity of aluminum in gates.

Ruff[14] actually measured the velocity of cast iron under foundry conditions. However, he used a downgate of large cross section with an ingate of small cross section. This probably resulted in higher velocities than would have been found if the ingate and downgate had been of the same cross section. He found velocities of about 7 ft./sec.

About 50 measurements were

made of the velocity of molten bronze flowing through gates of varying diameter, height and length. Other variables studied were the moisture content of the sand, pouring temperature and pouring time. The metals used for these experiments were gun metal, manganese bronze, and aluminum bronze. Several flat plates cast of these three metals under varying conditions of flow were examined for surface condition, and the mechanical properties were determined.

Melting Practice. High quality virgin metals and ingots of known composition were melted in clay-graphite crucibles in a 200-lb. capacity lift-coil induction furnace. Temperature was measured by means of bare wire chromel-alumel thermocouples with a standardized Leeds and Northrup millivoltmeter. This pyrometer had a time lag of about 5 sec. and was accurate to ±11° C. (±20° F.) at the pouring temperature of bronze. The melting procedures for the gun metal, manganese bronze and aluminum bronze are summarized in Table 1, and typical chemical compositions are given in Table 2.

Table 1

MELTING PRACTICE

	Gun Metal	*Manganese Bronze*	*Aluminum Bronze*
Crucible	Clay-Graphite	Clay-Graphite	Clay-Graphite
Furnace	High Frequency Induction	High Frequency Induction	High Frequency Induction
Melting Log:			
Copper	Melted under charcoal, deoxidized with 1 oz. 15% P-Cu/100 lb.	Melted under charcoal, deoxidized with 1 oz. of Al/100 lb.	Melted under charcoal, deoxidized with 1 oz. of Al/100 lb.
50-50 Fe-Al	—	2300° F (1260° C)	2300° F (1260° C)
50-50 Cu-Mn	—	2200° F (1200° C)	—
Sn	2060° F (1125° C)	2010° F (1100° C)	—
Al	—	—	2100° F (1150° C)
Zn	2010° F (1100° C)	2010° F (1100° C)	—
Pouring Temp.	2060° F (1125° C)	1800° F (980° C)	2010° F (1100° C)

Table 2

CHEMICAL COMPOSITIONS

	Composition; per cent					
Heat No.	*Cu*	*Sn*	*Zn*	*Al*	*Fe*	*Mn*
G25	87.53			9.24	3.01	
G23	87.84	8.14	3.98			
G24	87.26			9.71	3.03	
G32	57.92	0.60	39.06	0.86	1.00	0.56

Velocity Measurements. The data necessary for calculating the velocity were obtained by pouring a definite weight of metal through a specially designed mold in a known time. This mold was fitted with a pouring basin designed to maintain a constant liquid metal level, and was arranged so that the metal issuing from the ingate always fell below the lower level of the gating system in order to prevent back pressure (Fig. 2).

Molds were made of No. 2 Albany sand, and the pouring basins of dry core sand. The outlet from the pouring basin was of the same diameter as the gating system with which it was used.

Metal was poured into the side of the pouring basin until the metal level was ½-in. from the top, and was maintained at this level by constant pouring from the crucible for the duration of the test. The plug was removed from the opening in the basin when the desired metal level was reached, and as soon as metal began to issue from the gate a stop watch was started.

The metal was allowed to flow for about 10 sec. The plug was then replaced in the basin to prevent further flow; simultaneously, the watch was stopped. After the block of metal thus obtained had solidified, it was weighed and the volume velocity of the pouring calculated from the measured data; the volume velocity was then converted to linear velocity.

With the foregoing standardized procedures of melting and casting, the variables influencing the rate of flow were investigated, principally with gun metal. Emphasis was placed on geometrical factors such as the diameter (D), height (H) and length (L) of the gate (Fig. 2), because it was believed that these factors would have the most influence on velocity.

Other factors such as the moisture content of sand, pouring temperature, and pouring time were investigated primarily to determine their effect on the errors of measurement of the velocity. The velocity measurements with the various factors under investigation are listed in Table 3. Heats of manganese bronze and aluminum bronze were included

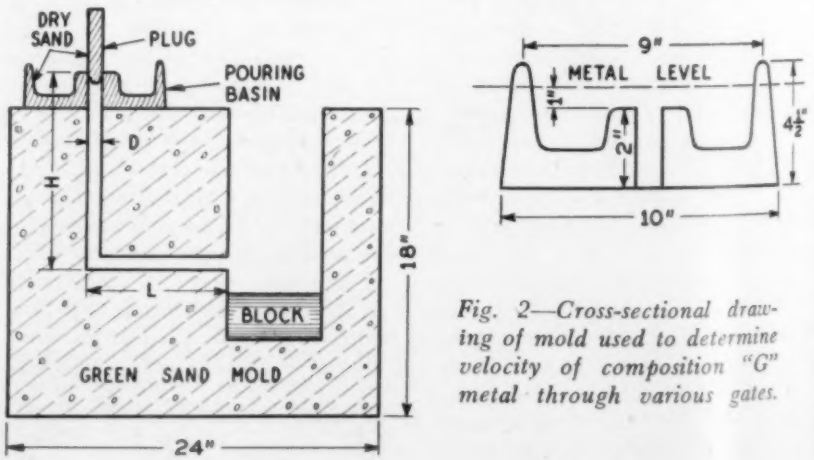


Fig. 2—Cross-sectional drawing of mold used to determine velocity of composition "G" metal through various gates.

Table 3

Velocity of Composition "G" Metal in Gates
(88% Cu, 9% Zn, 3% Sn)

Heat No.	Gate Diam., in.	Downgate Height, in.	Ingate Length, in.	Sand (% H_2O)	Pouring Temp., °C	Pouring Time, sec.	Velocity, ft./sec.
G-5	½	11	9	—	1125		2.84
G-5	½	11	6	—	1125		2.66
*	½	11	4	—	1125		2.90
G-14	½	8	9	9.0	1125		2.62
G-14	½	8	6	9.0	1125		2.66
G-6	½	8	4	—	1125		2.54
G-14	½	6	9	9.0	1125		2.58
G-14	½	6	6	9.0	1125		2.55
*	½	6	4	—	1125		2.42
*	¾	11	9	—	1125		2.33
*	¾	11	6	—	1125		2.05
*	¾	11	4	—	1125		1.58
G-15	¾	8	9	—	1125		2.25
G-15	¾	8	6	—	1125		1.58
*	¾	8	4	—	1125		1.45
G-10	¾	6	9	8.5	1125		2.32
G-10	¾	6	6	8.5	1125		2.14
*	¾	6	4	—	1125		2.04
G-16	1	11	9	—	1125		1.05
G-16	1	11	6	—	1125		1.32
*	1	11	4	—	1125		1.25
G-18	1	8	9	—	1125		1.38
G-18	1	8	6	—	1125		1.27
G-18	1	8	4	—	1125		1.37
G-19	1	6	9	—	1125		1.47
G-19	1	6	6	—	1125		1.47
*	1	6	4	—	1125		1.35
G-13	¾	8	4	6.3	1125		1.25
G-13	¾	8	4	8.5	1125		1.19
G-13	¾	8	4	10.0	1125		1.39
G-13	¾	8	4	12.0	1125		1.27
G-7	½	11	4	—	1135		3.10
G-7	½	11	4	—	1120		2.86
G-7	½	11	4	—	1105		3.10
G-7	½	11	4	—	1075		2.87
G-31**	½	8	4	—	1125		2.70
G-32†	½	8	4	—	980		2.62
G-40	½	8	4	—	1125	5.5	2.77
G-40	½	8	4	—	1125	8.9	2.88
G-40	½	8	4	—	1125	18.2	2.84
G-36	½	11	4	—	1315		2.27
G-36	½	11	4	—	1215		2.11
G-36	½	11	4	—	1150		2.16
G-36	½	11	4	—	1035		2.28

*Average of several heats
**Aluminum Bronze
†Manganese Bronze

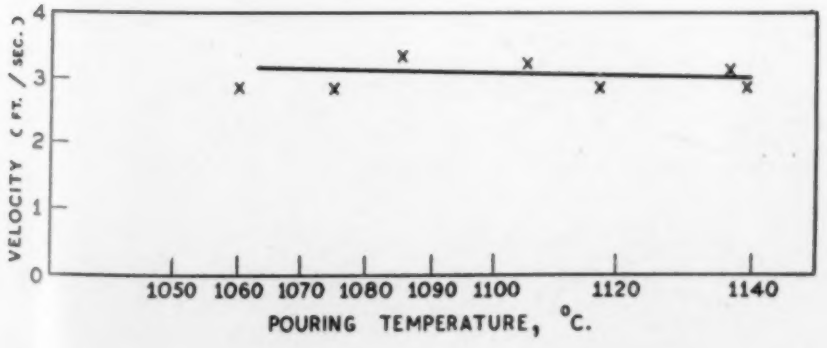


Fig. 3—Chart showing velocity of composition "G" metal vs. pouring temperature (11x½-in. downgate, 4x½-in. ingate).

for comparison with the gun metal.

Mold Design. The design of the mold was given careful consideration and that shown in Fig. 2 was chosen because it fulfilled the following conditions:

a. Permitted variations in the height, length, and diameter of gates.

b. Prevented back pressure from metal in the mold from acting on the metal in the gate.

c. Utilized a pouring basin that provided a constant liquid level, and a means for starting and stopping flow of metal into the gating system.

d. Provided a means of measuring the volume velocity (cu. in. per sec.) from which the linear velocity (ft. per sec.) could be calculated.

e. Allowed comparisons to be made between measured velocities and those calculated by means of the formula $V = K\sqrt{2gh}$.

Velocity Measurements. The velocity of composition "G" flowing through the gating system was but little affected by changes in pouring temperature, pouring time, or the water content of the sand (Table 3 and Fig. 3). Varying the downgate and ingate lengths from 4 to 11 in. had a negligible effect on the velocity of composition "G" (Figs. 4 and 5). Increasing the gate diameter from ½-in. to one in. decreased the linear velocity from 2.5 ft. per sec. to 1.3 ft. per sec. (Fig. 6).

Downgate Affects Velocity

The fact that the height of downgate had a negligible effect on the velocity was unexpected, and contrary to common theories and experiences with respect to fluids flowing through pipes under the influence of gravity. Equations for the flow of liquid through an orifice in the bottom of a large reservoir show that:

$$\text{linear velocity} = K\sqrt{2gh}$$

g = 32 ft. sec.^{2} (acceleration due to gravity)
h = height of the pipe in ft.
k = a constant.

Many formulas for predicting the rate of metal flow through gates are based on this equation [10, 13, 16]. It was believed that one reason for the negligible effect of height of downgate on velocity was that the liquid metal was running through a channel having permeable walls, i.e., a sand mold, whereas the conditions

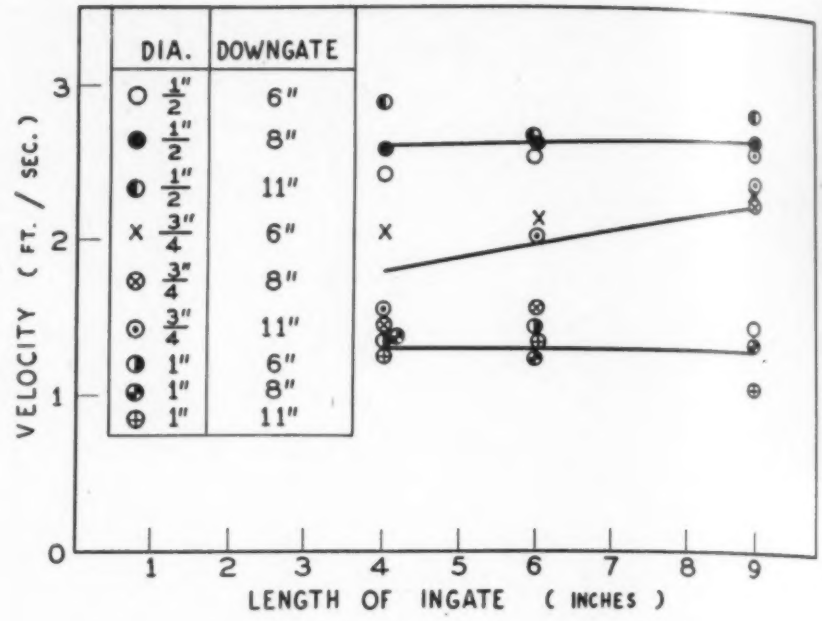
of the velocity equation assume a fluid running through a channel with impermeable walls.

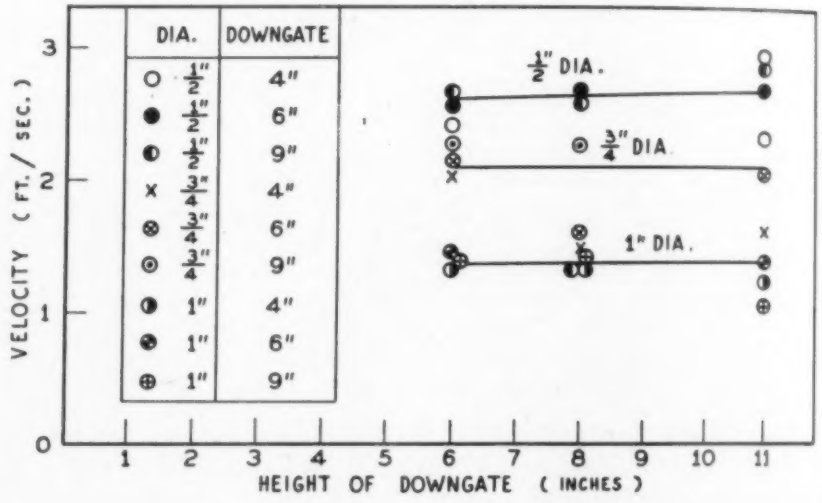
Varying the gate diameter had the most pronounced effect of any of the variables studied, the larger diameters resulting in lower linear velocities (Fig. 6). From consideration of the friction involved and the surface to volume ratios of large and small gates, the opposite effect might have been expected had the dimensions of the downgate itself been the controlling factor.

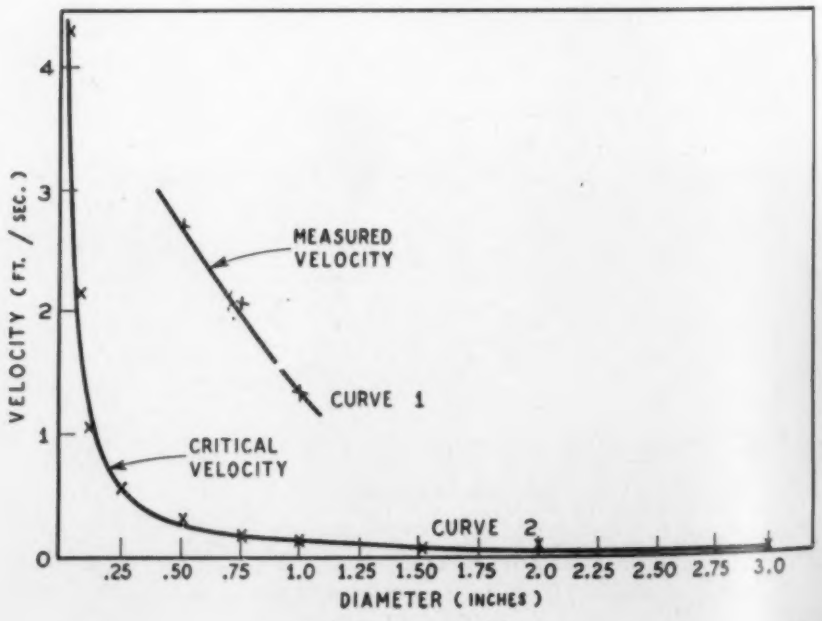
Experimental Errors. In general, control of the variables mentioned did not result in velocity changes greater than the experimental error. One source of error was variation in the height of the metal in the pouring basin. An effort was made to maintain the level at one in. above the top of the downgate. This level varied about ¼-in. Another source of error was the time measurement. An error of about ½-sec. may have been made in timing the insertion of the plug to stop the flow of metal.

Turbulence of Flow. The Reynolds' number has been taken as a criterion of turbulence, as previously explained. With the aid of this expression and with a Reynolds' number of 2,000, the critical velocity of bronze has been calculated and is shown in Fig. 6 by curve 2. This is the limiting velocity for a given diameter of pipe, above which velocity turbulence gradually sets in with any increase in velocity.

On the same graph, the actual or measured velocities are shown for various diameters of gate. The Reynolds' numbers for the actual conditions, with the respective gate diameters, are approximately ten times the Reynolds' number for the critical velocity of flow. This means that the measured velocities were nearly ten times higher than the ve-







Reading from top to bottom:
Fig. 4—Velocity of composition "G" metal vs. length of ingate.
Fig. 5—Velocity of composition "G" metal vs. height of downgate.
Fig. 6—Velocity of composition "G" metal vs. diameter of gate.

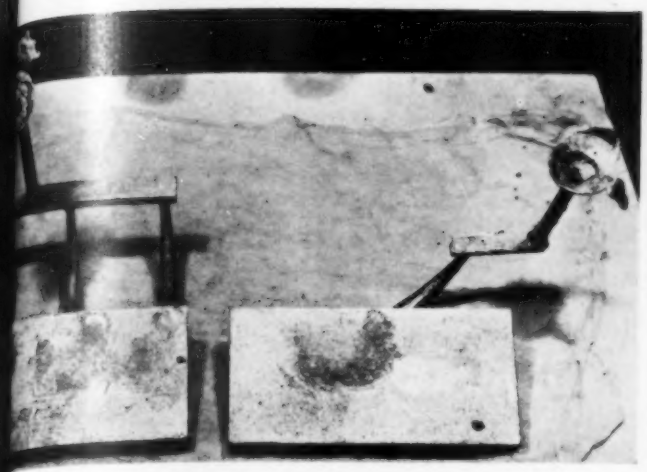


Fig. 7—Flat plates cast with gating systems employing "reversed" horn gates.

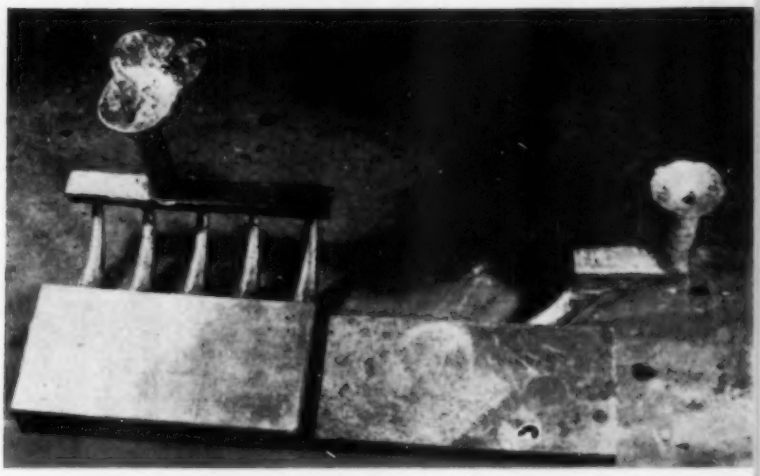


Fig. 8—Gating systems employing "reversed" and "regular" horn gates.

locities on the borderline of streamline and turbulent flow.

If curve 1 is extrapolated to curve 2, it is found that streamline flow theoretically should be obtained with gate diameters of approximately 1/16 or 1½ in. and proper velocities of flow, but these conditions obviously are impractical for foundry operation. The second part of this investigation concerns the practical significance of velocity of flow and its control by well-known methods of gating.

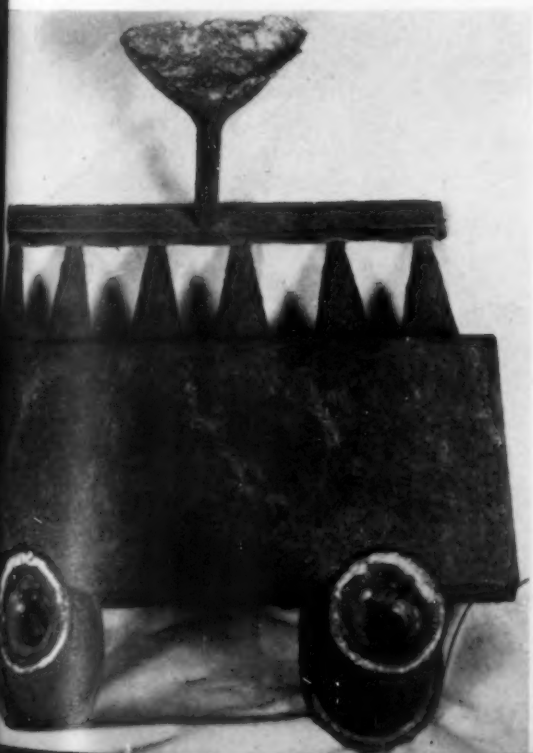
Practical Use of Controlled Velocity of Pouring

The velocity measurements described in the preceding section made it possible to predict the linear velocity of the three alloys flowing through the type of gate system used in this investigation (Fig. 2). Calculations of the Reynolds' numbers of these systems showed that the metal always was flowing in a turbulent manner.

Procedure. The effect of pouring under streamline and turbulent conditions on the dross-forming characteristics was then investigated with aluminum bronze, manganese bronze, and gun metal. Several 1x6x12-in. flat plates were cast with gating systems employing one, three, and five "reversed" horn gates and one "regular" horn gate (Figs. 7 and 8). These systems included various combinations of exit diameter and metal velocity, which resulted in a wide range of Reynolds' numbers.

Additional plates were then cast using a series of thin, flat, tapered gates which were expected to have the same dross-prevention properties as the five horn-gate systems without the inherent disadvantages of horn gates (Figs. 9 and 10). The plates were examined for surface defects and photographed (Figs. 7, 8, and 9). Tensile specimens were cut from the centers of the plates to show the effect of dross on mechanical properties (Table 4).

As shown in Figs. 7, 8, and 9, the surface of the casting gives an indication of the drossiness of the met-



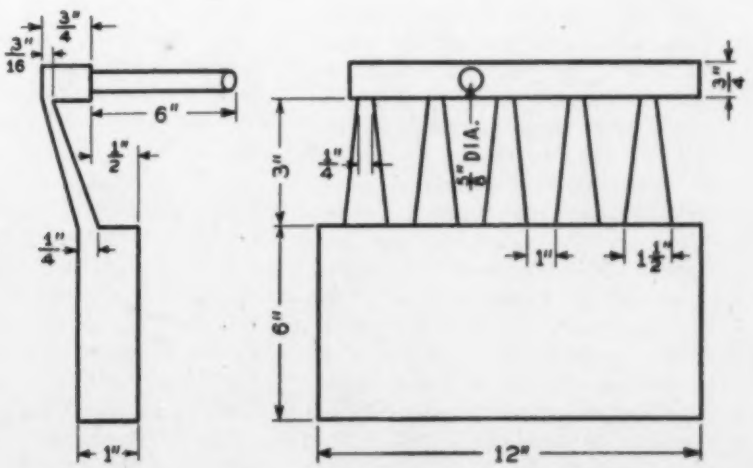


Fig. 9 (left)—Flat plate cast with thin, flat, tapered gate.

Fig. 10 (above)—Sketch of gating system employed in casting flat plate (Fig. 9).

al. Highly dross-forming alloys such as aluminum bronze show more dross on the surface of the casting when the pouring is more turbulent. The difference in the appearance of the surface of aluminum bronze in Fig. 7 is easily recognized when the turbulence of the metal was increased by using one horn gate in place of three horn gates.

Gun Metal Different

The situation is entirely different with gun metal because this is not a dross-forming alloy. As shown in Fig. 8, the velocity of pouring makes little difference in the surface of the casting of this metal. Even when the metal was sprayed into the mold cavity through the single horn gate (small diameter adjacent to the casting) no dross occurred on the surface of the casting.

The use of multiple horn gates to reduce the rate of metal entry into the mold cavity has many obvious disadvantages. This method was used only as a means to investigate the effect of rate of flow. It was found both empirically and by calculation for streamline flow that the product of the number of horn gates and their largest diameter (adjacent to casting) must be ten times the diameter of the downgate.

Aluminum Bronze

When this condition is satisfied, dross-free aluminum bronze may be poured with no dross formation on the surface of the casting. Fluted ingates of elliptical cross section may be used as shown on the aluminum bronze plate in Fig. 9 in order to produce a practically clear surface. The important part of the gating system is the point at which the metal enters the mold. At this point the flow of metal should be streamline for the best quality of castings.

Apparently, the turbulence at the constriction of the fluted ingate has little effect provided that the combined cross-sectional area is smaller than the cross-sectional area of the downgate and will permit the downgate to remain full when the metal is poured. The use of fluted ingates of elliptical cross section designed according to principles of fluid flow was believed to afford the most practical means for obtaining streamline flow of dross-forming alloys.

Tensile specimens taken from the aluminum-bronze plates showed wide variations in mechanical properties (Table 4). The plate cast under predicted streamline-flow conditions had uniformly high properties, while the other plates showed decreasing properties as well as an increasing scatter of results. Examination of the fractures of faulty test bars showed the presence of oxide inclusions which were produced from the turbulent flow of the metal.

Table 4

MECHANICAL PROPERTIES OF PLATES CAST WITH METAL POURED INTO VARIED GATING SYSTEMS

Heat No.	Horn Gates, no. Regular	Horn Gates, no. Reversed	Ingate Diam. xN / Downgate Diam.	Yield Strength, psi.*	Tensile Strength, psi.	Elongation, % in 2 in.
G 25-1	1		0.02	8,500	37,500	6
G 25-1A	1		0.03	27,750	70,000	16
G 25-1R		1	2.6	26,500	82,000	35
G 25-1RA		1	2.6	25,000	51,000	2
G 25-3R		3	4.0	24,000	70,500	19
G 25-3RA		3	4.0	25,000	78,000	26
G 25-5R		5	6.0	27,250	82,000	38
G 25-5RA		5	6.0	27,750	81,750	33
G 23-1	1		0.03	13,750	29,050	16
G 23-1A	1		0.03	11,500	18,550	1
G 23-5R		5	6.0	11,000	15,550	2
G 23-5RA		5	6.0	13,750	28,000	11
G 34-1R		1	2.6	23,750	50,000	3
G 34-1RA		1	2.6	27,500	73,000	7
G 34-3R		3	4.0	26,250	81,500	18
G 34-3RA		3	4.0	25,625	81,500	19
G 34-5R		5	6.0	25,625	65,000	14
G 34-5RA		5	6.0	26,000	66,000	10

*0.5% offset

Summary

a. Height of downgate and length of ingate appear to have little effect on the linear velocity of bronze in a gate of uniform cross section.

b. Aluminum bronze and manganese bronze form dross when poured under turbulent conditions, but are dross-free when streamline flow is attained.

c. Gun metal is less susceptible to dross formation than aluminum bronze or manganese bronze.

d. Turbulence is decreased by making the diameter of the gate at the junction of the gate and the casting greater than the diameter of the downgate.

e. Turbulence is reduced by using one or more tapered ingates, having the large end against the casting, and the total area of the small ends slightly less than the area of the downgate.

f. Plates of aluminum bronze cast with turbulent flow have mechanical properties inferior to those cast under streamline conditions.

g. Thin, flat, tapered ingates help promote directional solidification in aluminum bronze and manganese bronze.

The results obtained in this investigation and those reported elsewhere in the literature indicate that the following gating techniques are fundamentally sound:

A. For drossy metals (aluminum bronze, manganese bronze)

1. Use bottom gates wherever possible.

2. Take precautions in the use of horn gates so that the section size of the horn gate will not be large compared to the section size of the gated portion of the casting.

3. Gate thin sections by using thin, flat, tapered gates which impart all the advantages of horn gating without its harmful effect on feeding.

4. Keep the gating system full at all times. In a full gate dross will be retained by adhering to the point where it is formed. Care must be taken to avoid pouring dross into the gate from the crucible. If the gate is kept full, and no dross is added from the crucible, the use of dross sumps may be eliminated.

5. Use pouring basins where possible. These should be so designed that the metal is not poured directly into the downgate from the cru-

cible [illegible] into a reservoir offset from the g[illegible]e.

B. For nondrossy metals (tin bronze)

1. Direct riser pouring is permissible provided that care is taken to avoid excessive splashing or cutting of the sand.

C. General Recommendations

1. "Various portions of the casting should be so arranged that they are connected with a rising channel of increasing cross section."[8]

2. Gates should be placed to promote directional solidification and should be easy to remove.

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DISCUSSION

Chairman: D. FRANK O'CONNOR, American Saw Mill Machinery Co., Hackettstown, N. J.

Co-Chairman: HERMAN SMITH, Federated Metals Div., American Smelting & Refining Co., Pittsburgh.

MEMBER: For the benefit of those who might want to apply this data to practical problems in designing gates, the author spoke of the viscosity of molten metal in Reynolds' number. Would you assume it is constant for the various types of brasses and bronzes or could that be determined?

MR. HARDY: Some work has been done on the determination of viscosities of metals, but the high temperatures involved make precise measurement difficult. The *International Critical Tables*[10] list viscosities of several bronzes at several temperatures and these are all of the same order of magnitude. Possible ways to measure viscosity include measurement of the torque exerted upon a cylinder placed in a rotating crucible of metal, or measuring the damping effect on a pendulum swinging through a bath of metal. Clark, in his 1946 Exchange Paper with the IBF of England, lists a very good bibliography of various methods that have been tried. (See AMERICAN FOUNDRYMAN, July, 1946, p. 44.)

E. J. DUNN:[3] It is my understanding from what the author said that as you increase the diameter, the linear speed decreases. Now with constant heat and a constant source of metal such as you had, will you give an explanation of that? I had expected it to increase.

MR. HARDY: This puzzled us at first. We think that this anomolous result was due to the shape of the pouring basin, Fig. 2. We attempted to maintain one inch of metal above the gate. This was easily done with small diameter gates, but on larger diameter gates, at times a vortex formed over the down-gate so that the depth of metal there was less than one inch, so there was less pressure there to force the metal through the gate. I think that if we had used a deeper pouring basin this might not have happened.

MR. DUNN: Would you expect an increase then?

MR. HARDY: We expected an increase in linear velocity with large diameter gates, because for a given volume and weight of metal there is less surface on a large diameter. Therefore, there should be relatively less friction on large diameter gates. However, we found the reverse to be true.

GEORGE DALBEY:[3] I would like to ask if the Reynolds' number will be included in the final printing of the paper?

MR. HARDY: Most textbooks on hydraulics show how to use Reynolds' number. Walker, Lewis, McAdams and Gilliland[11] in the references at the end of this paper describe its use.

[3]Chase Brass & Copper Co., Cleveland.
[3]U. S. Navy Yard, Mare Island, Calif.

REPORT OF A.F.A. COMMITTEE ON

DIMENSIONAL STABILITY OF ALUMINUM CASTINGS

Aluminum and Magnesium Division
Casting Stability Committee

MANY SOURCES HAVE REPORTED that they have encountered dimensional instability in castings which have been machined. This is particularly troublesome to manufacturers of precision equipment, gauges, instruments, or aircraft accessories where dimensional accuracy must be maintained over a wide range of temperature such as from —75 to +350° F.

The difficulty does not merely lie in instability after the article is in use, change in dimension often occurs during machining or even during the drying of the paint used to coat the part.

The instability referred to is not the normal change brought about by the temperature coefficient of expansion, but is a permanent change which remains after the part returns to room temperature.

Causes of Instability

Castings may change as much as 0.001 in. per in., which may be far in excess of the required tolerance. Theories have been advanced to explain the cause of this instability and numerous suggestions made to overcome or avoid the difficulty.

Dimensional instability of castings is believed to be due to three factors:

1. Growth in volume.
2. Locked-up stresses due to casting or quenching.
3. Other contributing factors.

Growth in Volume.—Some aluminum alloys, particularly those containing appreciable amounts of silicon, will increase in volume on ageing at room temperature or at elevated temperature. This is called growth.

This is aptly demonstrated by the experiences with pistons in earlier days. Cast pistons were found to seize after being in use a short time. It was found that to prevent this the pistons had to be "degrowthed" or stabilized by proper heat-treatment. When such heat-treatments were used the pistons no longer seized.

In one investigation of stability, ¼-in. diameter diecast aluminum test bars were used to study growth. The bars of two aluminum alloys containing silicon were heated for 4 hr. at various temperatures. The changes in dimension are shown in Fig. 1 and 3.

All of the growth in these graphs is attributed to volume increase due to precipitation. The growth for these 7.5 per cent silicon alloys reached a maximum of 0.0010 to 0.0011 in. per in. when heated to 450° F. after casting.

Types of test methods and specimens for checking stability are described and curves are given to show relative stability of various aluminum alloys. Dimensional stability of aluminum castings is of particular importance to makers of precision equipment where dimentional accuracy must be maintained over a wide range of temperature.

Since maximum growth occurred at 450° F., other sets of these bars were heated at 450° F. for various times up to 16 hr. In Fig. 2 and 4 it can be seen that maximum growth was obtained within 4 hr. at 450° F., accordingly, some manufacturers have used the practice of aging at 450° F. for 5 hr. to obtain stability in castings.

By referring to the graphs it can be seen that if alloys such as those represented in Fig. 1 and 3 are subjected to paint baking at around 300° F., a growth of about 0.0005 in. per in. will occur if the casting has had no prior heat-treatment. Such a volume increase on finished machined parts may be grounds for rejection.

Aging Treatment

To avoid this from happening the part could be aged to its maximum growth at 450° F. before machining, in which case subsequent heat operations at temperatures lower than 450° F. would cause no volume change. As far as is known these growth curves are not reversible.

Alloys in the aluminum-zinc-magnesium series do not appear to be subject to such growth. Growth curves for two such alloys are shown in Fig. 5 and 6. From these curves it would appear that after any heating up to 700° F., growth will be less than 0.0002 in. per in. For Fig. 5 and 6, ½-in. diameter sand cast test bars were used.

A leading research laboratory also tested the aluminum-zinc-magnesium alloy of the composition shown

Presented at an Aluminum and Magnesium Session of the Fiftieth Annual Convention of the American Foundrymen's Association at Cleveland, May 7, 1946.

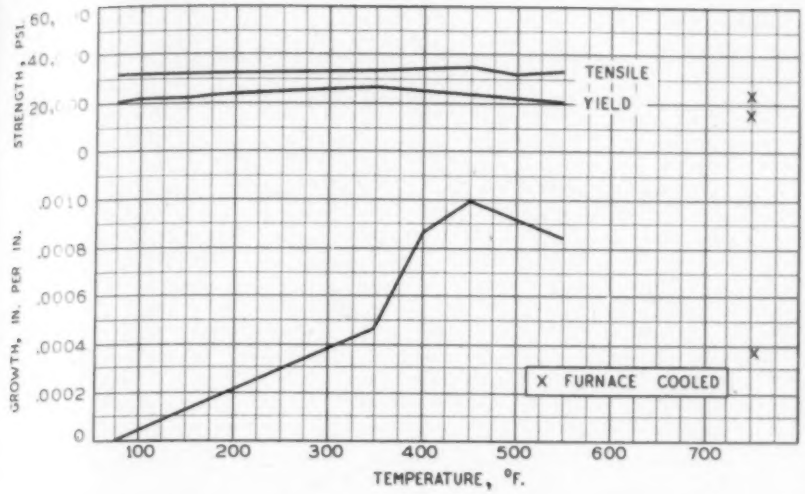


Fig. 1—Curve shows growth of sand-cast standard tensile test bars at indicated temperatures, 4 hr. at temperature, followed by air cooling. Aluminum alloy composition, per cent—Si, 7.5; Cu, 0.11; Mg, 0.31; Fe, 1.3.

in Fig. 5 for stability. Parabaloid sand castings 30½ in. in diameter and 6½ in. deep from mounting flange to top of rim of disc (Fig. 7) were cast.

The castings were aged for 10 hr. at 356° F. They were then machined and measurements were taken at 25 circumferential positions on each casting. These castings were then subjected to high and low temperature being brought back to room temperature after each exposure and as far as possible each castings was again measured in identically the same spots as for the previous measurements.

Temperatures ranged from —40 to +185° F. The average changes in dimension at each temperature are shown in Table 1.

These data agree closely with the results shown in Fig. 5 indicating maximum dimensional change to be in the region of 0.0002 in. per in.

Figures 1 through 6 also show the effect of elevated temperature heat-treatments on the physical properties of the alloys. In general it may be stated that ductility and impact properties are decreased by such heat-treatment.

Table 1

STABILITY OF AL-ZN-MG SAND CASTINGS AFTER MACHINING AND EXPOSURE TO INDICATED TEMPERATURES

Exposure Time, Hr.	*Temp., °F.*	*Dimensional Change, in. per. in.*
4¾	—40	0.00013
4	—20	0.00013
20½	0	0.00024
24	+150	0.000077
17¾	+170	0.00013
22	+185	0.000077

Average of 25 readings each.

Locked-up Stresses

Many cases of dimensional instability in aluminum castings have been found to be due to locked-up stresses within the castings.

Locked-up casting stresses are likely to occur when progressive feeding throughout the whole length of the casting has not been accomplished so that internal stresses are set up between sections of unequal thickness due to unbalanced contraction during solidification, or when the natural path of solid contraction is forcibly restricted as in sudden changes of direction.

Locked-up stresses sometimes occur in the casting due to unequal cooling in quenching from the solution heat-treatment.

Locked-up stresses if held in check by a strong enough metal section may constitute regions of concentrated tensile stress which will weaken the structure against shock or fatigue loads applied in the same direction as the stress.

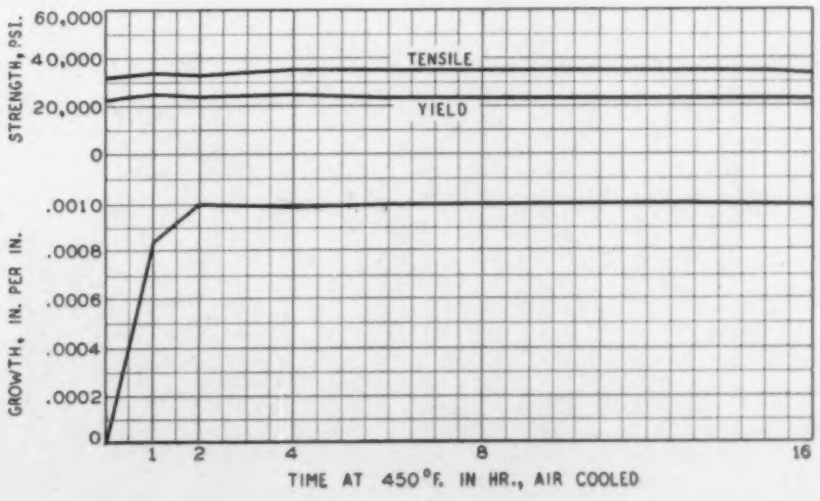
If the section is weakened by machining or by heating to elevated temperature, the locked-up stresses are relieved and the part distorts out of shape. This distortion can occur during machining or while the casting is in service, and is likely to destroy the desired alignment of the part or of an entire assembly.

Stresses Minimized

How to Remove or Minimize Locked-Up Stresses.—In cases where parts are subject to distortion during machining if used without previous heat-treatment having been applied, this is a good indication that locked-up stresses have been relieved by the machining operations.

In such cases it has often been found that heat-treating at 450-700° F. will stress relieve castings to a point where distortion is reduced to a minimum.

Fig. 2—Curve shows growth of sand-cast standard tensile test bars at 450° F. for length of time indicated, in hr., followed by air cooling. Aluminum alloy composition, per cent—Si, 7.5; Cu, 0.11; Mg, 0.31; Fe, 1.3.



There have also been cases where severe distortion occurred during machining as a result of locked-up quench stresses that occurred on rapid cooling from the solution heat-treat temperature. Such locked-up stresses can be minimized in three ways:

1. The more drastic the quenching, that is the colder the quenching medium and the greater the temperature from which the casting is quenched, the greater the tendency to quenching stresses. Therefore, to lessen the degree of such locked-up stresses the quenching rate should be decreased by quenching in boiling water, air blast, or still air.

2. Some castings are used in solution heat-treated condition with no elevated temperature ageing. In such cases any locked-up quench stresses are not relieved. A conventional aging treatment at 300-400° F. will tend to stress relieve such castings. The higher the aging temperature the more complete the stress relief.

3. Omit solution heat-treat and quench altogether by use of alloys that develop equivalent high strength without the necessity of heat-treatment or quench. Alloys such as those shown in Fig. 5 and 6 are in this category.

Other Stress Factors

Other Factors Contributing to Locked-Up Stresses.—These miscellaneous factors are grouped together since the preceding two classifications represent by far the greater part of the difficulties usually encountered. There have, however, been cases where fabricating processes have introduced locked-up stresses.

When tool pressures during machining are greater than the elastic limit of the cast section being machined, machining stresses occur. Later these stresses will cause distortion of the casting.

Support in Machining

Lack of support during machining was found to be a great cause of such stresses. The remedy is to back-up castings sufficiently throughout the area to be machined. Tool friction must also be avoided since it may heat the metal and thus reduce its strength. Another help is to add chucking lugs or bosses to the castings.

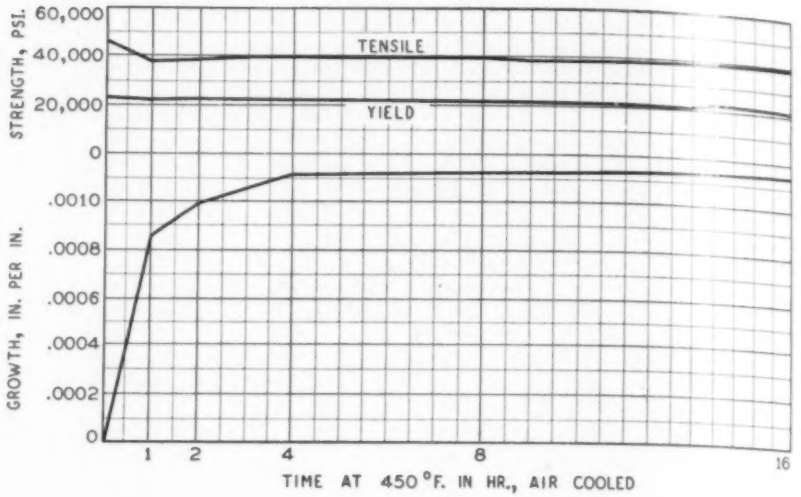


Fig. 4—Curve shows growth of sand-cast standard tensile test bars at 450° F. for length of time indicated, in hr., followed by air cooling. Aluminum alloy composition, per cent—Cu, 3.4; Si, 7.4; Fe, 0.66; Mg, <0.03.

Plastic deformation resulting in residual stresses may result from the straightening of castings which have warped in quenching or sagged at high temperature. To counteract this the usual practice is to straighten castings right after quenching, before the precipitation treatment, since the material is softest and most ductile then. The ensuing accelerated aging treatment then has a good chance to remove the resulting stresses.

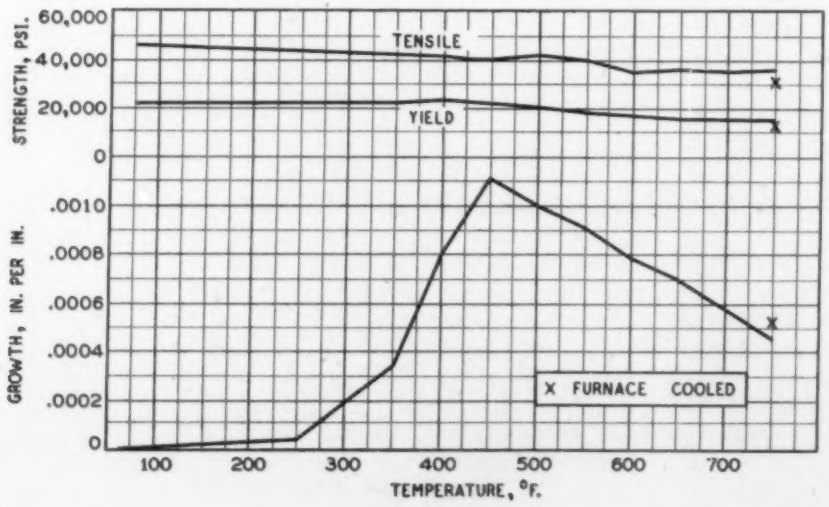
Small distortions after quenching can sometimes be remedied by heating the castings in boiling water and straightening them while hot. Another solution to this problem is the use of non heat-treated alloys such as those shown in Fig. 5 and 6.

Paint Baking

Sometimes it is found that paint baking at 300-350° F. will cause distortion in the casting. It has been shown, however, that a properly stress relief annealed casting will not distort during paint baking. This annealing should be done at a temperature above that at which paint baking will be done.

Welded castings should be cooled slowly if locked-up stresses are to be avoided. This can be accomplished by allowing the casting to cool down with the furnace, or by burying the casting in silica sand or some other material which will retard cooling,

Fig. 3.—Curve shows growth of sand-cast standard tensile test bars at indicated temperatures, 4 hr. at temperature followed by air cooling. Aluminum alloy composition, per cent—Cu, 3.4; Si, 7.4; Fe, 0.66; Mg, <0.03.



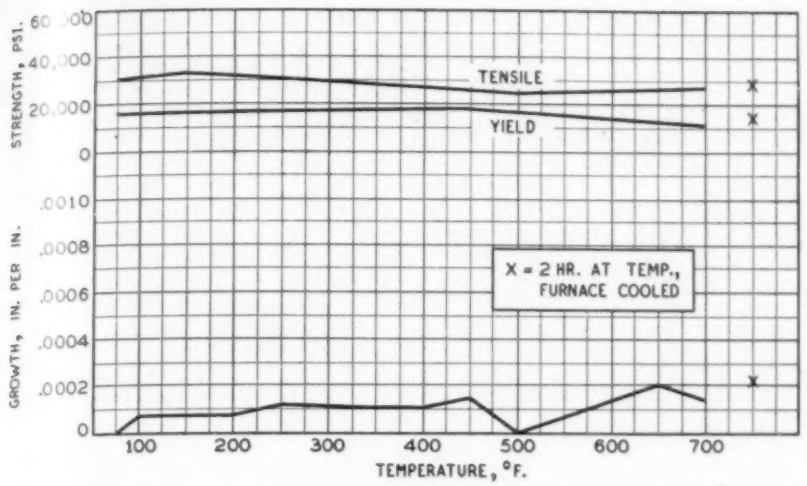


Fig. 5—Curve shows growth of sand-cast standard tensile test bars at indicated temperatures, 4 hr. at temperature followed by air cooling. Aluminum alloy composition, per cent—Zn, 5.5; Mg, 0.5; Cr, 0.5; Ti, 0.2.

or by building special insulated cooling chambers for use after welding. If stresses do remain after welding, they can be relieved by stabilizing the casting as outlined in preceding sections.

Stability Measurements

Measurements of Stability. — All foregoing information, with the exception of the data in Fig. 1 through 7 and Table 1, was compiled from the experiences of many engineers rather than from the experimental measurements.

The reason for this is that opinions of test pieces and the results obtained show no great agreement. Also, the methods of measuring stresses may vary considerably.

For example, there was excellent case of this sort which occurred with a magnesium castings. One experimenter used SR4 strain gages and subdivision technic to test residual stresses. By use of a dentist drill and hacksaw for cutting the specimens he found 4,000 psi. locked-up tensile stress.

Strain Gages Used

Other investigators repeated the test using the same type of gages but different cutting instruments, and found zero to several thousand psi. compressive stress in the castings. It was shown that the size of the cutting tool and the proximity of the cut to the gage both had large effects upon results due to the sensitiveness of the SR4 gages.

It was also suggested that although the technic using SR4 gages and careful cutting might be used successfully to determine dimensional instability as the result of locked-up stresses, it probably could not be used to determine volume changes due to precipitation hardening or changes due to growth at elevated temperatures because of the instability of the SR4 gages over a period of time.

It would appear therefore that the best specimen would be any intricate, thin section casting, and that great care should be used in measuring the residual stress.

Recent tests using brittle lacquer to detect residual stresses gives promise of a new technic that may be important in this field. The casting can be coated with brittle lacquer and allowed to dry. When a small hole is then drilled in the casting through the lacquer coating, stress patterns develop around the hole and are retained as cracks in the brittle lacquer thus giving a visible stress pattern.

The type of pattern developed indicates the type of stress, tension or compression which is in the casting. By varying the lacquer composition it is possible to obtain coatings which will respond to different degrees of stress. Thus it may be possible to test both the type and amount of residual stresses by using these brittle lacquers.

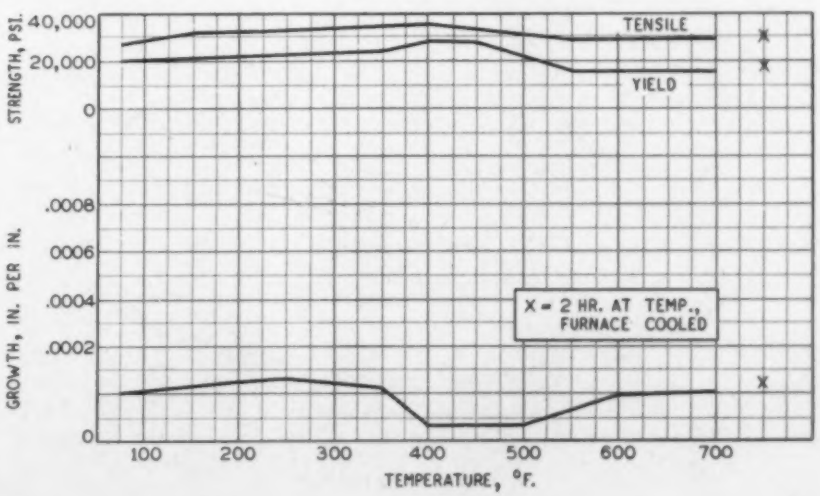
The tests described are all destructive tests and cannot be used and still retain the castings for use. Nondestructive tests using X-ray diffraction patterns have shown that residual stresses can be found by this method.

The perfection of this technic will avoid many cases of castings with locked-up stresses being machined, thus either making it possible to stress relieve the parts or to avoid expensive machining costs on parts that would be dimensionally unstable and would be out of alignment after machining.

Committee Personnel

Personnel comprising the A. F. A. Casting Stability Committee of the Aluminum and Magnesium Division are as follows: H. Brown, *Chairman;* D. Basch, E. J. Basch, C. E. Nelson and W. Sparrow.

Fig. 6—Curve shows growth of sand-cast standard tensile test bars at indicated temperatures, 4 hr. at temperature followed by air cooling. Aluminum alloy composition, per cent—Zn, 3.6; Mg, 1.8; Mn, 0.2; Cr, 0.2; Ti, 0.2.



DISCUSSION

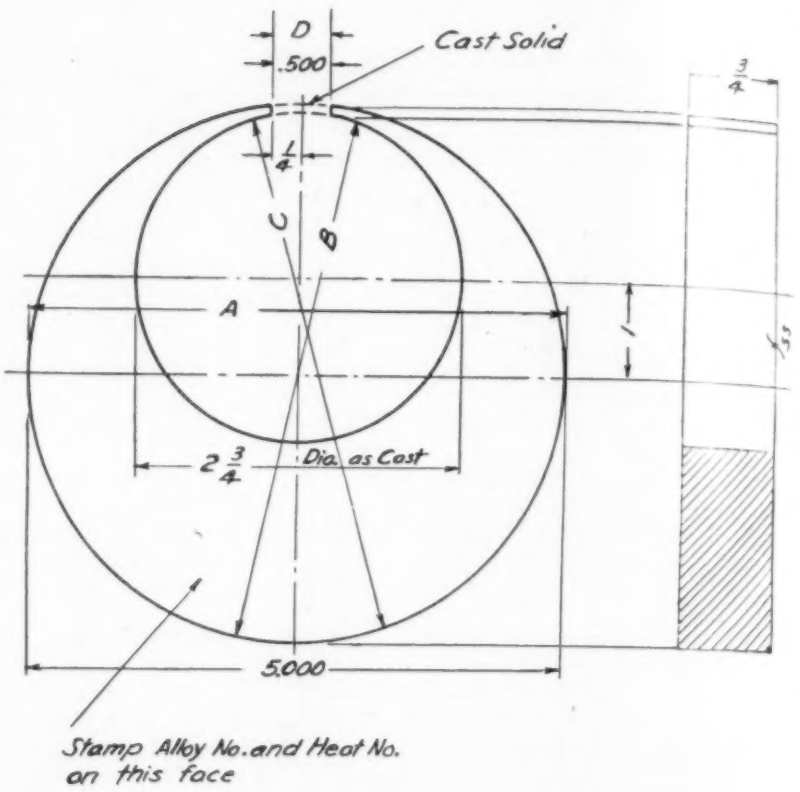
Chairman: W. E. MARTIN, National Smelting Co., Cleveland.

Co-Chairman: A. W. STOLZENBURG, Aluminum Co. of America, Detroit.

MEMBER: I am wondering whether this lacquer test might not be showing stresses in the lacquer as well as in the castings. Perhaps the lacquer is cracking of itself to some extent.

MR. BROWN *(author's closure):* The magnitude of the stresses in a brittle lacquer coating is low when the part, together with the lacquer coating, is held at a given temperature. Due to the creep phenomenon inherent in the lacquer, equilibrium is reached in the coating at a given temperature approximately one hour after an imposed load has been applied. Upon lowering the temperature below a critical value for a given lacquer coating, biaxial stresses are set up in this coating which may cause the coating to crack in a random pattern generally called "crazing."

A stress pattern developed on a part as the result of an imposed load would be influenced to a very small degree due to residual stresses in the lacquer if careful technique is followed. As the stress pattern is evaluated from the calibration bar, it would be reasonable to assume that the residual stress in the lacquer



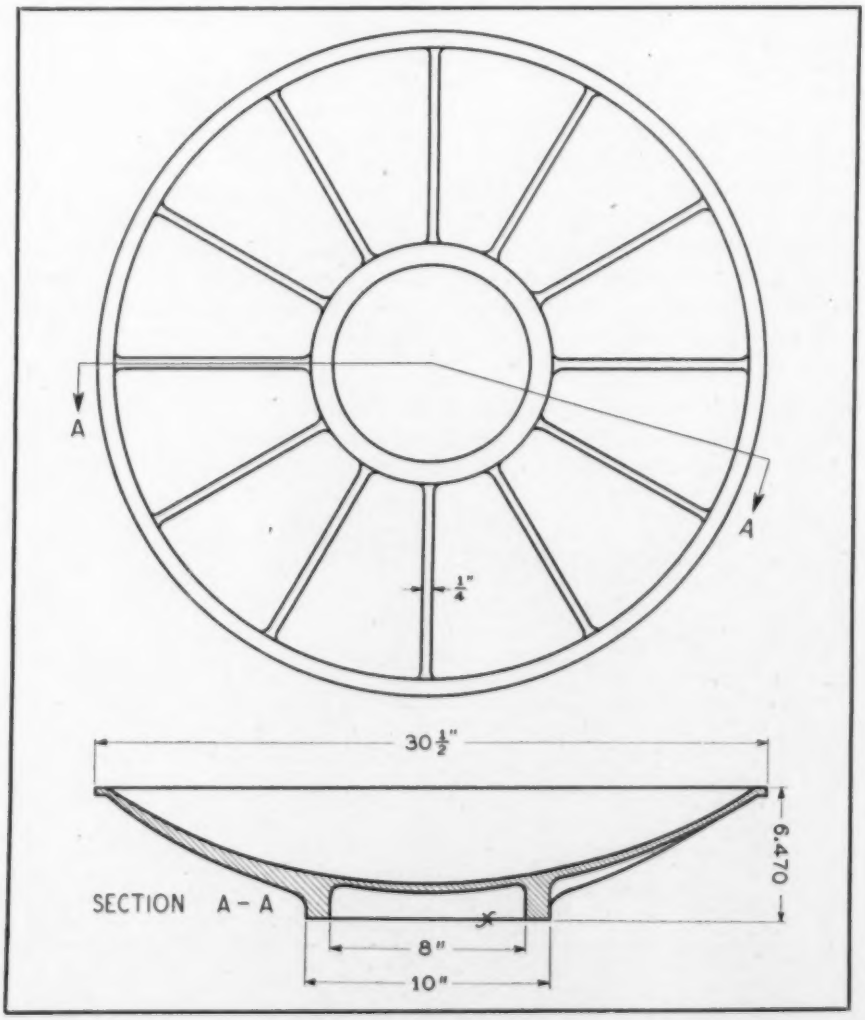


Fig. 7 (below)—Casting for stability measurements. Fig. 8 (above)—Distortion test specimen.

on a test piece would be of the same magnitude as the residual stress in the lacquer on the calibration bar when careful technique has been used.

Thus, residual stresses in the lacquer coating would add to the imposed stress due to load, with the result that the sensitivity as indicated by the calibration bar may be raised or lowered from a theoretical value. Thus, the influence of residual stress in the lacquer coating would only affect the sensitivity and, as we are measuring rate of strain to cause rupture in the coating, a reasonably close co-relation should exist between the strain causing rupture in the lacquer coating on the calibration bar and the lacquer coating on the part undergoing test.

The above applies, of course, where a given temperature and humidity condition prevails. Upon a change of temperature or humidity, if the temperature is decreased, thermal stresses are set up in the lacquer coating. This phenomenon is used to produce an overall lacquer coating pattern where it is desired to determine the direction of stresses throughout the piece and where the imposed load has not been of sufficient magnitude to cause the lacquer to crack.

In this case, a qualitative analysis only can be obtained. This is generally accomplished by directing a cool stream of air or by swabbing the part with water at a temperature several degrees below the surrounding air temperature. This technique is also used in determining on a qualitative basis residual stresses in machine members.

CLOTH TYPE DUST COLLECTORS

George A. Boesger
The W. W. Sly Manufacturing Co.
Cleveland

Cloth type filters collect finely divided dry dust at economical cost. Two general designs are available—the tube or bag type and the screen type. Mechanical members which support the cloth envelopes are connected to a rapping device for dislodging the dust pack from the cloth surfaces. Hoppers are provided below the screens into which the collected dust is deposited.

CLOTH-TYPE filters remove dust from the conveying air by flowing the air through specially designed filter fabric. Several variations in design are available, but the operating principle remains the same. The cloth-type filter is in general use primarily for the collection of finely divided, dry dusts, and accomplishes high recovery at economical cost. With the exception of the far more costly electrostatic precipitator, no dust collecting device can equal the cloth collector in efficiency.

Collecting only the larger particles of dust gives the appearance of being effective because it does eliminate a larger part of the visible nuisance. However, there are three reasons for collecting the fine dust:

1. Coarse dust particles settle near the point where the dust is created, but the finer particles float in the air and become a nuisance and create damage at considerable distance from the source.

2. Heavy dust particles may settle on a piece of machinery, but only the very fine particles work their way into bearings to cause irreparable damage.

3. It has been scientifically determined that larger particles of dust are stopped in the nose and throat of an exposed person and that only the finest particles find their way into the lungs.

The cloth-type collector is most adaptable as it will handle dust of any size. No dust is too coarse or too fine for the cloth collector. A cloth filter consists of a tight casing enclosing the filter fabric, and so arranged that all incoming air must pass through the fabric to the collector outlet.

Presented at a Session sponsored by the Safety and Hygiene and the Plant and Plant Equipment Committees at the Fiftieth Annual Meeting of the American Foundrymen's Association at Cleveland, May 6, 1946.

Dust Particles Filtered

Upon entering the case of the collector a reduction in air velocity occurs, serving to drop the coarser particles of material from the air stream. The remaining finer dust is conveyed to the cloth surfaces where it is filtered from the air stream, the cleaned air continuing on through the cloth to the collector outlet. The dust pack, or mat, thus formed on the cloth surfaces assists materially in improving the over-all recovery efficiency.

The collector casing is provided with hoppers into which the collected dust is deposited. The filter is fitted with a shaking device designed to dislodge periodically the dust from the cloth surfaces. This shaking operation occurs with air flow shut off, since the air pressure would otherwise tend to hold the dust pack against the fabric.

Installation Types

A connection may be made from the collector to the discharge side of the exhauster as a "blow through," or pressure-type installation; or to the inlet side, designated as a "draw through," or suction-type installation. The latter arrangement is most generally adopted because in this case the exhauster, or fan, handles only clean air and is not subject to abrasion by the dust particles.

Cloth Filter Designs

Two general designs of cloth filters are available—the tube or bag type, and the screen type. In the cloth tube or bag type, the filter fabric takes the form of a series of tubes or bags, open at one end and closed at the other.

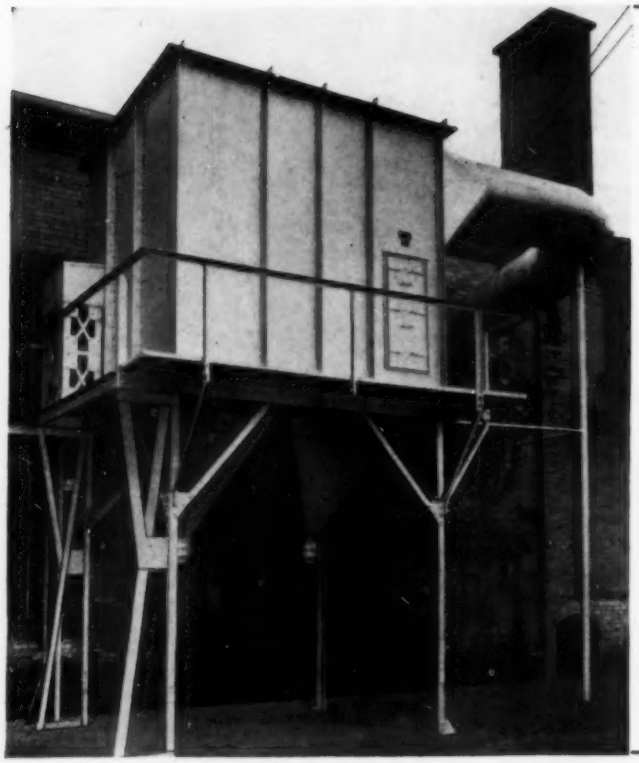
These tubes, or bags, are suspended vertically in the casing with the bottom, or open end, tightly sealed to thimbles in a plate located near the bottom of the casing. The dust-laden air enters the casing below the seal, or thimble plate, and passes upward through the cloth inlets (Fig. 1).

Dust is precipitated on the inner surfaces of the tubes, or bags, and the cleaned air continues on to the outlet at the side, or top of the casing. The tubes, or bags, are attached to a shaking device, usually connected at the top of the tubes, which is operated periodically to remove the dust pack from the inner surfaces of the fabric, from which it drops to hoppers which form a part of the collector casing.

Cloth Screen Collector

The cloth screen type collector is of more recent development. In this type of filter the cloth appears in the form of comparatively flat envelopes suspended horizontally and side by side in the metal casing, in a manner which permits installing a maximum cloth area in a minimum size of casing.

Open ends of the cloth envelopes are sealed to each other or to the casing, definitely dividing the casing into a dust chamber and a clean air chamber. The dust-laden air enters the dust side of the casing and must



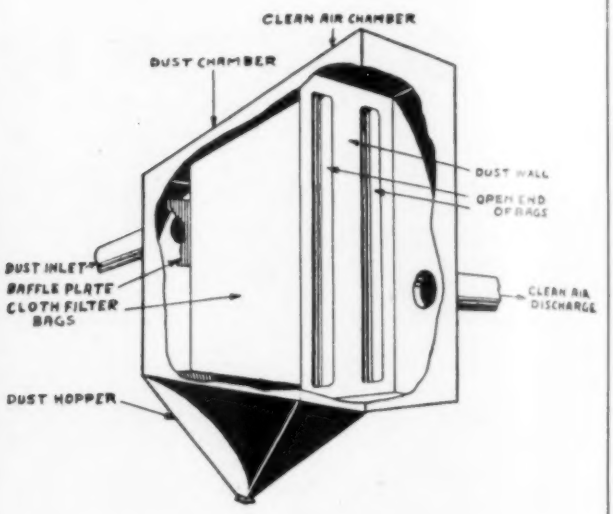


Fig. 1 (Left)—Typical cloth type dust collector installed outside of a foundry. Collected dust is withdrawn from the hoppers. Fig. 2 (Above)—Diagram shows passage of air through a cloth type dust collector.

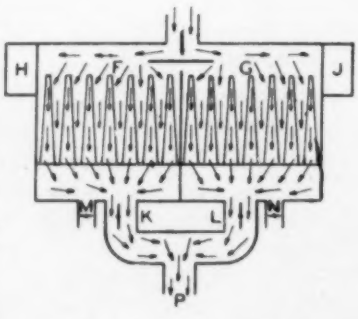
pass through the cloth from the outside to the inside of the envelopes. The dust is collected on the outside of the envelopes.

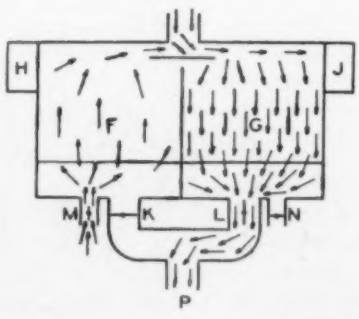
Mechanical members which support the cloth envelopes are connected to a shaking or rapping device, for dislodging the dust pack from the cloth surfaces. Hoppers are provided below the screens into which collected dust is deposited.

Many refinements in cloth collector design have resulted from extensive operating experience. They have brought about longer life; greater capacity per unit of space occupied; ease of access; reduced maintenance and repairs; and more effective cloth cleaning.

As the building up of the dust mat on the cloth surfaces results in a gradual increase in collector resistance, it is necessary periodically to vibrate the cloth surfaces to reduce the dust pack and the resistance to air flow. With the more common collector arrangement, known as the intermittent type, the cloth is cleaned by periodic operation of the shaking device, with the exhauster shut down.

Frequency and duration of these cleaning periods is determined by specific conditions. For average problems with average dust loading, these cleaning periods of 5 to 10 min. may take place after each 4 to 5 hours' operation. The cleaning period frequently can be decreased





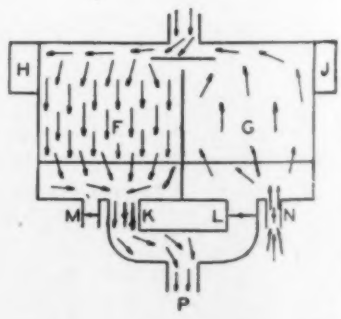
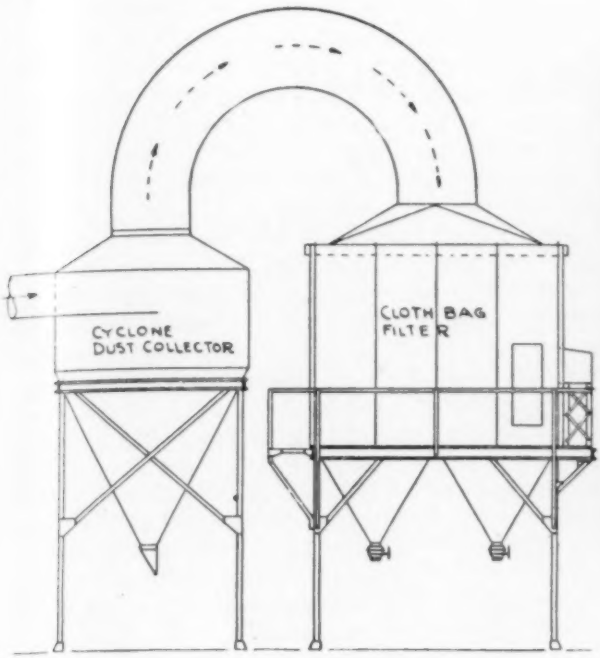


Fig. 3 (Left)—Diagram of air passage in an automatic-continuous dust filter. In this diagram both filter sections F and G handle the dust-laden air. Fig. 4 (Center)—In this diagram main damper K is closed, small damper M is open for reverse air-flow through filter F, to assist shaker drive H in removal of dust from bags. Fig. 5 (Right)—In this diagram main damper L is closed, small damper N is open for reverse air-flow through filter G, to assist shaker drive J in removal of dust from bags.



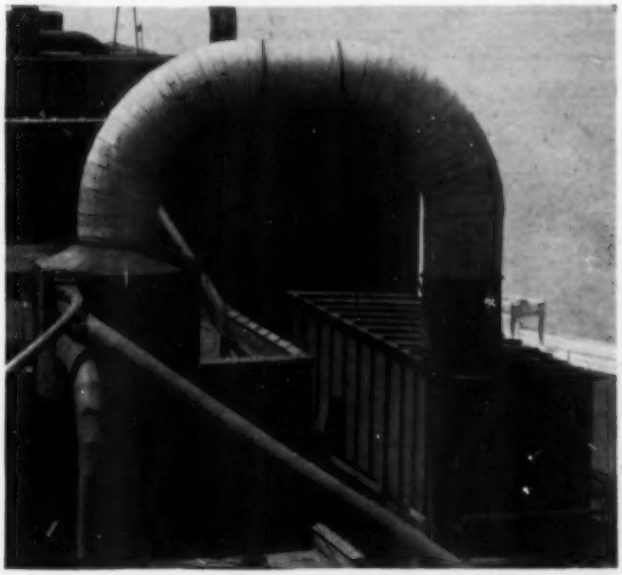


Fig. 6 (Left) with Fig. 7 (Right) show a cloth type dust filter with cyclone type "precleaner".

by installing additional presettling means ahead of the cloth collector to reduce the dust loading on the cloth.

Intermittent type of collector is economical in first cost and maintenance for use with operations and processes where short shutdowns for cloth cleaning are not detrimental. These cleaning periods ordinarily are arranged to coincide with the lunch periods or shift changes.

Continuous Type Collectors

Continuous type cloth collectors are used for process applications where these shutdown periods for cloth cleaning cannot be permitted, and continuous operation is essential (Fig. 2). The continuous type collector consists of two or more cloth filter compartments, or sections, each fitted with an individual shaking device and arranged with gates or dampers to permit alternately isolating each compartment from air flow to permit cloth cleaning.

This control of air flow and shaking of the respective compartments may be accomplished in regular cycle through manual operation, or by automatic control apparatus. In some instances, involving the handling of high dust concentrations, an extra cloth filter compartment, or section, is provided over and above those required, with the control so arranged that one compartment is constantly shut off for cleaning.

Continuous type collectors are substantially more expensive from the standpoint of first cost, operation and maintenance than the intermittent type collector. This is particularly true when the installation is of the automatic type, necessitating electrical control and timing apparatus.

Properly applied and maintained, cloth filters will provide effective suppression of finely divided as well as coarse industrial dusts, but several factors must be given consideration in their application. The more important limiting factors are temperature, moisture, and chemical qualities of both the dust and the conveying air or gas.

Maximum Operating Temperature

The limiting temperature for cotton filter fabric is approximately 175° F. This limit may be somewhat extended through the use of wool filter fabric. The use of asbestos, glass cloth or plastics does not appear to be satisfactory at this stage of development, because proper physical qualities have not been developed to withstand the requirements of this service.

When containing moisture, the conveying air and collector temperature must be maintained above the dew point to prevent condensation within the collector, which might otherwise result in sludging the dust on the cloth surfaces. The chemical properties of the air, or gas, and the materials being handled, must be analyzed to determine if a deleterious reaction on the fabric or the collector parts is likely.

Cloth Filter Factors

Other factors governing the cloth filter recommendation for each application are the air volume handled; nature, characteristics and amount of material to be collected; and range of dust particle size and specific gravity. The size of a cloth filter, i.e., the filtering area, is governed by consideration of these factors. Filtration velocity used with intermittent type collectors will vary with specific problems of average nature in the range of from one to four lineal ft. per min. This is more commonly termed "ratio," meaning one to four cu. ft. of air per min. passing through each sq. ft. of filtering area. With continuous type collectors, the "ratio," or filtration velocity, frequently is increased.

High dust loadings are reduced by presettling devices installed ahead

of the cloth filter. This presettling may be accomplished by a centrifugal or cyclone collector, settling or baffle chamber, or other devices effective for partial precipitation.

Wire Mesh Spark Screen

When connected to systems where there is a possibility of sparks or incandescent particles being drawn into the system, a wire mesh spark screen is installed in a compartment ahead of the cloth filter section for trapping and extinguishing these particles.

This device is also useful for removing or trapping nominal quantities of lint or fibrous material from various classes of operation. The wire spark screen frequently is fitted with a rapping device operated from outside the casing for shaking down accumulated dirt and lint. As a further protection in applications where sparks may occur, the filter fabric may be treated with a special flameproofing solution.

When handling dusts of explosive or combustible nature, cloth filters should be fitted with vents, or stacks, arranged automatically to open and relieve pressure, should an explosion occur.

Discharge of dust from collector hoppers should be given careful consideration to prevent dispersing fine dust to the working room or the surrounding territory during the emptying operation. Standard cloth collectors are provided with hoppers having dust-tight discharge gates fitted with canvas tubes. By proper manipulation of the canvas tube the dust emitted during the emptying operation can be kept at a minimum.

This is accomplished by twisting the tube before opening the hopper gate, then gradually untwisting it and allowing the dust to slide slowly through the tube to the receptacle or truck. The hopper gates are, in some instances, fitted with clamps for the attachment of bags to which the collected dust is discharged directly. Receptacles, trucks, etc., with tight covers connected to the hopper gates by removable spouts are sometimes used.

A practice of considerable merit is to enclose entirely the space below the collector casing surrounding the hoppers. This prevents air currents from distributing dust which may be emitted during the hopper emptying operation. Access to the hopper enclosure is had through a tight door in the end or side. As a further step, the enclosure can be provided with an exhaust connection to maintain a slight negative pressure therein and thus prevent emission of dust.

When mechanical or continuous dust discharge is desirable, the collector hoppers are fitted with rotary air-lock type valves, which permit dust discharge but prevent air leakage, although the collector may be in operation. The rotary valves ordinarily discharge into a tight conveyor for returning the material to the process or to a central disposal point. In some instances the dust is discharged from the hoppers into a water sluicing system and discharged as sludge into settling pits, lowlands or sewer.

Acknowledgment

The writer wishes to acknowledge assistance from Mr. Andrus, of the American Foundry Equipment Co., and from Mr. Vedder, of the Pangborn Corp., in the preparation of this paper.

LINING AND PATCHING of CUPOLA REFRACTORIES

E. J. Lally
Forest City Foundries Co.
Cleveland

IT IS A KNOWN FACT that the cupola is the heart of the foundry; therefore, for good operation it is essential to use the best refractories available with the most advanced practices of lining and patching.

The shell of the cupola should conform to a perfect circle as nearly as possible. Any irregularities allow excessive expansion and contraction between the joints of the lining bricks.

In lining the well of the cupola, any of several methods may be used. Selection of a method depends upon the size of the shell and the well diameter desired. One method is placing of the 2- or 2½-in. brick next to the shell and cupola block in front, the cupola block being exposed to the metal. In another method the cupola block is placed next to the shell and the 2- or 2½-in. brick exposed to the metal.

This latter method offers the advantage of allowing replacement of the straights with but little effort. Another method utilizes 4-in. circle blocks next to the shell and faced off with cupola blocks. Still another is in the use of two cupola blocks, one in front of the other. In still another method, 2- or 2½-in. straights alone are used, depending upon the well diameter desired. The final method, say in a No. 3 cupola, is the use of a cupola block alone in the well.

All of the foregoing methods require the use of a soft, well-mixed mud between the joints of the bricks, the bricks themselves being dipped in a "clay soup." In all of the methods mentioned, "soup" is poured freely between the shell of the cupola and the bricks next to the shell to fill any voids.

It should be mentioned at this point that in lining the cupola well, the joints of the bricks of the course next to the shell and the joints of the face brick should be staggered. This helps to eliminate any possibility of molten metal seeping through the bricks to the shell. After the well has been completely lined, the face brick are coated thoroughly with a commercial refractory wash.

Having lined the cupola well, the next step is to properly fit the tuyeres in place. In the case of the individual tuyeres, it is a good policy, if possible, to bolt the tuyeres to the shell. Caution should be used to insure that a solid, tight fit is obtained between each tuyere, especially if the tuyeres are not bolted to the shell.

In the case of the continuous tuyere, it is important that the shell at this point conforms to as perfect a circle as possible. This is to insure a close fit so that no air is lost in passing from the windbox through the tuyeres.

Proper selection of methods and materials for cupola lining and patching has much to do with the efficiency of the foundry melting. Good materials properly applied are important factors in cupola melting practice.

The type of lining used between the tuyeres and to a point 2½ ft. below the charging floor depends upon the size of the shell and the diameter that is desired.

It has been advantageous in the lining of a cupola to have a ring bolted to the shell about 2 ft. above the melting zone, so that when only the melting zone is relined the brick above the ring is held in place. The number of rings in the stack of a cupola depends upon the height of the stack.

The three courses of brick immediately above the tuyeres are placed so as to form an overhang of 2 to 3 in. over the tuyeres. This aids in preventing the tuyere plates from being burned off by the molten metal as it passes by the tuyeres, and also helps to keep the metal from dripping into the tuyeres. This overhang can be obtained by an extra 2- or 2½-in. straight being placed in the three courses.

The section between the overhang and a point 2½ ft. below the charging floor is lined according to the diameter desired.

In one lining method, a 2-in. straight is placed next to the shell and a cupola block in front of the 2-in. straight. Another method utilizes two cupola blocks next to each other to form a course. If the cupola is small, the stack short, and only small tonnages melted, a single cupola block next to the shell is sufficient lining. Here again, the bricks

Presented at a Refractories Session of the Fiftieth Annual Meeting, American Foundrymen's Association, at Cleveland, May 7, 1946.

are dipped in a "soup" and the joints are sealed with a well-mixed clay. Any void between the back lining and the shell is eliminated by pouring "soup" between the shell and the back lining.

Again it is important when lining the stack to stagger the face block in regard to the back lining to prevent the joints from coinciding. This practice is continued up to 2½ ft. from the charging floor.

From 2½ ft. below the charging floor to 2 ft. above the charging floor a refractory block is used as a back lining. The face block is an iron brick. Here the iron brick is used as a face block to withstand the punishment received when the charges are thrown into the cupola.

These iron brick usually are hollow, so that "soup" can be poured into them to allow for contraction and expansion. The last row of iron brick above the charging floor usually is bolted into place. The course of bolted iron brick acts as a support for the arch brick used in the top part of the stack. The arch brick between the iron brick and the top of the stack seldom has to be replaced because it serves only to protect the shell from the flame of the cupola and has no other function.

After the cupola has been completely lined, a coke fire should be placed in it and allowed to burn for 24 hr. This will insure a completely dried cupola.

Patching A Cupola

The amount of work necessary to patch a cupola depends upon the amount of burn-out, which in turn depends upon the type of operation. If the cupola is charged unevenly, air distribution poor, melting rate high for the cupola size, or if the patching has been done sloppily, the amount of burn-out is affected accordingly. Therefore, in cases of consistenly uneven burn-out it is the responsibility of the melter to find out which of the foregoing factors causes the trouble.

After the drop has been cleaned up, the inside of the cupola should be sprayed thoroughly with water. A much cleaner job of chipping is obtained by wetting the inside of the cupola.

Once the cupola has been chipped, the first thing to consider is repair of the well. If the method of placing the cupola block next to the shell and the 2- or 2½-in. straight next to the metal is used, it must be made certain that the 2- or 2½-in. brick are in good shape. If not, they must be replaced.

If the cupola block is exposed to the metal and the well is in need of repair, a rammed mix that is fairly dry should be rammed into the eroded area, preferably with an air hammer. After the well has been repaired it should be washed with a clay soup or a refractory wash.

Materials Used

A number of different refractory materials are used for cupola patching. Some foundries use sandstone cut with the grain; others use sandstone cut against the grain, Newcastle firestone, fire brick, and some use their own recipe for a rammed mix. Any of these materials may be used as long as it gives good results.

It is important to note here that no matter what type of stone is used for patching, it should be dried so that all moisture is driven out. This will prevent "popping" of the patching material when bed is burned in.

After the melting zone of the cupola has been chipped *clean* and the well has been repaired, the next step in patching is to form a ring immediately over the tuyeres to act as a ledge or support for the remainder of the patchwork. This ring consists of small pieces of stone and well-mixed mud.

The type of mud used for patching varies with every foundry. Some of the mixes used are fire clay and new molding sand, fire clay and silica sand, silica sand and gannister, and fire clay alone. It is important to remember that no matter what mixture is used for patching, it should be mixed 18 to 24 hr. ahead of time.

Once the ring over the tuyeres has been made, the melting zone can be patched. If an excessive burn-out occurs in one area, it is better to tear out the lining to the shell and put in a new lining in that area. It is much more practical to do this than to try and patch over the area with patching material.

The ability of a patcher to work the patching stone into the variously burned areas is of great help in maintaining the original diameter of the cupola. It is up to the patcher to place a good layer of clay on the lining to hold the patching stone in place. Any areas that are burned deeper than others have to be patched with a thicker piece or pieces of stone so that no depressions are left in the patching.

If a certain area has only a slight burn-out, it is well to patch with a thin piece of stone so that the original diameter is not reduced. Finally, in an area that has had practically no burn-out it is wise to merely wash over it with a clay soup and leave it alone, because sooner or later it will burn off.

The writer would like to mention that any voids formed by the joints of the patching stones must be filled in with small pieces of stone. This helps to keep the larger stones in place and prevents them from slipping.

After the work of patching the cupola has been done, the patched areas should be washed with a thick clay soup.

In conclusion, if two cupolas are available and they are run alternately, it is preferable to dry the patching for at least 2 hr. by means of air and gas torches or a wood fire.

DISCUSSION

Chairman: C. E. BALES, Ironton Fire Brick Co., Ironton, Ohio.

Co-Chairman: C. S. REED, Chicago Retort & Fire Brick Co., Chicago.

MEMBER: What is the patching diameter?

E. J. LALLY: The patching diameter is 8 in.

MEMBER: How do you perform this patching operation?

MR. LALLY: In our practice, we operate two cupolas alternately. After the man has finished his patching job, he puts up the bottom doors, rams in the sand bottom, and puts in the bed, all in one operation before he goes home that evening. The cupola is ready to start the next morning. At 12:30 that night we start a small fire with the torches. There are four portholes in the walls of the cupolas through which we put burners. We burn those burners about a half hour every night, just to start the coke.

With this particular type of operation or practice we have been getting a satisfactory drying of our patch and a hot cupola the next day. If you had only one cupola it would not get a chance to dry.

CO-CHAIRMAN REED: I have heard it recommended but ran into some difficulty on this question of back-up. I know of a case where a new cupola which was lined with a 6-in. block backed up by another 6-in. block split its rivets on the first heat. The solution to that problem, I was told, was the insertion of cardboard spacers at least four or five to the ring on the outside lining. Have you had any experience either with the 6-in. or

4½-in. backup that you would like to bring out?

MR. LALLY: It has only been a short while that we have had this No. 8 cupola in operation. We did not know exactly what lining to use. We did not know how many tons a day we were going to run because we put a new conveyor line in. I was told the requirements were 11 tons an hour. I decided that I would try two 6-in. cupola blocks in the melting zone. In the well, I am using a cupola block and 4-in. circle block.

I was amazed when the first heat was tapped. We had hardly any burnout at all. It was very low and to this day that cupola burns low. It does not burn high. At times it is somewhat deep, depending upon the tonnage we take out of it. I took out 107 tons one day. That cupola is producing approximately 11 tons an hour, 8-10 hr. a day.

No, I have never heard of the experience Mr. Reed spoke of. I have had cases of a No. 6 cupola just using the 2-in. block next to the shell. Cupola block was exposed to the metal in the melting zone, and the metal went to the shell before the heat was over. I found out that the trouble was in the brick.

I would like to mention again that the 2-in. block in the well is a help. I know it is conventional to put the cupola block next to the metal and when it gets bad, to put in a new well. In my No. 6 cupola we put the 2-in. or 2½-in. brick exposed to the metal. When it gets bad my patcher asks me to look at it. If it is bad enough, he chips it right out. I have seen him put ten brick in at one time.

T. W. CURRY[1]: Are you putting those brick edgewise? Is there molten metal and slag against the 4½-in. face of the brick?

MR. LALLY: That is right.

J. A. BOWERS[2]: I do not know how many times we have relined our cupolas, but we have never tried to put a back-up or any other means of taking care of expansion when using cupola blocks. As a matter of fact, we put the brick against the shells as close as we can. Of course, you can not all the time because there may be a rivet or bolt head or an angle holding up the cupola block, and we actually pour the soupy fireclay behind the brick just exactly as Mr. Lally described. We have never ripped any of our shells. If you were to get a brick that had a high expansion coefficient, you probably would. As a matter of fact, we did rip a shell one time with sandstone but never with clay cupola block. Frankly we favor, as far as the melting zone is concerned, on our 72-in. cupolas putting one row of 96-72 cupola block just above the tuyere and then we drop back to a 96-78, dry the lining out and patch the 30 in. just above the tuyeres with regular 9-in. straights to close all joints, if possible, on a brand new lining.

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[1]Lynchburg Foundry, Lynchburg, Va.
[2]American Cast Iron Pipe Co., Birmingham, Ala.

DUTIES AND FUNCTIONS OF THE INSPECTION DEPARTMENT AS SEEN BY MANAGEMENT

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WHAT IS MANAGEMENT'S ATTITUDE toward Inspection? There are obviously three schools of thought on this subject, namely, (a) Management that has no Inspection Department, (b) Management that has an Inspection Department and recognizes and supports it only as a necessary evil, and (c) Management that has an Inspection Department and supports it.

First, let us consider the plant that has no Inspection Department because of decision of Management. This situation may be both good and poor. If Management feels that Inspection is unnecessary and does not have an Inspection Department because it is not interested, then it is the author's opinion that Management has much to learn. This could be looked upon by many as an undesirable situation.

Quality Considered

On the other hand, if in that plant Management has succeeded in building up an organization of outstanding supervisors that think in terms of Quality as well as Quantity, then that would approach ideal manufacturing, and Inspection might not be necessary. Such a situation should not be criticized unfavorably, but should be commended. Very few plants, however, have achieved that degree of perfection.

Secondly, it is an unfortunate situation where through tradition Management has an Inspection Department, but does not support it, feeling that an Inspection Department is a necessary evil. In such a case the author would be inclined to feel that an inferior product would be produced, but if difficulties arise because of the inferior product, without a doubt then the Inspection Department will be criticized. That is all the value there is in such a case, and the existence of the Inspection Department has been justified as a good place to put the blame.

The author recalls taking the Factory Manager of such a plant on an inspection trip through a plant some years ago. He had been with his company for many years and admitted that he had treated his Inspection Department as a necessary evil too long. He then became vitally interested in bolstering and supporting his newly revamped Inspection Department. Management did not support his policies.

Thirdly, let us consider plants where Management has an Inspection Department and supports it. This is most desirable from the standpoint of the functioning of the Inspection Department. Here you will find the average Inspection Department doing a good job. Naturally, if an Inspection Department does not do a good job, then of course by virtue of Management's desire to have a good Inspection Department you can easily expect that Management involved will make the necessary adjustments to get the desired results.

Generally speaking, the desire of Management as pertains to the Inspection Department will be to have an Inspection Department equal to its desire.

It is the plant where Management recognizes the importance of an Inspection Department that the author should like to consider as the basis of developing this subject.

At the author's plant the philosophy of wanting to have a good Inspection Department is followed and to do this the Inspection Department must be supported to the utmost. We find that it is one of the most economically sound phases of our business.

Since this subject is in reality divided into two phases, i. e., Duties of the Inspection Department and then Functions of the Inspection Department, it is logical to take the subject of duties first.

The author of this paper has considerable experience with Inspection Departments and is well qualified to write on its duties and functions. Safeguarding the quality of the product is the fundamental duty and shipping of quality castings to the customer is the function of the Inspection Department.

Inspection Duties

Duties of the Inspection Department.—Fundamentally, the duties of the Inspection Department are to safeguard the quality of the product being manufactured. Naturally, foundrymen are interested in Inspection as it pertains to castings.

It is not the intent of the author to tell you how you should inspect your castings or how you should set up your Inspection Department. Inspection is inspection whether in a foundry or in any other industry, and the author feels that the fundamental stated, i. e., safeguarding the quality of the product, applies to the

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Inspection Department in any industry.

You will note that the author said, "Safeguarding the quality of the product." Management would be entirely wrong if it were to assume that the Inspection Department were fully responsible for the quality of its product. The author is inclined to believe that this is an important item that sometimes is overlooked.

Safeguarding quality is entirely different from building quality into a product. In various papers the author has presented in the past, he has emphasized the fact that you build quality into a product; you do not inspect quality into a product. Quality in a product has to be considered from the inception of the product. You must think Quality while designing. You must think Quality when making tooling, which in the case of the foundry represents largely your patterns, molding equipment, etc.

In actual manufacture, which in the foundryman's case is the making of castings, you must instill in your entire organization the desire to have the workmanship reflected in the quality you desire. Your Inspection Department through co-operation can help, but in the final word their duty is still to inspect and to do their work so that your quality is safeguarded as it is shipped to your customers to the degree you desire.

Inspection Functions

Functions of the Inspection Department.—The author analyzes the functions of the Inspection Department as the means to shipping of Quality castings. Naturally, the Inspection Department has to function in the interest of Management to fulfill Management's desire to furnish a product satisfactory to its customers. Here again it is not the author's intent to go into specific details of the minute methods of Inspection, but rather into some of the fundamentals that an Inspection Department may use as a guide in doing its work.

First of all, and most important, is to know the policy of Management and to work to fulfill that policy as near as possible.

Secondly, to do this you must have the proper Inspection Department organization. This organization must be made up of well qualified employees and placed in the most logical assignments to take full advantage of their knowledge, skill and tact.

Thirdly, reference has just been made to proper placement of the inspectors based on their knowledge, skill and tact. Naturally, in such placement most consideration must be given to the man to supervise the Inspection Department. This individual must be well versed in all details of the product. The man so selected must also be given maximum authority for his responsibility of seeing that the best possible judgment is used in making decisions.

Final Decisions

This authority should give him power to make final decisions, but he must be big enough to consult and use advice of other department heads. There are many parts that are not strictly in conformity with specifications which will still function satisfactorily in the finished product.

The authority to make this kind of decision must be given the individual in charge of the Inspection Department by Management and supported by them. This is necessary to develop good judgment which, after all, is vital to the proper functioning of an Inspection Department. Inspectors without good judgment can make unnecessary and unreasonable demands that prove to be economically unsound.

One place where good judgment is of utmost value in foundry inspection is in the case where costly castings might have a blow hole in a surface which is to be machined. It is relatively easy for the foundry inspector to determine if the finish cut will clean up at the point of the blow hole. If it will, there is no reason for the casting to be rejected by either the foundry or the ultimate user.

Liaison Work

In the case of borderline cases such a program may require considerable liaison work between the foundry and customer. The importance of this liaison work cannot be over-emphasized. One of the most important details of this is to be assured that decision by the foundry is in line with the desires of the proper authority of the customer. This authority varies in many companies. Usually this is better accomplished by the two Inspection Departments working together.

Certainly there would be no justification in rejecting castings of the type that will clean up satisfactorily. By the same token, however, it is economically unsound to put too much preliminary investigational work on a small inexpensive casting where the effort expended exceeds the value of the casting.

Fourthly, to function best the Inspection Department must work directly under top management. It cannot and must not be subordinated to any lesser division of the organization. This does not mean that it cannot be co-operative with the other departments, for in this respect it must be most co-operative. It must not, however, subordinate itself to any other department in the entire organization.

Fifthly, often times Inspection Departments err in adhering to the strict philosophy that their sole function is to be assured that the product is in full accordance with all established specifications. Basically, if they function in this manner, they are right. However, the functioning of the Inspection Department can become more valuable and not only accomplish their work, but do it better, if they will go just a little farther.

Deviation Analysis

This is the ability to analyze the cause for either the parts themselves or the finished commodity not being in accordance with the specifications. If the Inspection Department can find the cause, it aids the entire program materially if it is able to pass it on to interested members of other interested departments the reason for such deviation.

True it is that this information may not be used by the other departments, and members of the other departments may even take the attitude that the Inspection Department is trying to usurp their powers and dictate how to do their job. The author does not advocate this going a little farther for that purpose, but does feel that this added information can and should be used in a co-operative manner by those who are directly involved.

Sixthly, this same philosophy followed as to the reason for rejection of a part, as to why the part is wrong, or why, being wrong it will not function properly, can be the cause of widespread respect throughout the plant in gaining respect for the judgment of the inspectors.

Too many times the inspectors' remarks are simply to the effect that the part is not according to specifications. However, the inspector who is able to show why the part is not according to specifications in addition to knowing it is not, gains much respect for himself and his department.

Seventhly, installation of pride in the company and product in the minds of inspectors materially aids them in their work. It should be a source of pride to any Inspection Department that the product sold meets the approval of the public or customer purchasing the commodity.

Pride in Quality

Thinking in terms of castings, how much pride does your foundry's Inspection Department have in the fact that a large number of the castings you ship are ultimately rejected by customers because of defects the inspectors in your foundry should have discovered and relegated to the cupola for re-melting instead of having shipped to the customer? By the same token to what degree are they interested in the ultimate user's welfare if they realize that certain casting defects, such as cracks, might result in the death of a person through failure of the casting in the field?

The foregoing items which are so vital to the functioning of an Inspection Department are only a few of the many that could be cited, but should be sufficient to bring to your attention enough to aid you or the men in your Inspection work to do a better job.

The author is vitally interested in doing everything possible to improve Inspection as it pertains to foundries. He considers it one of the outstanding privileges he has enjoyed to have served as the first Chairman of the A. F. A. Inspection Committee. This committee under the subsequent able leadership of Messrs. Warner and Leverenz has worked hard to bring further to foundries' attention the importance of good inspection in foundries.

The work of the Armed Forces Inspectors stationed in foundries and industries using castings during the war should have stressed the importance of good inspection in the foundry industry. Too many times foundries think of their production in terms of yield as it leaves the foundry, and are prone to be critical of inspectors who reduce this yield by rejecting defective castings.

Consistent Quality

If these inspectors use good judgment, they should be encouraged rather than adversely criticized, for after all, the yield you are really interested in is not just your foundry indication, but that indication plus the tonnage rejected by your customers. Castings rejected by customers, especially after partial or full machining, represent a loss not only to the foundry, but to the customer, for under standard practice his labor in machine work is lost completely.

In the author's opinion it is the foundry which constantly ships good castings that is the one which will survive coming competition. Such a foundry will find a good Inspection Department a definite factor in post-war survival. We are now in a position where most foundries still do not have sufficient capacity for added business.

Other Fabricating Methods

However, you as foundrymen should not lose sight of the fact that other industries are figuring ways and means of taking work away from the foundry and any item now made in a foundry which later has to be made in a fabricated part because of economy or quality, lessens the prestige of the foundry industry in general. The author favors castings in design where they can be used and does not like to see a casting replaced by a part fabricated in a different way just because you have failed to make good castings.

The author brings this paper to a conclusion by asking you of Foundry Management, "Are you accepting your responsibility to your customers by supporting a good Inspection Department?"

DISCUSSION

Chairman: H. W. WARNER, Allis-Chalmers Mfg. Co., Milwaukee.

Co-Chairman: E. G. LEVERENZ, American Steel Foundries, East Chicago, Ind.

MEMBER: Would you care to comment on the relation of inspection to salvage?

MR. JOHNSON: I do not know just what your thoughts are since in my paper I did not touch on salvage. In discussing the general manufacturing plan rather than the foundry alone, the salvage should come directly under the Inspection Department. In my paper I referred to the liaison work necessary between the foundry and the customer or, as far as that is concerned, any manufacturing organization and the customer. I also referred to the fact that the authority varied in many cases.

I am surprised that the question has not been asked whether inspection should come under engineering or some other department. If you have a well organized inspection department, with the salvage directly under inspection working with the ultimate customer, your inspection department should be the liaison organization because it has a good over-all picture. Through that liaison work the salvage is tied in with inspection and you are not apt to get into trouble.

There are a lot of schools of thought which pertain to salvaging, especially as applied to castings. I know that in inspection work one of the most critical things pertaining to the inspection of gray iron castings was the question of welding. There are cases where welding is feasible and there are cases where welding is not feasible, in my opinion.

I do not see how the salvaging department could salvage a casting without some liaison work with the customer with whom they are dealing. In my opinion, if they have a good inspection department they will be well satisfied with the ultimate results.

MEMBER: How far would you go with salvage? Would you make minor salvage repairs on 50 per cent of the castings, on 20 per cent, or on all of them?

MR. JOHNSON: I would not want to talk in terms of percentages. There again it centers on liaison work between the two companies and personally, I feel there are many castings that are salvaged that should be usable without any salvage. I think sometimes people get too picayunish.

For example, in the case of a surface condition, where it is merely appearance and nothing else, I have seen foundries endeavor to weld those and grind them off. In my opinion it was a poorer casting than if they had not even touched it. I would rather see those minor defects left in because then you are better able to judge the casting. When it has been welded you do not have as good idea as you do if it had not been welded.

I cannot answer your question on a percentage basis. I still think it is just a question of good judgment and sound understanding between the two inspection departments.

GRAY IRON WEAR RESISTANCE

Comparison of gray irons showing good to bad metal-to-metal wear resistance has proved that variations in microstructure are associated with wear performance. Superior wear and galling resistance of gray iron is achieved with a complete and fine pearlitic matrix with A.F.A. Type A graphite. Coarse pearlite, free ferrite or cementite in the matrix and A.F.A. Types D and E graphite decrease the wear and gall resistance.

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INVESTIGATIONS by Paul Lane[1], E. K. Smith[2], Frank Dost[3], and others have shown that certain microstructures of gray iron are particularly vulnerable in metal-to-metal wear applications. Many of these same investigators also show that certain other gray iron microstructures offer the maximum resistance to wear and galling.

Those structures recognized as having the best wear properties are illustrated in Fig. 1. These may be described as having a fully pearlitic matrix with the A.F.A. Type A graphite. Those gray irons that show unsatisfactory resistance to galling and wear may show varying degrees of one or the other of the structures illustrated in Figs. 2 and 3.

Scope. In this discussion, comments are confined to those machine parts which occasionally fail because of differences in the cast iron structure and not because of inadequate lubrication or high bearing pressures, etc. In piston rings, engine and pump cylinders, machine tool ways, etc., it has been found that gray iron parts with random A.F.A. Type A graphite and a fully pearlitic metal matrix perform more satisfactorily than those irons with the structures illustrated in Figs. 2 and 3.

Furthermore, gray irons containing varying amounts of the poorer wearing structures of Figs. 2 and 3 have poorer wear resistance in proportion to the amount of these structures that is exposed to wear. However, a small amount of the vulnerable structure becomes the weak link in the wear resistance of that gray iron. The important feature, therefore, of achieving the optimum wear and galling resistance of gray iron is to provide the desirable structure throughout the metal.

Gray Iron Structures Vulnerable to Wear. From the reports of various investigators, and from the writer's experience with piston rings, cylinder liners, machine tool ways, etc., various gray iron structures may be classified in order of their decreasing wear resistance:

A. Refined and completely pearlitic matrix with A.F.A. Type A graphite (Fig. 1).

B. Coarse pearlitic matrix with A.F.A. Type A graphite.

C1. Pearlite with a small amount of free ferrite matrix with A.F.A. Type A graphite (Fig. 3A).

C2. Pearlite with a small amount of free cementite matrix with A.F.A. Type A graphite (Fig. 3B).

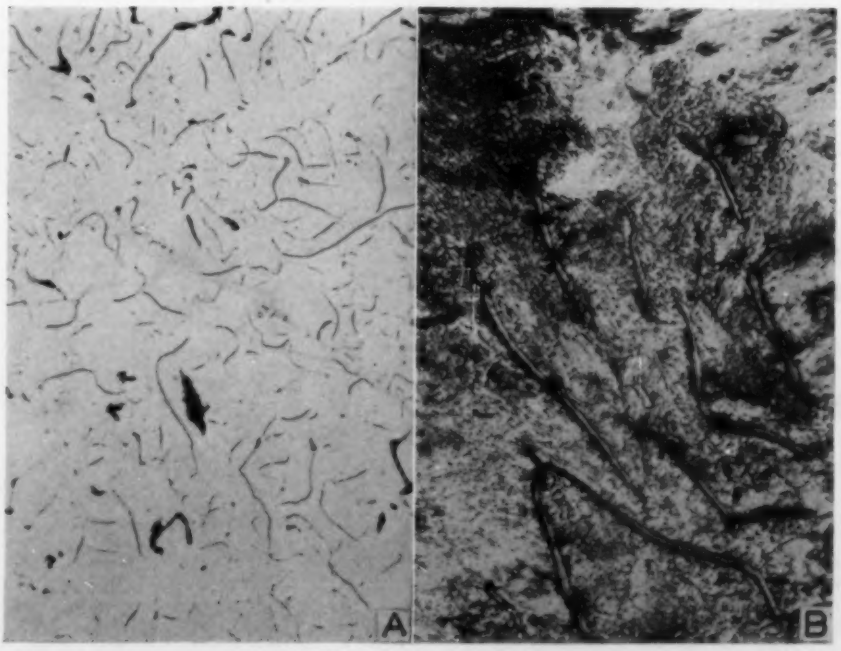
D. Abundant ferrite matrix with A.F.A. Type A graphite.

E. Abundant free cementite matrix with any form of graphite.

F. Secondary ferrite matrix associated with dendritic graphite, A.F.A. Type D or E graphite.

However, this discussion is not so

Fig. 1 (A)—Desirable graphite structure for wear resistance (A.F.A. Type A). Unetched. 100X. (B)—Fine and completely pearlitic matrix. Nital etch. 500X.



much on the degree of wear resistance of different gray irons as on preventing the undesirable structures for optimum wear properties in gray iron machine parts.

To achieve a refined pearlitic matrix free of primary ferrite (condition *A*) in gray iron, a minimum of 190-200 Brinell hardness should be aimed for with the use of those alloys which refine the pearlite without tending toward producing cementite. It is obvious that if a coarse pearlite (condition *B*) prevails, the area of ferrite particles exposed to wear can become sufficiently large so that galling or seizure can take place. Therefore, a careful balance of the carbon, silicon and alloys to the casting sections and properties desired will readily maintain the refined and fully pearlitic matrix, provided the graphite structure is proper.

Fig. 2 (A)—Dendritic pattern of graphite (A.F.A. Type D or E). Unetched. 100X. (B)—Gray iron matrix associated with dendritic graphite. Small white areas are ferrite in small graphite flakes. Nital etch. 500X.

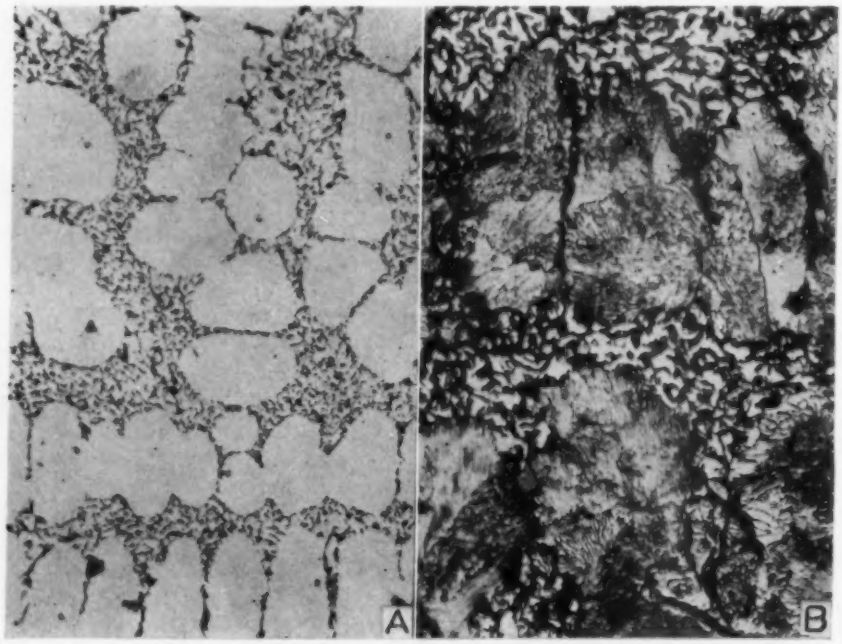
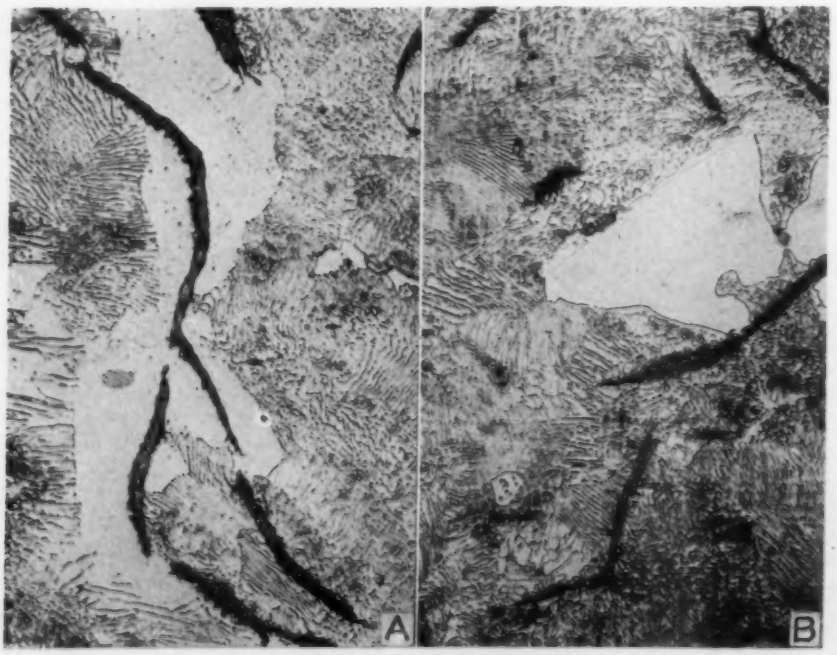


Fig. 3 (A)—Ferrite (etched white) and pearlite matrix of soft gray iron. Nital etch. 500X. (B)—Cementite (white) and pearlite matrix of hard gray iron. Nital etch. 500X.



In the foregoing discussion on matrix structure control by means of manipulating the carbon, silicon and alloys, it is pointed out that graphite structure control was not mentioned. However, gray irons with A.F.A. Type D or E graphite structure (Fig. 2A) have strong tendencies to produce matrix structures of Fig. 2B. It is to be noted that the white spots in Fig. 2B are ferrite almost surrounded by small graphite flakes. It is this condition of small metal particles associated with small graphite flakes that is most vulnerable to wear. A study of the influences or factors causing this structure[4]:

1. Increasing chill capacity as affected by the composition.
2. Rapid freezing.
   a. Thin sections.
   b. Low pouring temperatures.
3. Decreasing total carbon and/or increasing steel in melting mixture.
4. Increasing melting temperatures above 2850° F. (superheating).
5. Degree of oxidation during melting. In the order of increasing oxidation effects, melting units can be arranged as follows:
   a. Cupola.
   b. Indirect arc furnace.
   c. Direct arc furnace.
   d. High frequency electric furnace.
   e. Crucible and air furnace.

Factors 1 and 2*a* often are beyond the foundry control, since the composition, design and mechanical properties cannot be changed because of specifications and demand on the castings. Factors 2*b*, 3, 4 and 5, and to some extent factors 1 and 2*a*, can be controlled by suitable foundry practice.

Control of Desirable Gray Iron Structures. Fortunately, all of the factors influencing the production of Types D and E graphite can be overcome by careful foundry control, and their effects on graphite structure can be efficiently prevented by adequate inoculation of the metal just before pouring the castings.

Inoculation is a gray iron foundry

term meaning the addition of silicon, or its combination with other metals; iron, nickel, calcium, chromium, etc. From the author's work on inoculation, these late additions have two effects:

1. They provide final deoxidation to melt (very common metallurgical practice, indeed).

2. They provide seeding of the melt with relatively concentrated silicon, which initiates early graphitization during freezing.

The dendritic pattern of Types D and E graphite indicates late graphitization or supercooling[5, 6].

Inoculant Control

It is obvious that as the potency to produce Type D and E graphite increases, the need for more inoculation or more powerful inoculants increases. The effectiveness of these inoculants in controlling the desired graphite microstructure (Type A) depends upon the type and amount of the inoculant used. The capacity of the iron to become supercooled and to throw Type D or E graphite is dependent upon the influence of the Factors 1 to 5 previously mentioned. The greater this capacity to throw Type D or E graphite, the more inoculant required, for example:

Cupola Iron with TC, (per cent)	*Required Ladle Silicon Addition, (per cent of metal)*
3.20—3.35	0.20
3.00—3.20	0.30
2.75—3.00	0.40
2.50—2.75	0.50

An iron with 3.25 per cent TC poured into ⅛-in. sections or superheated to 2950° F. may require more than the indicated amount because factors 2*a* and 4 may be more pronounced. One limiting factor must be borne in mind, i.e., the casting skin effect of a few thousandths of an inch will show the Type D or E graphite and cannot be prevented. Since this skin is machined off before placing into metal-to-metal wear service, all the foundryman need do is to add inoculant in sufficient amount and the proper power to hold the undesirable graphite structure to the desired thinness.

A few details about inoculation might be mentioned which the foundryman should bear in mind:

A. Temperature of metal inoculated should not be below 2750° F.

B. The inoculant should be fused into the metal without burning. Powdered silicon, for example, burns readily with a blue flame and is, therefore, ineffective.

C. Adjustment of the composition is necessary to accommodate the ladle addition of silicon and its combined alloys.

With a little experience the foundryman soon learns how much inoculant to add to control the structure of his particular irons in the sections involved.

In the writer's experience with gray irons poured into piston rings, cylinder liners and blocks, machine tool castings, etc., no case has been found where the graphite structure could not be controlled, except in the skin of the castings. This skin effect in extremely light castings has been consistently held to .020 in. with adequate inoculation.

Conclusion

For superior metal-to-metal wear, it has been shown that gray iron should have a fine and fully pearlitic matrix with A.F.A. Type A graphite. This structure can be adequately controlled with the proper balance of total carbon, silicon and/or alloys, and with adequate ladle inoculation for the metal and castings involved.

Foundries should find no production difficulties, therefore, in meeting a gray iron structure specification for superior wear performance.

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DISCUSSION

Chairman: R. G. MCELWEE, Vanadium Corp. of America, Detroit.

Co-Chairman: W. E. MAHIN, Armour Research Foundation, Illinois Institute of Technology, Chicago.

T. E. EAGAN[1] *(written discussion):* The author has emphasized the fact that type A graphite with a fine pearlitic matrix gives the best wear resistance. He has also reviewed the methods of control necessary to get desirable gray iron structures. The only thing in the way of comment on this paper that I could make is to confirm his work.

After a good many years of study of the wear resistance of Diesel engine cylinder liners in which measurements of wear were taken under actual operating conditions instead of laboratory wear tests, we found that, in general, a type A graphite structure with a fine pearlitic matrix gave excellent wear results. There was only one exception to this and that is the use of copper as an alloy.

For some time we used one per cent copper as the alloy for the refinement of the pearlite matrix, and we experienced a considerable amount of wear, some of it being as high as 0.040 in. per 1,000 hr. of operation. Investigation showed that every one of these liners had the one per cent copper, that the microstructure was type A graphite with fine pearlite.

The exception to this was that if a minute amount of chromium or molybdenum (0.15 per cent or less) was present the wear was excellent. The evidence that we obtained through these actual tests was so conclusive that we are willing to state that the addition of copper without other alloys such as molybdenum and chromium will give poor wearing liner material.

We confirmed the fact that the Brinell hardness of the liners must be above 190 to 200, and our specification called for a minimum of 200. The actual range is between 200 and 250. However, I am wondering if we are not wrong in stating that the best wear resistance of any gray iron is one that is free of primary cementite.

Would it not be better to say that the best wear resistance of a machinable gray iron is one that has type A graphite distribution with fine pearlite structure? Because of the fact that most liners are of such a design that they must be machined we seldom, if ever, use a chilled iron.

Therefore, we have very little evidence of the wear resistance of an iron that contains rather large amounts of primary cementite, and until we have evidence to the contrary, we may be misled by making too broad a statement.

The statement that Mr. Sefing makes that he is considering metal to metal wear resistance should be emphasized. Other types of wear require different types of materials to give proper results.

For instance, in abrasive wear a much better type of material would be one that contains a considerable amount of cementite. In fact, any type of abrasive wear instead of friction wear requires a harder matrix than does the straight friction wear.

There are two factors which control wear resistance. One is the metallurgy of the iron used, and the second is the design of the parts. Our investigation has shown that there is a limit to which metallurgists may go in improving the

wear resistance of cast iron.

In our own case, this limit seems to be about 0.0005 in. per 1,000 hr. of operation. However, by proper design such as oil cooling of pistons, proper ring design and better control over lubrication, this figure can be vastly improved so that in the end we experience phenomenal wear resistance. This should be true because if the two metals never touch each other, which is theoretically possible by the use of proper lubrication, we should experience no wear whatsoever.

H. W. LOWNIE[2]: I am disappointed that Mr. Sefing did not present more data on the correlation between microstructure and wear resistance. The use of inoculation as a method for controlling the structure has been discussed but there seems to be a missing link between the resultant structures and their wear resistance.

MR. SEFING: I think that if I were to have presented sheaves of data on wear resistance vs. structure, no one would have believed the data anyway. All that we can do is take service castings that have performed satisfactorily and those that have performed unsatisfactorily and inspect them and see what the score is.

MEMBER: We have made a study of such service castings as engine pumps, engine cylinders, and piston rings. We have found consistently that those that performed well had the A.F.A. type A graphite, and those that performed poorly had the dendritic graphite.

We had 42 piston rings that were taken out of engines after 8 hr. because they scored the cylinder and there were 82 piston rings that performed satisfactorily in the engines up to 250 service hours. Our laboratory shows that you can go to almost double bearing pressure dry test with a good graphite structure than the dendritic type.

MR. LOWNIE: This experience ties in pretty well with some wear resistant applications that I have worked with.

Mr. Sefing has made a point about the difference between the use of inoculation for chill control and its use for structural control. My thought is that the mechanism must be the same in both cases.

MR. SEFING: When you are pouring a soft iron, the gray portion of the casting will usually show a good A.F.A. type graphite. The edges of all castings may show a few thousandths of an inch dendritic pattern. The wearing surface is machined off. If you are machining off ¼ in. of metal you have to hold the dendritic graphite beyond that ¼ in. If you are working with piston rings and machining off 0.020 in., you must move the dendritic pattern to less than 0.020 in. and that is a tough job.

MR. LOWNIE: In that case you are controlling chill and controlling structure at the same time and by the same mechanism.

MR. SEFING: I am not ready to accept that they are exactly the same thing.

MR. LOWNIE: You made the statement that the effectiveness of inoculation can be determined by determining the amount of silicon picked up in the ladle. I do not think that this is correct.

MR. SEFING: If you are making a 0.30 per cent silicon addition cupola mix, we will say a 40 per cent steel mix, the 0.30 silicon may be inadequate when you use fine material (say 20 mesh) so that you burn up about half of it. Then you have to add 0.6 per cent silicon in order to get 0.30 per cent silicon into the metal.

H. C. AMTSBERG[3]: I am inclined to agree with Mr. Lownie on this structural control and disagree with the point of having a certain specified amount of inoculant added to an iron of particular chemical composition.

In our work at Westinghouse we have found that although we may have the same chemical composition in an iron, we may have to add different amounts of inoculant in order to control the chill. We find in controlling the chill we do modify our structure to a certain extent.

MR. SEFING: The amount that I suggested is for a 3 per cent carbon iron and it is probably more than a production foundry would need. It is true that the amount of silicon to be added will be varied by the proportion of the scrap and with variations in oxidation in melting.

CHAIRMAN MCELWEE: It is difficult to make accelerated tests to compare with field tests. However, I know of one occasion when we made a study that was an accelerated test. It is a worm and worm wheel. The worm was SAE 1045 steel and the worm wheel was cast iron. You have extremely high unit pressures, and in this particular case the company that was making these worms attempted to make a study. We cooperated and found confirmation of what Mr. Sefing said. If there was any tendency toward whorls or dendritic graphite formation in a very short time the ferrite particle would pull out, would weld itself to the worm with resultant cutting tool effect. The test wheel failed shortly.

[1]The Cooper-Bessemer Corp., Grove City, Pa.

[2]Battelle Memorial Institute, Columbus, Ohio.

[3]Westinghouse Electric Corp., E. Pittsburgh, Pa.

FOUNDRY REFRACTORIES...

THEIR PROPERTIES AND APPLICATION

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OF THE COMMERCIAL REFRACtory materials, fireclay refractories are used in largest quantities in the foundry industry. The properties of fireclay brick are frequently considered toward the end of securing those products which may be used to best advantage in the cupola and accessories, malleable furnaces, annealing furnaces, and other refractory constructions. Individual mention will be made of the various properties of fireclay brick, both physical and chemical.

Pyrometric Cones

Refractoriness. Obviously, a material which is supplied for heat resistance should have a high fusion point. More accurately, the ability of refractories to withstand temperatures causing softening is referred to as the pyrometric cone equivalent (P.C.E.). Many factors influence the softening point of refractory materials. Because of these, the comparison of an unknown refractory with a standard and known refractory composition has been found to permit most satisfactory evaluation of the temperature-resisting qualities of the unknown.

Standard pyrometric cones are available, the bending temperatures of which have been carefully determined. In testing a given refractory product, a small sample is finely ground and molded into a similar cone-shaped unit. This is then installed on a cone pat along with standard cones and heated according to a predetermined, definite schedule. At the point the sample cone is observed to bend and touch the refractory pat, the condition of the standard cones is recorded (Fig. 1). This permits satisfactory evaluation of the refractory qualities of the sample tested.

In discussing refractoriness of fireclay brick, it should be emphasized that this refers to the pyrometric cone equivalent determined as has been described, and not to the temperature which the particular brick will withstand (Fig. 2). Fireclay brick may have a P.C.E. of Cone 32½, corresponding to a temperature of 3131 F. However, under a load of 25 psi these brick will tend to show measurable deformation at a temperature of about 2250 F.

Furthermore, at temperatures in the neighborhood of 2250 F, this refractory may show some dimensional changes, even though little or no load is applied. These properties will be dealt with further.

Chemical Analysis. The chemical composition of a refractory is extremely important but requires care in properly understanding its relation to a given set of service conditions. It is only necessary to consider the satisfactory application of highly siliceous brick where a refractory more basic in nature seems indicated in order to understand that the chemical analysis of a fireclay brick needs considerable interpretation to make it a useful guide. In fireclay brick, the ratio of silica to alumina is important in gaining the preliminary estimation of a given refractory. The iron oxide content is not especially important as long as it is held within reasonable limits.

An outline of the important properties of the various refractories used in the foundry—indicating their practical meaning to the foundry operator and describing particular applications.

Physical Properties Important

Lime, magnesia, and alkalies tend to lower the refractoriness of fireclay brick, so the percentages of these are important. However, the effect of given percentages of lime, magnesia, or alkalies may vary through considerable limits, depending upon the brick in question. It might be said that chemical analysis provides a worth-while preliminary estimation of the refractory, but the physical properties generally are far more important.

Porosity-Permeability. In cupola work, it has long been recognized that low porosity is desirable in refractories, particularly when fluid and corrosive slags must be resisted. High heat duty fireclay brick may show porosities roughly within the range 10-30 per cent. Accordingly, the porosity of a cupola block may be an important consideration in the length of service life secured in a melting zone. Here again, other considerations must be taken into ac-

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count, since porosity in itself should not be the sole criterion used to estimate the performance which may be expected. Chemical composition may readily influence the service life.

For example, it is known that highly siliceous blocks may in some instances resist the fluxing action of the cupola slag better than less siliceous brick, even though the porosity of the former is considerably higher. This is further borne out by the splendid performance of silica brick in some cupola operations, where high steel charges are used and exceptionally severe conditions prevail. The silica brick have substantially higher porosity than the generally used fireclay cupola blocks.

Permeability Important

In more recent years, the importance of permeability in a refractory has been recognized. This property differs from porosity in that it provides a measure of the type or kind of pores, in addition to quantity. Where the pores in a refractory are connected, it can reasonably be expected that slag attack and corrosion may proceed much more rapidly than in a refractory of equal porosity, but one in which the pores are independent or only infrequently connected.

In the permeability measurement, air under pressure is passed through the refractory sample with definitely controlled and standard conditions. The result is reported as the volume of air passing through a unit thickness of the refractory in a given period of time.

The full import of permeability measurements has not been developed up to the present time. However, it has been interesting to note that superduty fireclay brick may have from three to ten times the permeability of dense high heat duty fireclay brick. Since superduty fireclay brick has not been widely used in cupola operations, although frequent trials have been made, their higher permeability may possibly offer some explanation. In some nonferrous melting furnaces, where the brick are in direct contact with the molten metal, certain low permeability-high heat duty fireclay brick have shown definite superiority over brick of similar chemical composition, but of higher permeability.

Strength-Abrasion Resistance. The mechanical strength of refractories has frequently been given undue attention. There are, of course, applications in which maximum mechanical strength is required, but for a great many refractory constructions the strength needed in a brick is only that which will insure handling up to the job without excessive cornering and breakage. The natural appeal of a mechanically strong brick is obvious, but often in obtaining high mechanical strength of product other much more important properties are sacrificed. This can also be a consideration in the handling of refractory brick around the plant. Where handling is rough, strong brick are required in order to arrive on the job in good condition.

Frequently, improvements in handling procedure may be made without material increase in cost. These handling improvements may then permit taking advantage of the other important properties in the brick that do not necessarily go along with the highest mechanical strength.



Fig. 1—(above) Appearance of test cones after heating.

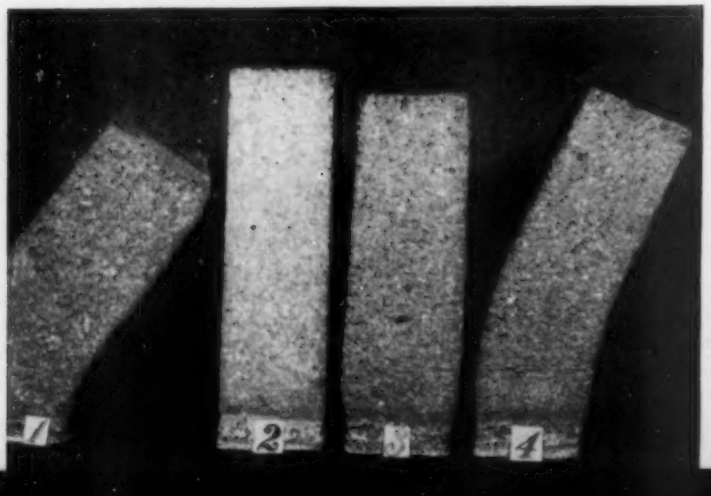
Fig. 2—(left below) Fireclay brick after heating to a temperature of approximately 3150 F.

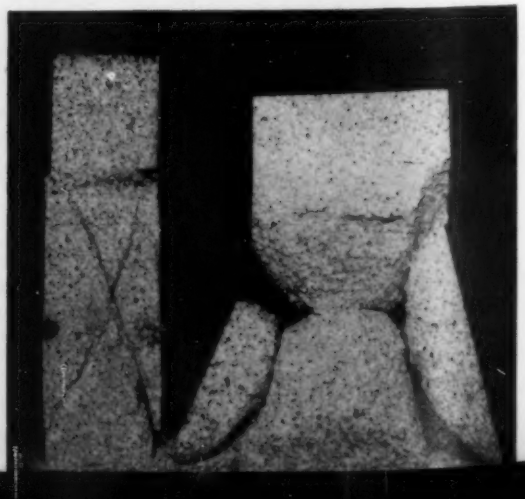
Fig. 3—(right below) Appearance of fireclay brick after ASTM crushing test.

Strength Determinations

Methods of determining strength of refractories deal principally with the determination of cold crushing strength, modulus of rupture, and abrasion resistance. Cold crushing strength, according to the ASTM procedure, is determined on a 9-in. straight brick stood on end in the conventional testing machine. A load is applied until the brick is crushed (Fig. 3). The result is then reported in the load in pounds per square inch required to crush the brick.

The modulus of rupture is determined on the brick laid on flat in span across two knife-edge supports. Load is applied at the center of the span by the testing machine, and the result is reported in pounds per square inch according to the usual modulus of rupture formula. The determination of abrasion resistance is not as precise a test as either the cold crushing strength or modulus of rupture test. A commonly used abrasion test consists of a modification of the "rattler test" as used on paving brick. The samples are placed in a steel drum, along with a definite weight of iron balls. The drum is





then revolved a definite number of times, after which the brick are removed and weighed. From the weights before and after this test the loss is determined (Fig. 4).

In the malleable furnace sidewalls and bottom, the strength of the refractory may be important. In the walls, the cutting action of misdirected flames may frequently be substantial. In the bottom, particularly in operations where cold charging is used, the brick may be subjected to considerable mechanical abuse. In the upper section of the cupola lining, depending upon the charging methods, abrasion may be an important factor in consumption of the refractory lining. In walls of many heating and annealing furnaces, negligible mechanical abuse is involved and, in such constructions, the best combination of over-all properties may be secured if exceptionally high mechanical strength is not demanded.

Right—top to bottom:
Fig. 4—Typical appearance of blast furnace brick after 600 revolutions in paving brick rattler machine.

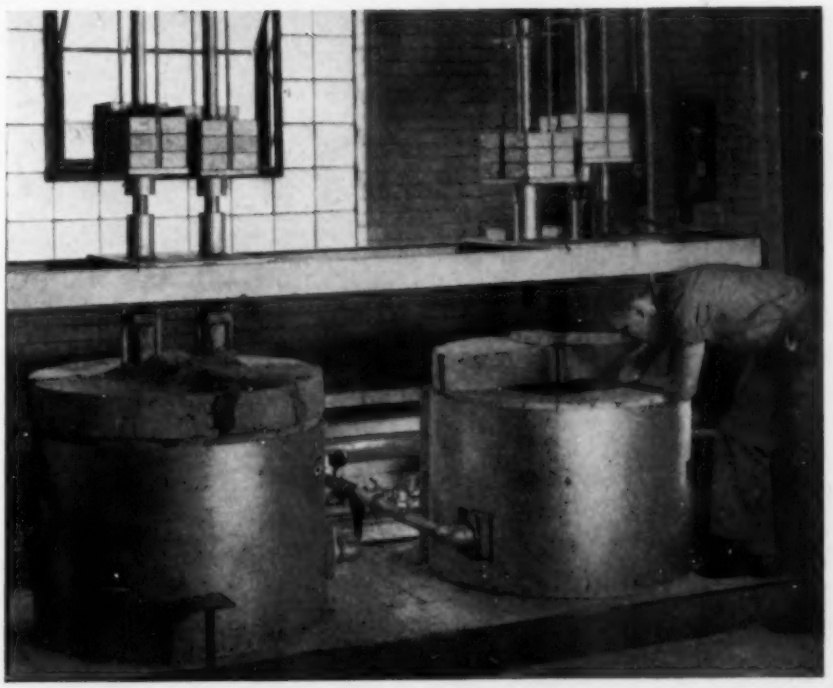
Fig. 5—Furnaces for testing load-bearing properties of refractory brick.

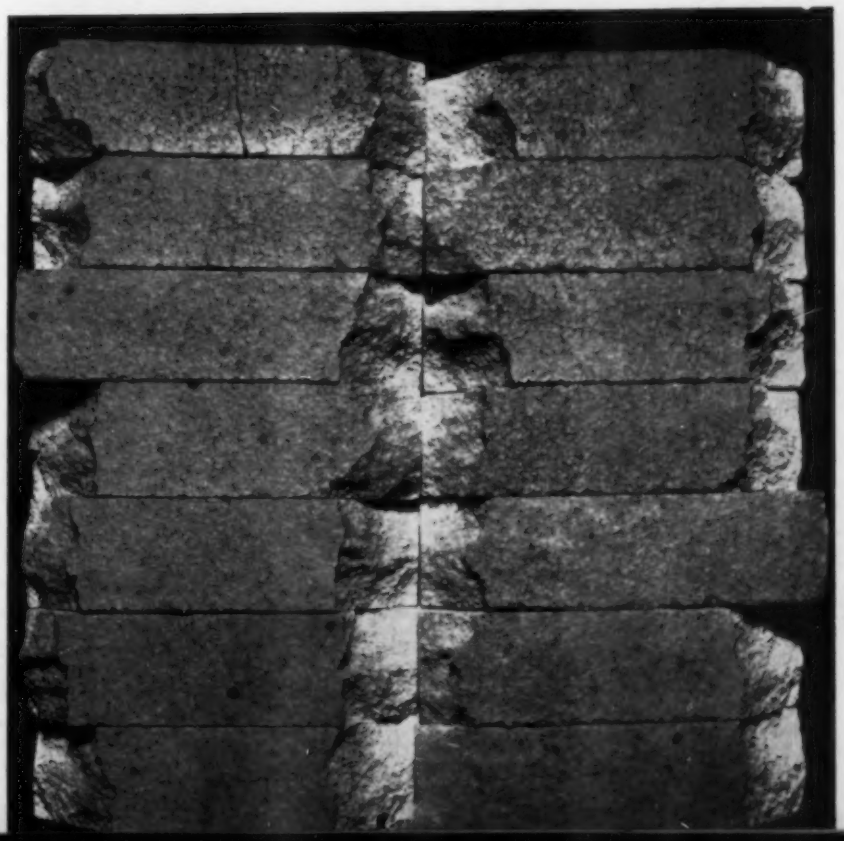
Fig. 6—Typical appearance of high heat duty brick after panel spalling test. Preheat temperature, 1600 C (2912 F).

Volume Stability at High Temperatures. The importance of this property is rather obvious, since refractories of the proper quality after being installed are expected to be reasonably free from permanent dimensional changes. In the case of excessive shrinkage in service, wide joints may develop. With excessive shrinkage, the affected portion of the brick may be so structurally changed that it separates from the body of the wall.

High shrinkage may so alter the form of construction that failure results. With excessive permanent expansion, the furnace structure may be disrupted, the brick may be pinched off at the hot ends, or other similar undesirable conditions develop.







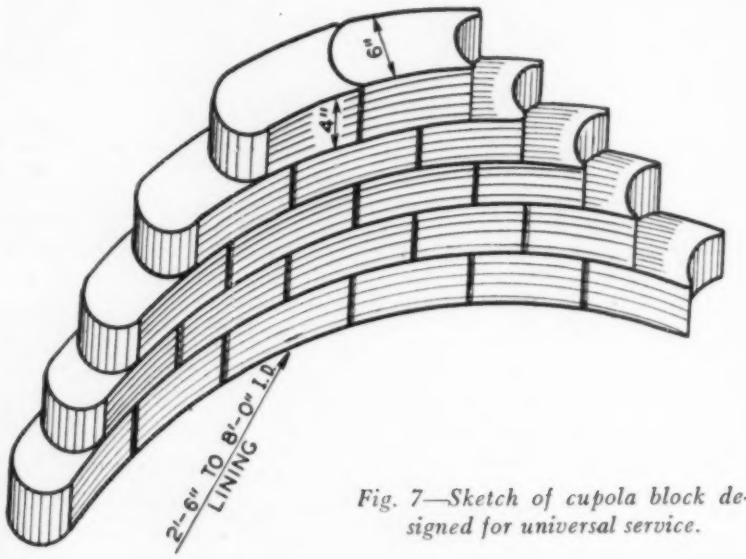


Fig. 7—Sketch of cupola block designed for universal service.

The test to determine the volume stability of refractories consists of heating samples of the full-sized brick according to a definite schedule of heat treatment. With high heat duty fireclay brick, a temperature of 2550 F is used, and this temperature is held for 5 hr. With the superduty fireclay brick, the reheat temperature is 2910 F, also held for 5 hr. After the furnace has been allowed to cool, the brick samples are measured to determine dimensional changes which have taken place as a result of the heating.

Bung Brick

Of the commonly used foundry refractories, bung brick are one class which should be volume stable at high temperatures. This is also true of superduty and high-alumina brick which are used in malleable furnace bottoms. In the cupola, it is naturally desirable that volume stability be maintained with the refractories, but with the relatively lower temperatures and the increased importance of other properties the volume stability of cupola refractories is not generally the major consideration.

Strength Under Load at High Temperatures. In foundry practice, there are relatively few constructions wherein the brick are called upon to withstand heavy loads in the high temperature range. However, in some furnaces, there may be piers supporting relatively heavy loads and, should these piers be immediately adjacent to the fireboxes, the load-bearing properties may be important.

The test for load-bearing properties is conducted in a small cylindrical furnace (Fig. 5). The 9-in. brick are stood on end and a load equal to 25 psi of cross section applied to the 4½x2½-in. face. The high heat duty fireclay brick are subjected to a temperature of 2460 F for 1½ hr. and the superduty fireclay brick to a temperature of 2640 F for the same time.

Under the load, the brick tend to subside, depending upon the amount of liquid formed within the brick structure and the lubricating effect of this liquid on the still solid refractory particles.

Load-Bearing Properties

It was mentioned that under a load of 25 psi the high heat duty fireclay brick tend to subside at a temperature of approximately 2250 F. The superduty fireclay brick support the same load to a higher temperature, namely, in the range of about 2330 F, and intermediate heat duty brick will support this load at a considerably lower temperature, approximately 2175 F.

Where conditions of load and temperature must be resisted, it frequently is necessary to go to the high-alumina brick, especially those containing corundum. These brick have load-bearing properties far exceeding the high heat duty fireclay brick, or even the superduty fireclay brick.

Spalling Resistance. Resistance to spalling has been given a great deal of attention by refractory users in all fields. This may be because failures from spalling are so spectacular. Spalling may be caused by a great number of conditions; therefore, only the most general will be considered here.

Temperature Shock

Before the advent of the superduty fireclay brick, it usually had been considered that a coarse grind and relatively light burn were important in high heat duty fireclay brick if the brick had to withstand substantial temperature shock. High porosity was thought desirable to achieve spalling resistance. In the superduty fireclay brick, extremely high density and low porosity were achieved; furthermore, these brick frequently were extremely hard burned.

However, these refractories showed exceptionally high resistance to thermal shock. The results so obtained have not discounted the previous experience with coarse-textured, light-burned fireclay brick, but have demonstrated that spalling resistance is not the relatively simple matter previously considered.

Resistance Test

The laboratory test for thermal spalling resistance consists of installing fourteen 9-in. straights in the panel which forms the door of a test furnace. This panel is given prescribed heat treatment for 24 hr., which in the case of high heat duty fireclay brick is at a temperature of 2910 F, and with superduty fireclay brick, 3000 F. Following the heat treatment the panel is alternately exposed to the cooling of a blast of moist air and to heating at 2550 F. This schedule is continued for a given number of cycles, following which the loss in weight of the panel is determined (Fig. 6).

There are a number of constructions in foundries in which the spalling resistance of the refractory is most important. For example, spalling conditions may be severe in the upper sections of cupolas. Where

charging methods are such that the upper lining is not subjected to excessive abrasion, some advantage may be taken of modifications in texture of the cupola blocks to secure improved spalling resistance.

Another important construction, insofar as spalling resistance of the refractories is concerned, is in malleable furnace bungs. Here the refractory is exposed not only to high temperatures, but to such heat shock as may be involved in starting up the furnace, or in removing and replacing the bung brick. In the hand-fired malleable furnaces, loose-textured, handmade bung brick frequently serve satisfactorily.

Dimensional Tolerances

In previous years when most of the furnaces were so fired, handmade brick were generally used. However, with the introduction of the powdered-coal firing more severe conditions resulted. Accordingly, the present bung brick was developed. Most bung brick are now made by the power-pressed process. They are not only more resistant to thermal shock, but also have desirable close dimensional tolerances made possible by the power-press process.

It frequently is desirable to consider the spalling resistance of refractories in the construction of heating and annealing furnaces. These furnaces may operate at relatively low temperatures, and intermediate or low heat duty brick appear adequate. However, temperature changes over the period during which the furnace is in operation may be sufficient to produce cracking and spalling of the intermediate or low heat duty brick. High heat duty brick, though not required for the temperature involved, nevertheless do provide a margin of quality as far as spalling resistance is concerned.

Cupola Blocks

Many types of spalling are caused by conditions other than temperature shock. For example, insufficient provision for thermal expansion may cause pinching and spalling of refractories. Improper design of arches may be another cause. In the cupola, use of cupola blocks having the taper not fitting the circle to be turned may result in pinching and spalling of the cupola blocks. A means of insuring a good fit may be in the use of a cupola block designed for universal service (Fig. 7). Another advantage of this block would be in the simplification of stocks where a number of different sizes are required. Trials now in progress should demonstrate the utility of this design.

DISCUSSION

Chairman: C. E. BALES, Ironton Fire Brick Co., Ironton, Ohio.

Co-Chairman: C. S. REED, Chicago Retort & Fire Brick Co., Chicago, Ill.

MEMBER: Would the speaker elucidate briefly on the basic principles of selecting the proper cupola block, considering the temperature, chemical erosion, and abrasion conditions?

MR. BRASHARES: We cannot generalize too much on cupola brick from one foundry to another. There are the high production shops with severe service as contrasted with those in which relatively moderate conditions prevail. It has been the belief for some years that a good cupola block should have low porosity.

There may be some difference of opinion regarding the importance of low porosity, but in any event, I believe equally or more important is the property of low permeability. Assuming that the consideration is limited to high heat duty fireclay brick and that the refractoriness of this class is met, then the qualities of low permeability and low porosity are generally believed important.

As to abrasion resistance, this property is frequently over-emphasized. For example, during the past few years a number of small foundries have changed to a less abrasion-resistant brick for the top portion of their cupolas. Where previously these foundries used a dense, extruded, hard burned brick in this section, they have obtained improved results by using a more open textured handmade refractory.

While this handmade brick may not be the type usually preferred in this section, its more open texture apparently permitted higher resistance to spalling. Where spalling was a greater factor than abrasion in the destruction of the refractory, then it is readily understandable that the handmade brick should give the better results. Here again, we should not generalize too extensively from a few examples, even though substantial savings resulted at these particular plants.

The chemical attack on cupola blocks, particularly in the melting zone, is an important consideration. Since high heat duty fireclay brick may show a silica content of 50 up to 75 per cent or even higher, the range in chemical composition is rather wide. Silica is known to be resistant to iron oxide attack, so increasing silica content in a fireclay brick may be important toward diminishing chemical attack. It is of interest to note that in severe operations where the charge contains a high steel content, even silica brick containing 95 per cent SiO_2 have been used successfully.

MEMBER: How should the foundryman evaluate a given product furnished by a brick manufacturer in order to find out why a superduty brick in one case will spall, and in the other case will run to greater erosion than the one that has spalled? Is the examination of a fracture of the brick with regard to grain size of any help in judging whether the right brick was selected, or whether it went somewhat beyond the line of permeability?

MR. BRASHARES: The examination of fracture is very worth while. The importance of grain sizing, however, may be somewhat questionable. Assuming that cupola blocks fit the circle in which they are used and that pinching is not a factor, nor are the blocks broken by impact, then a rough fracture usually denotes thermal spalling. This results from temperature shock caused by rapid heating or cooling.

A smooth appearance generally denotes chemical attack. Where spalling may be followed by substantial chemical attack, some difficulty is experienced in identifying the spalling. In evaluating a cupola block for the section in which the major amount of wear in a given cupola takes place, it is desirable to consider not one property of this block, but several. Those which are generally of greatest importance are porosity, permeability, chemical analysis, and spalling resistance. Undue emphasis on a single property frequently leads to false conclusions.

MEMBER: Would you recommend a maximum specific pressure for acid and basic brick, considering the burning qualities?

MR. BRASHARES: The actual crushing strength, cold, of most refractories is far in excess of any loading they are generally called upon to withstand in service. However, it should be kept in mind that the load a refractory will withstand cold is far different than that which it will bear at higher temperatures. A high heat duty fireclay brick may have a cold crushing strength of 2000 psi. However, under a load of 25 psi, it will tend to subside to some extent at a temperature of about 2250 F.

Increase in load and temperature will cause further increase in the subsidence. Therefore, we should consider that with fireclay brick subsidence under load is gradual. With silica brick, the most acid commercial refractory, subsidence takes place through a very narrow range. For example, a good silica brick under a load of 25 psi will withstand a temperature in the range of 2950 F. Failure of the silica brick under load at high temperatures is by shear rather than gradual subsidence.

Magnesite brick, the most basic commercial refractories, also show this same type of failure under load at high temperatures, although their softening range may be slightly greater than with silica.

QUALITY CONTROL METHODS

Quality "cannot be inspected into castings." Basic principles of directional solidification, in conjunction with metallurgical and foundry processing, must be understood. Close collaboration of designer and foundryman, with a mutual understanding of all requirements, results in a quality product through properly coordinated process controls. Interpretation of acceptable standards of steel casting quality is of utmost importance. Service requirements of the casting should be thoroughly understood, and specifications properly written for the grade or quality needed. Acceptable standards should be clearly defined and mutually agreed upon beforehand.

J. W. Juppenlatz
Chief Metallurgist
Lebanon Steel Foundry
Lebanon, Penna.

QUALITY CONTROL, when applied to castings, means that the character and over-all properties are consistently and uniformly maintained from casting to casting according to predetermined standards. Before quality and its control can be properly understood, the level of quality required must be established as dictated by the intended service application.

Quality levels of castings are constantly changing and, with the passing years, improvement of foundry processes has resulted in the production of castings approaching perfection. Methods of checking the individual casting for the level of quality have been refined, and more exacting methods of inspection have been developed.

The foundry, with its diversified products for various service applications, must have standard levels of quality adjustable to each type of service application. While a higher level of quality than chosen would be satisfactory for purely economic reasons, the highest quality level may not be required. Therefore, a satisfactory quality level for each casting design should be established according to satisfactory service performance so that production castings can be compared to definitely fixed standards of acceptance.

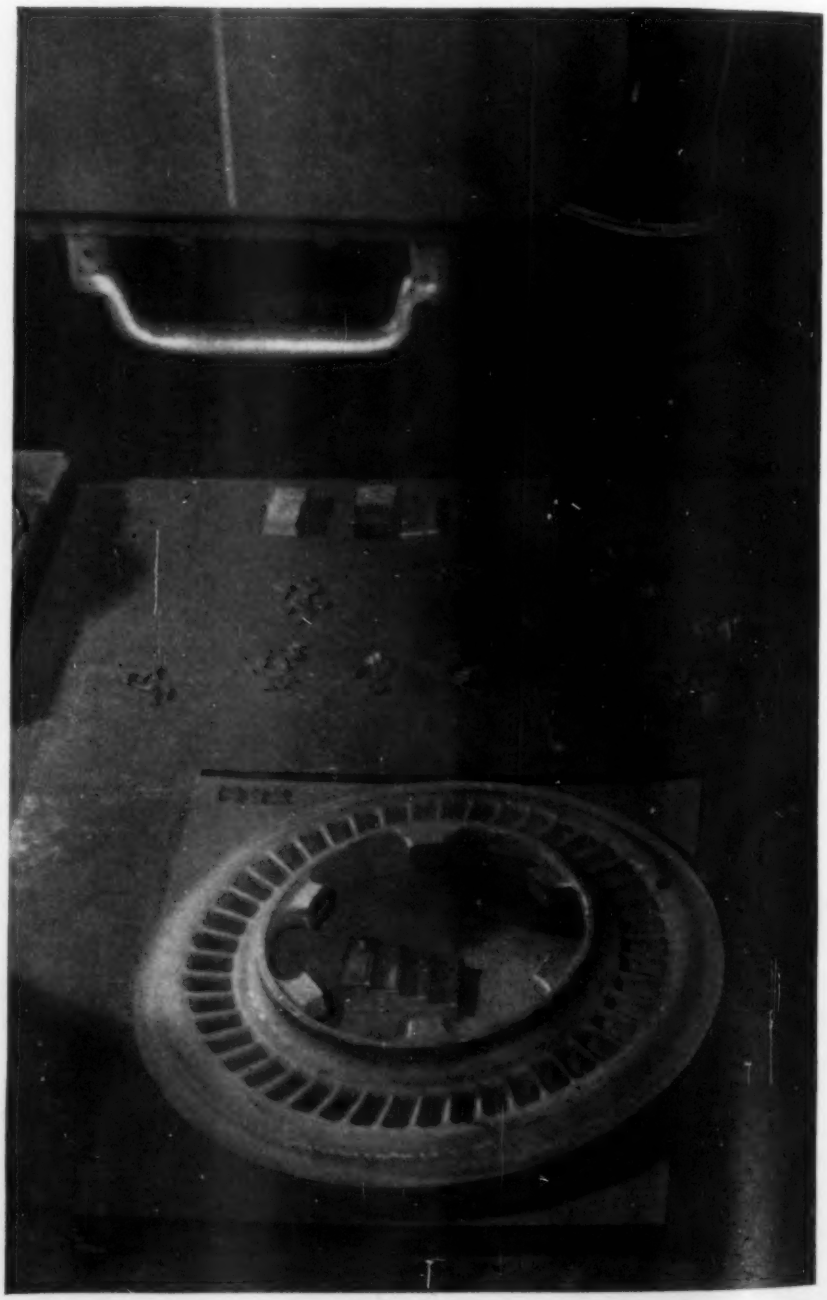


Fig. 1. 250-kv x-ray machine being used for examination of austenitic alloy supercharger casting. This is an example of production work subject to 100 per cent radiographic inspection. Practical section limitation about 1½-in. thickness.

Qualities of castings involve chemical, metallurgical, physical and soundness properties. Since the major portion of service failures is attributable to unsoundness, such as surface and subsurface defects, this discussion is aimed to cover applicable methods to determine qualities of casting soundness and requisites to promote their production. Illustrations in the paper pertain to steel casting production, but the scope may include other types of castings.

Records show that quality control of steel castings for railroad use started many years ago. Regular production castings were subjected to severe overloads and often tested to complete destruction. These tests afforded the engineer and steel foundryman fundamental information as to the quality of the casting, the effect of variable degrees of external and internal defects, as well as valuable design data. With these test results, coupled with actual service records, standards of acceptance for railroad steel castings were formed and now are reasonably well understood.

Service Performance

Many railroad castings are in service today after 35 years of satisfactory performance, yet these castings may contain surface and internal defects of varying proportions. These defects are not considered injurious, whereas similar defects in a highly stressed airplane landing gear, for example, could not be permitted.

When the highest level of casting quality and uniformity is desired, the production of such parts requiries: (1) strict adherence to design principles for directional solidification, (2) exacting process controls for manufacture, (3) non-destructive inspection for the standard of quality required.

The highest standards for casting quality require more confining and exact process controls of each operation, since the quality must be engineered into the casting and cannot be derived by inspection means alone. In order to be assured of the high quality desired, certain extra or special methods of inspection are involved, such as radiography, magnetic particle surveys, or other searching non-destructive tests for hidden discontinuities of the minutest nature.

Naturally, these special operations add appreciably to the cost of the manufacturing process and, since economic considerations cannot be neglected, the desired quality level must be carefully selected with that factor in mind.

Repair Welding

Steel castings, like steel ingots, are subject to surface and internal defects. These defects must be found and removed when required but, unlike billets or blooms, must also be filled in by proper repair welding. Satisfactory repair welding can be accomplished with rods capable of producing comparable physical properties after final heat treatment.

Failure to observe the proper foundry control and design requirements leads to defects such as refractory or mold sand defects, internal shrinkage, cracks and tears, gas porosity, slag, etc.

Inspection for quality several years ago depended largely upon visual examination alone. Other tests for soundness were destructive and, obviously, were not used on

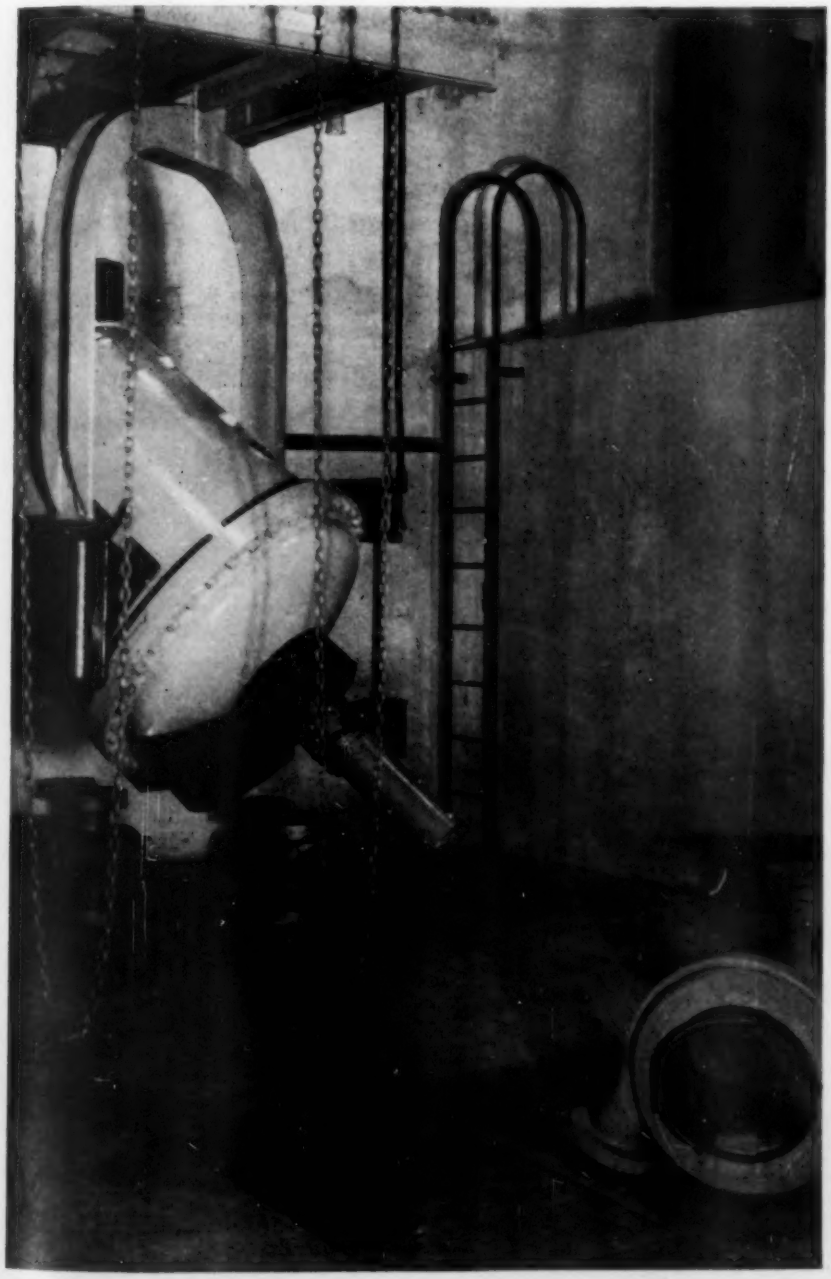


Fig. 2. 1000-kv x-ray machine being adjusted prior to exposure. Practical section limitation about 6-in. thickness.

production castings. Visual examination still remains an effective means of quality inspection for commercial grades of steel castings; however, when the requirements become more exacting, other methods of inspection are employed.

Radiography is the most common and presently accepted method of non-destructive testing (Fig. 1 and 2). Details of procedure need not be included here since excellent technical articles are available.[1] Since 1926, when some of the first high pressure cast steel fittings were x-rayed, gamma-ray (radium) techniques (Fig. 3) have been developed for the heavy sections of castings involved. The cost of films and the operating expense has limited this inspection to critical castings.

Testing Pilot Castings

Recently, radiography has been accepted by many foundries as a working tool for the complete exploration of all pilot castings before entering production. With such practice on pilot castings, the foundry is aware of internal shrinkage or other defects and may improve methods or make design changes to overcome these difficulties before production is started.

This method of procedure has definitely raised the general quality level that may be expected from steel castings, since properly controlled repetitive methods should yield equally sound castings without the expense of 100 per cent radiographic inspection.

Radiographic working standards with proper interpretation are essential and must be clearly understood by both the foundry and casting user. Incomplete knowledge or lack of workable standards for acceptance may defeat the useful purpose of this test.

Power Ratings

X-ray machines as used for casting inspection have increased in power ratings, thereby decreasing time of exposure and increasing sensitivity and applicable thickness ranges. Two-million volt x-ray machines are now in operation on heavier sections of castings, resulting in radiographs with greater sensitivity. The 20-million volt betatron has been announced as capable of penetrating about 18 in. of steel with considerable sensitivity and latitude.

With these newer methods and improvement of technique, radiographs proper are now being produced at a higher level of quality. Casting defects are more clearly defined with less general fogging and, if compared with acceptable standards produced under the older methods, allowance should be made for the increased degree of sensitivity attributable to the improvement in the radiographic technique alone.

With these improvements of process and technique for superior sensitivity of defects, complete new sets of radiographic standards pertaining to each method of radiography should be made available to the foundry.

Stereoscopy, i.e., three-dimensional radiography, promises to be of aid to the foundryman by locating defects normally disclosed by radiographs. This is done by two separate radiographs of the same casting with a focal spot movement of known quantity. Then, by calculation with distances and other factors, and the aid of a stereoscope, the location of the defect can be determined. Sharply defined x-rays permit accurate determinations of size and position of the defect.

Magnetic Particle Test

Within the past decade the magnetic particle test has gained prominence as an inspection method applicable to magnetic types of castings. The American Society for Testing Materials has adopted a tentative standard[2] for this method, and several technical articles are available on the subject.

The magnetic particle test is primarily limited to surface defects. When the proper flux density is applied with a subsequent magnetic powder application, surface discontinuities are quickly indicated by a buildup of the magnetic powder. A casting may be completely surveyed at a lower over-all cost than is possible with radiography.

In some instances, sensitivity of the magnetic particle test is greater than radiography due to the plane of the defect, which generally is at right angles to the surface. Small cracks of only a few thousandths of an inch in depth are revealed, which the average x-ray test would not disclose. Of course, defects of this nature normally are not visible to the eye.

The magnetic particle test, on the other hand, has not proved to be of any particular value in disclosing internal defects and, therefore, does not replace radiography but rather supplements it. The magnetic particle test usually is applied only to the higher types of quality castings (Fig. 4), where service requirements justify the added expense.

Fluorescent-Penetrant Inspection

Fluorescent inspection is now being applied to both non-magnetic and magnetic types of cast materials. This method is useful only for the determination of surface defects. Internal defects are not disclosed, therefore the end result of this method is similar to the magnetic particle test. The equipment required is not extensive, nor is any part of the operation difficult.

The casting to be inspected is immersed in, or sprayed with, a prepared highly fluorescent, light, water soluble, penetrating oil. After approximately a 10-min: soaking period, the oil is allowed to drain from the casting and the excess oil on the surface is washed away with warm water.

After drying, the casting is ready for examination in a darkened room under a violet-ray light (black lamp), with or without a developing powder dusted over the surface. Any surface defects then appear to fluoresce with a degree of brilliance in accordance with size and direction (Fig. 5).

Supersonic testing, while a comparatively new method, is now being used for the inspection of castings. The instrument (Fig. 6) sends supersonic vibrations through the casting under test, and measures the length of time it takes these vibrations to penetrate the material, reflect from the opposite side, or an internal defect, and return to the sending point. An oscilloscope pattern is produced on the screen, providing a visual indication of any defects which may be present in the material under test (Fig. 7).

Supersonic testing is limited to relatively smooth, flat surfaces pro-

viding good contact for the searching unit crystal. Sandcast surfaces are too rough and require hand grinding. The method is not yet applicable with uneven and curved surfaces.

There appears to be no limitation as to section thickness for the foundryman, as sections up to 15 ft. have been successfully penetrated with a good degree of sensitivity. Small, irregularly shaped castings with thin-wall sections are frequently difficult to test supersonically if they do not provide suitable surfaces from which the beam can be projected into the body of the casting at right angles to the available testing surface.

The minimum wall section is of about ½-in. thickness. Testing of thinner sections is sometimes practical when the sound vibrations can be passed through the flat edge, which actually represents a much heavier section (width).

Metals should be dense and with a cast grain of reasonable size. Austenitic alloy castings with large crystal size and variable crystal orientation cannot be penetrated, which probably is due to sound vibrations from the crystalline grain boundaries.

Sensitivity of the present equipment appears to be adequate for the exploration of castings. It is reported that defects in regular steel can be detected if the defect size is 0.1 per cent of the distance from the testing surface. At a testing distance of 5 in., a defect of 0.005-in. thickness should be easily detected.

Records of defects are not re-

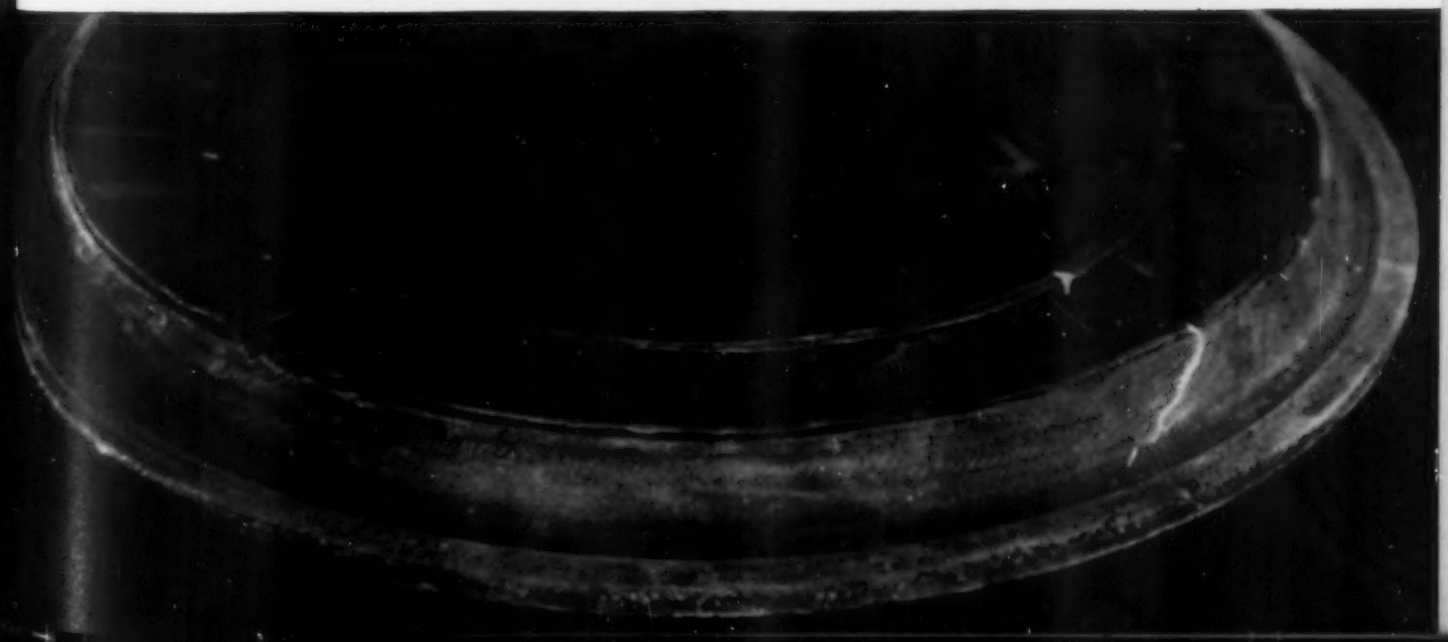


Fig. 3. Gamma-ray exposure of high strength steel alloy airplane landing forks prior to shipment. Another example of 100 per cent radiographic inspection.



Fig. 4. Magnetic particle testing on a high pressure low alloy valve body during the cleaning process. Surface defects which would not be found by normal visual inspection are easily detected. Operators are checking the areas for bore defects.

Fig. 5. An austenitic alloy supercharger casting being inspected by the penetrant-fluorescent method. Surface defects, which are cracks in this instance, fluoresce with a brilliance that would not be disclosed by visual examination alone.



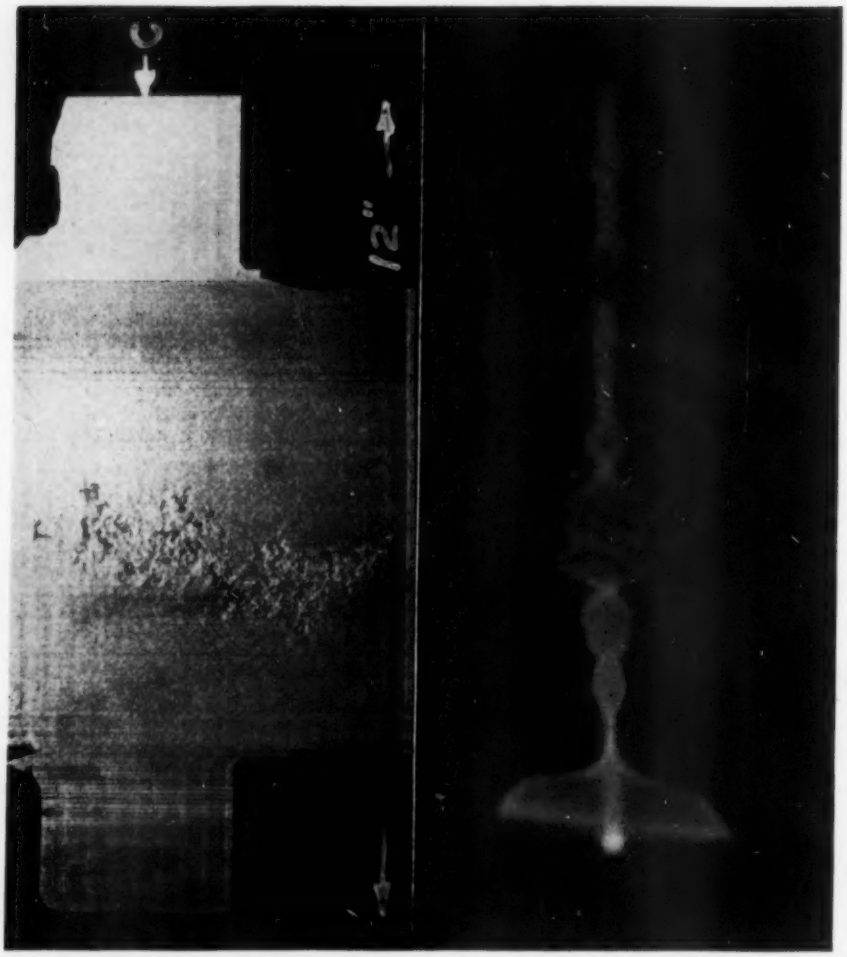


Fig. 7. A 12-in. casting sectioned showing internal shrinkage. The quartz reflectoscope crystal was placed at "C" with the resulting oscilloscope pattern shown at the left. The two large pattern reflections represent the sending point "C" and the opposite wall. Variations of pattern within these sections represent discontinuities.

corded, as all images are visually read from the oscilloscope pattern. Defects can be disclosed only when the beam reflections pass over the area at right angles. The location on one plane is established, and by calculation the distance from the surface is estimated. In order to explore all areas, complete scanning of all sections is required. This appears to be a tedious method and, if full coverage is required, radiographic examination might prove to be more economical.

The supersonic reflectoscope is an instrument which provides non-destructive testing of castings that other methods do not afford. It supplements radiography. This is a new tool for the foundryman, and it is now commanding considerable attention for the exploration of castings.

Other Non-Destructive Tests

Tests such as hydrostatic and air-pressure tests are employed in the steel foundry. Proof tests or static loading to a point just below the yield point, with measurement of any permanent set, are often made on rough or finished castings.

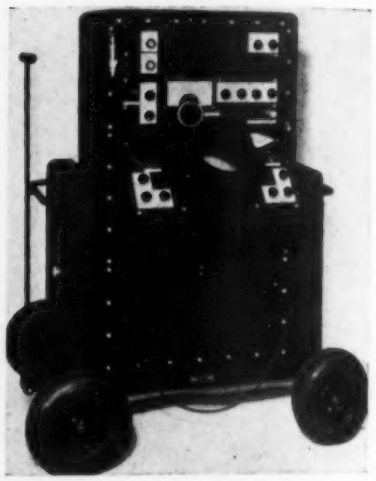
Fig. 6. Supersonic testing equipment capable of penetrating massive sections up to 15 ft.

Jig-gaging on production castings as a final inspection is used (Fig. 8) as a means to insure adequate stock for machining, including machine tool fixtures. Individual layout procedure, with thorough checking of dimensional tolerances and wall thickness, is always a recommended practice on pilot sample castings.

It is difficult to indicate limits of dimensional tolerance in a blanket manner, but plus or minus 5 per cent or plus or minus 1/16-in. wall thickness in the thinner wall sections has been accepted for certain classes of steel castings.

Design Precautions

Quality steel castings are not dependent upon non-destructive testing methods alone. Quality castings begin during the early stages of design. Close collaboration between the designer and the steel foundryman is vitally important, but only too often neglected. Good, practical designs, which permit the foundryman to carefully plan his production methods, with directional solidification possibilities built into the design, are essential for highest quality.



It is not possible, for example, to feed metal from thin sections to thick sections, since the thinner section freezes early, while the heavy section still is molten. Directional solidification involves controlling the rate of solidification so that freezing normally will occur at the thinnest cross-section and gradually progress from that point toward the heaviest cross-section involved, which then may be adequately fed with hot metal from a suitably placed riser.

With the tapering and blending of wall sections built into the original design, sounder castings with freedom from internal shrinkage can be produced. Other methods of indirectly controlling directional solidification are available, such as the differential thermal gradients result-

ing from the planning of heading and gating, effective chilling of areas or padding (tapering) of sections, which padding subsequently must be removed (Fig. 9).

Designs with restraint should be avoided, as these often result in hot tears. Blending or streamlining heavy sections into lighter sections always is desirable, as well as the avoidance of abrupt angles with inadequate fillets. Proper designs always afford the foundryman an opportunity of feeding the casting adequately, thus avoiding internal shrinkage.

Good quality castings cannot be produced from every design, and this should not be expected. Many complex or "impossible" steel casting designs can best be produced with two or more castings welded into a unit.

Process controls of actual foundry operations are not only desirable but essential for the highest quality levels. Each and every step of production must be fully controlled in order to coincide with previous or subsequent operations.

Properly made pattern equipment is equally as important as design. Dimensional accuracy is necessarily dependent upon the pattern and core box equipment. Exceptionally close tolerances of castings are reproduced with proper pattern plus



Fig. 8 (top). A heat treated, high tensile alloy casting being jig-gaged in target fixture before shipment. This inspection insures proper and adequate finish allowances with over-all dimensional tolerances suitable for subsequent machining.

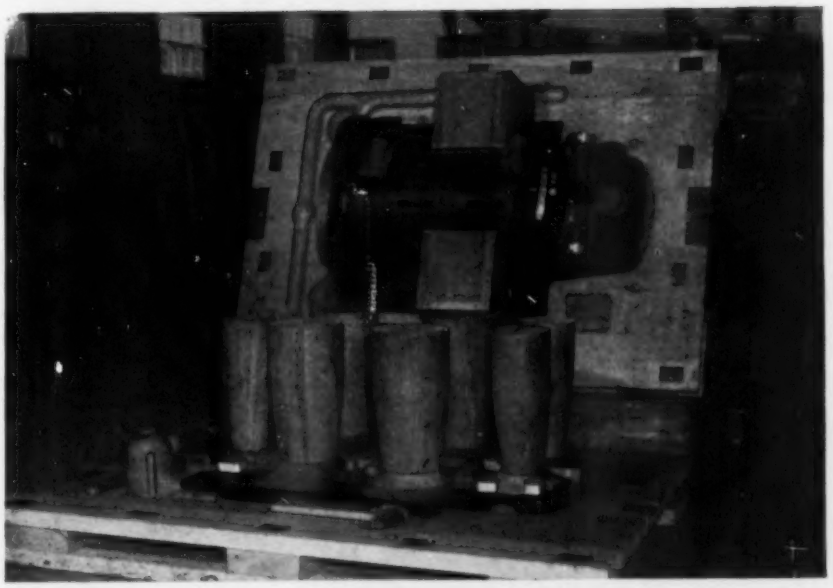


Fig. 9 (center). Pattern equipment used for pump case casting Cope and drag pattern boards are shown with heading and gating carefully planned for a casting to meet radiographic standards for internal soundness. Note the padding required under each riser for directional solidification.

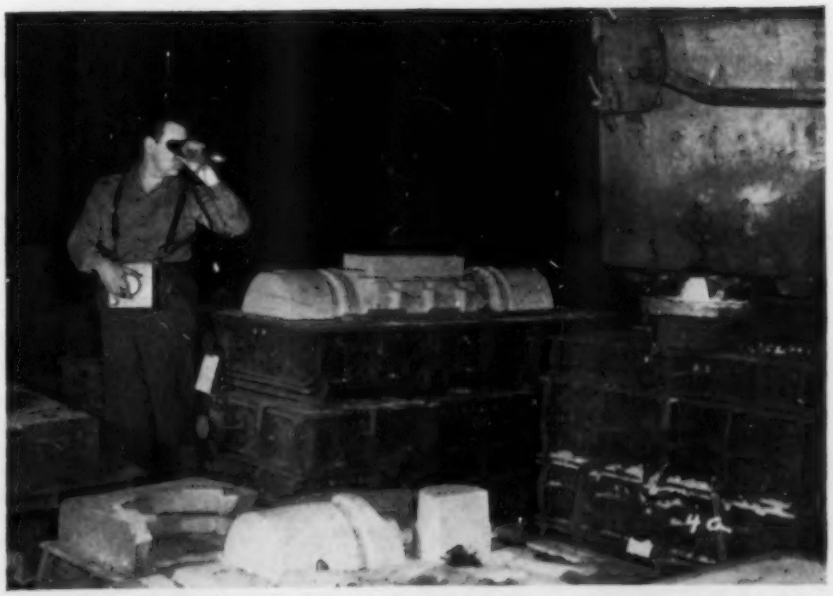


Fig. 10 (bottom). Checking the pouring temperature with an optical pyrometer. Mold as well as metal has been produced under process controlled conditions.

good molding equipment and technique. Surprising dimensional results are possible when engineered equipment and controls are properly maintained.

From this point on, the quality of the casting remains in the hands and head of the steel foundryman. He must carefully plan his methods of heading, gating, molding and pouring. The most preferable method of heading and gating is to incorporate these essential parts of the rough casting on the pattern equipment, usually mounted integrally on a board (Fig. 9). This step leaves little to the individual molder's judgment and insures duplication of the method on subsequent castings once good practice has been established.

Process controls covering all other phases of operation are required. Variables in the molds and cores are minimized by material controls and consistent check of manually controled operations. Good molds to receive metal of the proper metallurgical qualities are essential.

Pouring is one of the foundry operations which it is most necessary to control. Metal temperature varies, with consequent effects upon fluidity. Pouring rates invariably are manually controlled. Variation of both temperature and pouring rate vitally affects the soundness and cleanliness of the castings.

Temperature checks may be optically recorded (Fig. 10), and speed of pouring adjusted accordingly so that rates of freezing may be uniform and directional. Since these pouring variables are recognized as difficult to control, ample safety allowances must be made in design, heading and gating methods.

Adjusting Pilot Castings

After production under known conditions, the pilot casting is subjected to visual and non-destructive examinations by one or more of the previously outlined methods. If this sample casting fails to meet the quality or grade desired, changes usually are made in the basic foundry planning, or the customer's engineering department is requested to make adjustments in design. Once the pilot casting has met the quality requirements, production may start and, under the same controlled methods, similar quality castings may be reasonably expected.

Production castings, made by a satisfactorily controlled process, normally should possess no major defects. Refractory or other blemishes may occur, usually of surface origin. These minor defects must be visually located and repaired by welding, using a welding rod which produces properties comparable to those of the parent metal after final heat treatment.

Heat-Treating Factors

Under process controlled operations, it may be assumed that a satisfactorily sound casting has been produced, and which is dimensionally correct and ready for subsequent heat treatment. Heat-treating operations should be conducted under well controlled conditions, in uniformly heated and properly constructed furnaces.

Simple annealing treatments still are satisfactory for lower tensile strength requirements but, within the past few years, homogenizing, normalizing, quenching and tempering (or any combination) under controlled conditions has had a great effect upon the use and performance of high tensile steel castings. Cast armor is a good example of a casting on which exacting controls for quality must be maintained in heat treatment since ballistic testing of each casting is not possible.

Of course, after heat trentment, physical properties of the castings are checked by process control methods such as tensile, impact, hardness and other tests. Hardness tests often are taken on a percentage of castings from each heat treatment batch and, when required, 100 per cent of the castings are so inspected.

Visual final inspection of all castings is essential. Additional inspection of high quality castings may entail a 100 per cent inspection by one or more of the foregoing nondestructive methods. When satisfactory quality is consistently found and maintained, the frequency of testing may be substantially reduced.

Inspection by any method adds to the ultimate cost of the casting. Unnecessary inspection is, therefore, a waste of man hours and should be held to a minimum. This can be accomplished by a thorough understanding of the service requirement, with specifications properly written for the grade or quality needed.

Interpretation of acceptable standards of steel casting quality is of utmost importance. Frequently, inexperienced inspectors have misinterpreted requirements of acceptable standards of quality which directly resulted in waste of useful castings, or an excessive amount of rework and repair. Acceptable standards, therefore, should be clearly defined and mutually agreed upon before undertaking any commitment.

References

1. *Radiographic Testing of Metal Castings*, ASTM Tentative Standard E-15-39T.

2. *Magnetic Particle Testing and Inspection of Steel Castings*, ASTM Tentative Standard A272-44T.

DISCUSSION

Chairman: C. H. LORIG, Battelle Memorial Institute, Columbus, Ohio.

Co-Chairman: J. A. RASSENFOSS, American Steel Foundries, East Chicago, Ind.

G. VENNERHOLM[1]: I would like to bring to the attention of those who may not have had an opportunity to examine some of the new instruments developed for temperature readings of the liquid bath, that several such direct reading instruments are now available.

We have installed the Brown instrument which consists of an insulated thermocouple which is inserted right in the molten metal and the instrument which utilizes a rayotube. The latter is placed into the bath and a stream of air blown through the tube permitting temperature readings to be taken on the metal wall formed by the air bubbles.

The Brown instrument being a direct reading unit, is very accurate, but has the drawback that it requires considerable time to replace the refractory tube and remount the thermocouple after each reading.

The L. and N. instrument, on the other hand, offers greater possibilities because readings can be taken more or less continually thereby allowing a complete temperature log to be constructed for each heat.

I believe that these instruments will be of great help to the steel foundries as they will enable the operator to study the reactions taking place in the furnace with a much better understanding than has been possible in the past.

H. H. BLOSJO[2]: The author showed a casting with a hub on each side, in connection with the supersonic method of testing. In checking the soundness of this hub from one end, there seemed to be no effect from the sides of the hub. Does this test check only a straight line or does it cover an area?

[1]Ford Motor Co., Dearborn, Mich.
[2]Minneapolis Electric Steel Castings Co., Minneapolis.

MR. JUPPENLATZ: In answer to that question, my observation would lead me to believe that it just covers the entire mass from point of crystal contact "C", Fig. 7. Published technical illustrations by Sperry clearly illustrate oscilloscope patterns through solid sections and those containing varying degrees of defects. With the oscilloscope pattern shown in Fig. 8, the quartz crystal was placed at point of contact "C," Fig. 7.

This causes vibrations to be sent from the quartz crystal in a path of about ⅛ in. in diameter through the casting. The reflections go straight and rebound to the crystal on the off cycle. The two large oscilloscope aberrations represent the sending point and back wall respectively. The small aberrations are indications of voids, or shrinkage in this case, and their relative position.

In order to make a survey of the entire casting, the quartz crystal would have to be slowly moved over the surface of the casting, noting the oscilloscope pattern. It is a rather tedious, slow method of examination, especially if you desire to cover all areas, all surfaces.

H. H. JOHNSON[3]: Have you found the control chart method useful as an aid in controlling the various steps of the process?

MR. JUPPENLATZ: Yes, we have used some mathematical methods during the war. Under Army specifications a great deal of testing was reduced. I am thinking particularly of the physical tests. Even chemical testing was reduced due to the process controls in accordance to the degree of control that was exhibited in the individual foundries. We are acquainted with those methods.

[3]National Malleable & Steel Castings Co., Cleveland.

MEETING FOUNDRY COMPETITION WITH A SOUND COST PROGRAM

A. E. Grover
Cost Engineer
Cleveland

DURING World War II, management was primarily interested in producing the implements of war which were so necessary to an early cessation of hostilities. There was a tendency among manufacturers, including producers of castings, to accept the collective planning necessitated by the war, and guarantee of cost recovery, as a justification for a general letdown in cost control.

Those conditions no longer exist. Management must now recognize the importance of studying present cost procedures to ascertain if changes should be made to control postwar costs for use as a basis for establishing a dependable sales policy for each item of production.

Sales Policy Necessary

Whether or not price ceilings are continued, there will surely be no restrictions placed on foundries who may bid for business at prices below any established ceiling price. Therefore, the wise foundryman will enter competition with a scientifically prepared basis for establishment of a sales policy for each casting to be produced, to put him in a position to select the most profitable work.

As a matter of fact, the very existence of any foundry may well depend upon how much intelligent effort it has put into analyzing the cost of the various castings it will make.

In many foundries more effective cost procedures will have to be inaugurated and this must be done even if it takes more clerks to do the job. It is distressing to find so many cost accounting systems designed and operated primarily to control shop labor and to satisfy the requirements of loaning banks and Federal agencies, with little attention given to the preparation of costs and statistics so necessary for meeting competition.

A tremendous amount of study is being given to establish operation standards intended to increase the number of units produced. No expenditure seems to be too great to accomplish this. Management, however, too often frowns on necessary expenditure to study the other activities of the business and to prepare bases for intelligent cost estimating, necessary to tell management what items to manufacture and at what price to accept orders.

No matter whether prices are established by supply and demand, competitive bidding, or established Government ceilings, there is no law that compels the manufacturer to accept or reject specific orders for manufacture. There is, however, an obligation that every corporation management owes to its stockholders. It is to manage its affairs so that their investment may be protected and at the same time earn all the profits the traffic will bear. To accomplish this there must be good merchandising and good merchandising can only be had by a sound cost program. Sound costing procedures require the application of considerable common business sense.

We hear a lot about a cost system for the small foundry and another kind of system for the large foundry. Contact with activities in both small and large foundries has proven to the writer that the cost problems of the small foundry are the same as those of the large foundry, and that the solution of these problems are the same, provided the small foundry makes the same kind of castings as the larger competitor. The difference lies in the number and size of the cost calculations.

The determination of the proper base to use in establishing cost standards is all important and must be predicated on a thorough analysis of the several cost elements as related to the specific activity. It must always be definitely borne in mind that dependable results can never be attained by averaging unrelated elements.

Securing business in a standard commodity market requires inclusion

Dependable normal costs can be determined only on the basis of the company's capacity to produce with the present operating facilities. Past records must be used only as a guide in establishing fair and reasonable costs to be applied to the present potential capacity of the foundry. The cost establishing procedure herein set forth may be used in the gray iron, steel, malleable and non-ferrous foundries.

Presented at a Foundry Costs Session of the Fiftieth Annual Meeting of the American Foundrymen's Association at Cleveland, May 9, 1946.

of material cost at the prevailing market price at the time the product is to be manufactured. Otherwise, sales prices based upon actual costs that are higher than the market, might cause the refusal of business that would, if it had been accepted, allow the company to realize an operating profit.

On the other hand the inclusion of material costs at current market prices, which exceed actual costs, causes the inclusion of expenditures for materials in excess of the amount calculated in the estimate on which the sales price was based, thus unwisely giving the customer profits earned by the company through good purchasing.

Primarily, the company is organized to make profit through the conversion of materials into a condition that will be usable to its customers. It is the measuring of this conversion into the various items of product that concerns labor cost accounting.

Time is, of course, the real foundation of all labor costs. Upon time, are based all methods of wage payment, also all methods of distributing costs of labor to individual items of product.

Labor Standards

Dependable labor standards are based on the *normal time* that it should take to perform the operation, and at labor rates prevailing in the locality in which the plant is operating. It makes little difference whether wage payments are made on the basis of premium, piece, or straight hourly rates, just so long as the principle of basing the cost upon the time in which a job *should* be performed is adhered to.

Foundrymen generally base their estimating data on normal operations, either by:

1. Averaging actual operation costs over past years,
2. Using tonnages produced in periods of estimated normal operation,
3. Using an arbitrary per cent of tonnage shipped in the best six months of the last ten years or more, and
4. Consideration of the physical capacity of the foundry to convert metals into shapes and sizes suitable for the customer's use.

Averaging actual operating costs over past years is an unsound basis, because the average is very likely to be different from actual conditions today.

Using tonnage shipped in periods of estimated normal operation does not take into account the constant changes in types of castings from year to year. For instance, 1940 might be considered a normal year and the production have been 500 lb. per molder day. The foundry might by increased efficiency of its men or by having a different type of casting in the sand produce 700 lb. per molder day. This basis does not take into account the variations in weight produced by different molders in the same day or by the same molder on different days.

Using an arbitrary percentage of tonnage produced in the best six months in previous years lends itself to the same irregularities in production per molder day as the use of estimated normal operation.

Consideration of the physical capacity of the foundry to convert metals into shapes and sizes suitable for the customers' use eliminates these inaccuracies and irregularities of job costs caused by changed operating conditions as well as changes made in the product.

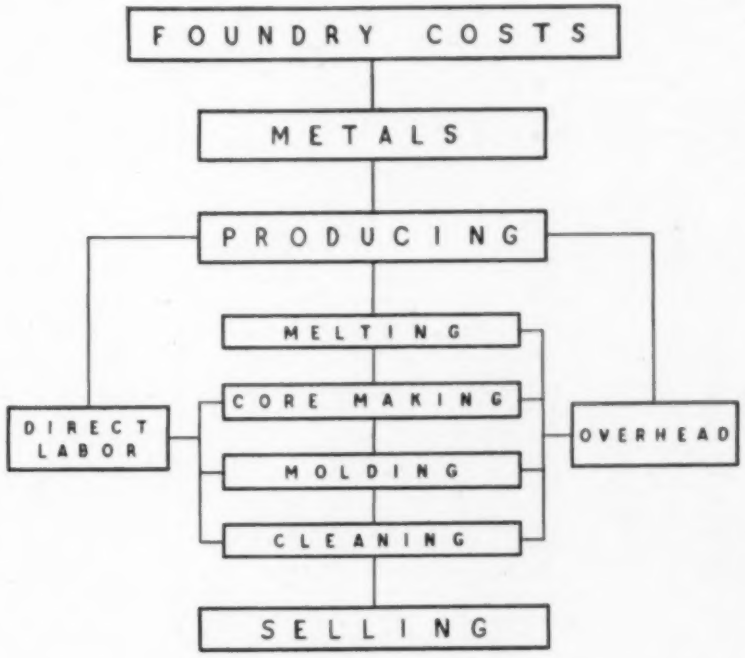


Fig. 1—Chart shows foundry costs broken down into three major classes: metals, producing and selling.

The following illustrations are intended to assist the reader in understanding the principles of cost engineering which the author will present. The writer, in the illustrations to follow, has selected a small gray iron foundry as an example because there is really little difference in principle between cost procedures adaptable for it and the larger foundries. The same principles may be logically applied to steel, malleable and non-ferrous foundries by taking into account some additional operations not required to produce a simple gray iron casting.

In Fig. 1 the writer has divided foundry costs into three major classes: Metals, Producing and Selling. He has further divided producing into the four basic operations necessary to convert the metals into shapes and sizes suitable for the customers' use: melting, core making, molding, and cleaning. To measure the cost of each of these operations the expenditures are divided into direct labor and overhead. Further explanations and definitions of direct labor and overhead probably are unnecessary.

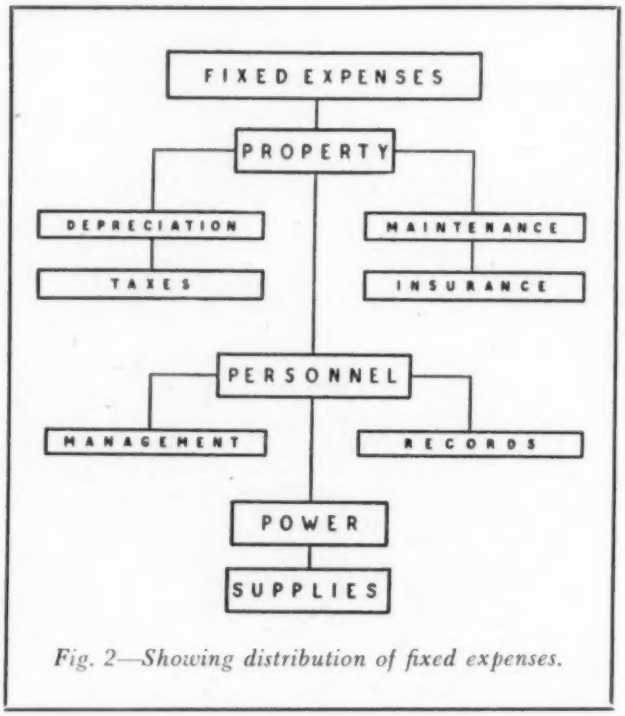


Fig. 2—Showing distribution of fixed expenses.

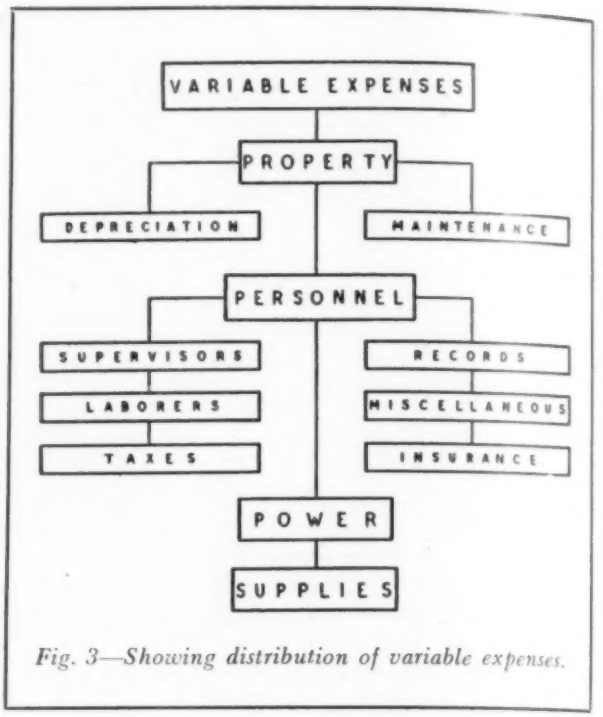


Fig. 3—Showing distribution of variable expenses.

For the purpose of establishing a basis for meeting competition with a sound cost program it is essential to predetermine the probable "break even" or "out of pocket" point. This necessitates the consideration of those expenses of a fixed character separately from those that vary with different volumes of production.

Fixed Expenses

In Fig. 2 fixed expenses are classified as property, personnel, power and supplies.

I. Property expenses:

1. Depreciation—Obsolescence of plant facilities only,

2. Assessed property taxes,

3. Maintenance and repairs to operating facilities,

4. Insurance on operating facilities.

II. Personnel expenses:

1. The salaries and expenses of executives, stenographers, clerks and others directly concerned with the management of the business, that do not vary because of fluctuations of production,

2. The salaries, wages and expenses of employees engaged in collecting, recording and reporting operating costs and statistics, that do not vary because of fluctuations of production.

III. Power expenses:

That portion of the power bill estimated to cover lighting the foundry.

IV. Supplies expenses:

Supplies used in maintaining operating facilities that do not vary because of fluctuations of production.

Variable Expenses

In Fig. 3 variable expenses are also classified as property, personnel, power and supplies.

I. Property expenses are:

The usage factor of depreciation, and

Maintenance of land, buildings and equipment.

II. Personnel expenses are:

The salary and wages of supervisors, laborers and those employees engaged in collecting, recording and reporting operating costs and statistics. These salaries and wages fluctuating approximately with varying volumes of production.

III. Taxes and insurance on wages of employees.

IV. That portion of the power bill estimated to operate the production facilities.

V. Supplies used in producing.

So long as there are normal periods, booms and depressions, individual companies must necessarily base the sales policy of each commodity at standard cost predicated on normal operations.

Operating Standards

Many companies establish their operating standards from averages of past performance. This is all right if the past performances have been comparable with present conditions. Such a basis of past performance is feasable only if the following conditions obtain:

1. Raw material prices must have been the same as prevailing today.

2. The wage rates and methods of bonus or premium payment, then prevailing, must have been the same as today.

3. The volume of business must have been in the same ratio to capacity of the plant, as the expected production.

If normal costs are based on averages of actual during high, normal and low production then the periods must be of equal length and with the same volume in each period.

It could happen that a particular foundry did not operate to capacity at any time in the past. Others may not have kept records of tons produced. Then there are foundries whose production per man per day has changed due to a different type of work. Many foundries have in-

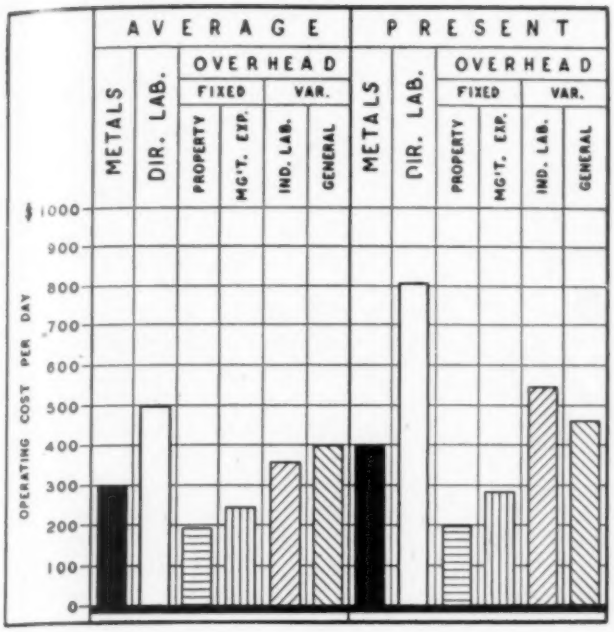


Fig. 4—Showing standard operating costs set up on past performance as a basis.

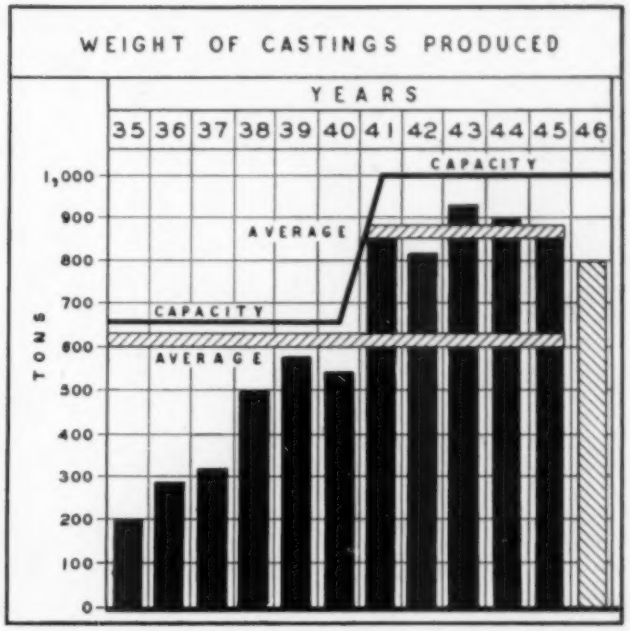


Fig. 5—Showing normal costs established on the basis of tonnage production.

creased their tonnage by the introduction of modern equipment or premium wage systems, often both.

Past Performance

The writer prepared Fig. 4 for the purpose of illustrating what could happen to the average foundry, whether it be large or small, when using past performance as a basis for setting up normal standard operating costs, that can be intelligently used in establishing sales policies. By sales policies, the writer does not necessarily mean setting the selling price. That price may be already established. However, a sales policy means whether it is wise to accept business or reject it.

Costs

The writer has classified costs as:

1. Metals,
2. Direct labor,
3. Overhead.

He has further classified overhead as fixed and variable. Fixed overhead includes property expenses and management expenses. Variable overhead includes indirect labor and general expenses.

These same classifications are shown separately for average of past performance and present operating conditions. Metals expense, direct labor, fixed property expenses, management expenses, indirect labor and general expenses are indicated in Fig. 4. The ordinate is operating cost per day.

You will observe that there is no consistent or relative increase in the different classes of costs. For instance, the metals cost has increased due to economic conditions.

Increase in Expenses

There is no appreciable increase in fixed property expenses. However, if the foundry has increased its producing facilities, or reestablished values of old facilities on the more equitable basis of cost to reproduce, then there would be an increase in fixed property expenses.

Of course, there is and will be some increase in management expenses, as well as an increase in general expenses. The largest increase, of course, will be in labor.

One very important fact brought out in this exhibit is the utter absence of relation of the different classes of overhead to the direct labor. Therefore, a sound cost program cannot possibly be had if direct labor is used as a basis of allocating these expenses to the various operating departments and production centers. If this method is used it should be superceded by a method that analyzes each kind of expense caused by a specific department or production center and charged direct to it. Such expenses as cannot be so charged can be more equitably distributed on the basis of direct labor hours than direct labor dollars.

Normal Costs

The writer previously mentioned using tonnage as a basis for establishing normal costs. This calls for some analysis. Assume that a foundry has had experience as shown in Fig. 5.

The black columns represent tonnages produced over a period of eleven years. In 1941, due to increased business brought about by the war program, the foundry increased its operating facilities, as indicated by the black line.

Averaging the tonnage over the eleven years results in 620 tons. Rather a poor basis when the foundry has a capacity for 1,000 tons.

If we average the tonnage over the five war years we arrive at 875 tons. After the stress of war production, the foundry could not be expected to operate at that per cent of capacity, so we would be more sensible to set the normal or standard something as shown by the '46 column. If tonnage is used as a basis for establishing normal costs, then there must be uniformity in the class of castings produced.

Dependable normal costs can be

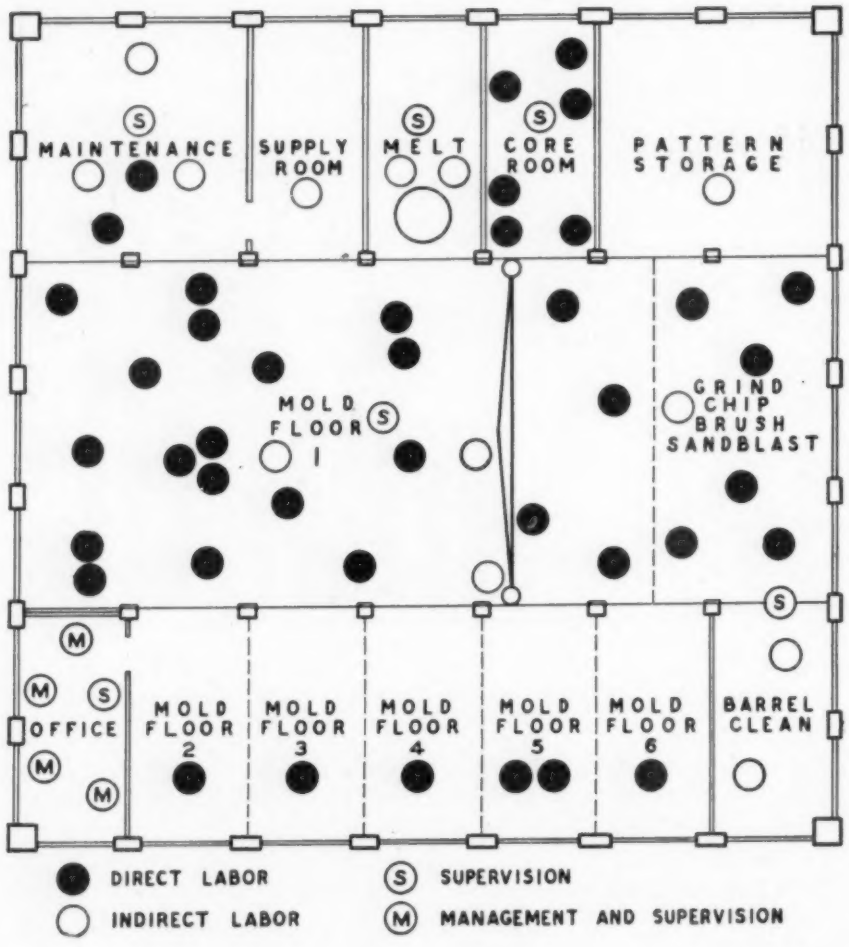


Fig. 6—Layout of Foundry.

determined only on the basis of the company's capacity to produce with the present operating facilities. Past records must be used only as a guide in establishing fair and reasonable costs to be applied to the present potential capacity of the foundry. In other words, there are certain items of cost that can best be determined by reference to past records in order to establish ratios of actual expenses of different characteristics to the volume of work performed.

There is a simple yet effective way to establish normal operation for a small jobbing foundry. The same procedure can be used in larger foundries because the costing problems of the large foundry, serving the same kind of customers, are the same as the small foundry. This procedure may also be used in steel, malleable and non-ferrous foundries.

The building is first laid out into departments representing the nature of work in each as shown in Fig. 6.

The floor area is determined for each department and subdivision. This area serves as a basis for departmentalizing:

Depreciation on Buildings,
Taxes on Buildings,
Insurance on Buildings,
Repairs on Buildings,
Heating Cost,
Lighting Cost.

Each department is then "manned," assigning to each a sufficient number of employes to utilize the area and equipment to capacity for the regular established working hours in the day.

The utility value of each department and subdivision is determined and used as a basis for allocating:

Depreciation on Equipment,
Taxes on Equipment,
Insurance on Equipment,
Interest on Investment (a return on invested capital applied on the estimate sheet as a form of profit).

Having thus established the physical capacity to produce under present operating conditions, cost sheets are prepared for each department and subdivision of a department. As molding is the principle operation in a foundry, it is necessary that the capacities of the other departments and subdivisions be in proper proportion to it.

The author recommends that you review the subject as covered in the March, 1937, issue of *Foundry* under the title of *Common Sense Cost Estimating*. There you will find a complete and comprehensive explanation of the method of establishment of this dependable basis for calculating costs of individual jobs.

Equally in importance to the establishment of the physical and personnel capacity of the foundry is forecasting of the normal working hours for each unit of production.

Table 1 shows the method the writer used in another industry to forecast normal hours of production.

There are 365 days in the calendar year; 52 Sundays; 6 legal Holidays; 52 Saturdays. We also anticipate that a two-week vacation allowance will be in order.

Taking these losses of working time into consideration there remains 243 days for actual production. Believing that industry will go on an eight-hour day, there will be 1,944 available working hours for each producer and other individuals.

It is not possible to utilize the producing facilities 100 per cent of the time so an adjustment must be made to allow for unavoidable down time. The figures are used to illustrate the principle and must not be accepted in your plant without a proper study of your own activities and expectations.

Table 1

NORMAL WORK DAYS AND HOURS OF PRODUCTION

Days in one year		365
Less—Sundays	52	
Holidays (Legal)	6	
Saturdays	52	
Vacations	12	122
Working days in a year		243
Working hours per day		8
Working hours per year		1944

	Working Hours	*Efficiency, per cent*	*Normal Hours*
First shift	1944	90	1750
Second shift	1944	85	1652
Third shift	1944	80	1555

BASIS FOR ESTIMATING			
PREPARED BY _____ DATE _____			
	NORMAL		
SHIFTS / HOURS	ONE	TWO	THREE
PRODUCTION COST			
VARIABLE			
LABOR			
EXPENSES			
TOTAL			
FIXED			
TOTAL			
SELLING COST			
ADMINISTRATIVE			
TOTAL COST			
USE OF CAPITAL			
DEP'T. _____ PRO. CEN. _____			
DEP'T. _____ PRO. CEN. _____			
DEP'T. _____ PRO. CEN. _____			

Fig. 7—Form of table for cost estimating.

ESTIMATE SHEET				
WEIGHT IN MOLD	30,000			
LESS GATES & SPRUES	10,000	LABOR	OVERHEAD	
GOOD CASTING	20,000			
METAL COST	$ 360		VARIABLE	FIXED
LESS G & S REMELTED	100		$ 260	
PRODUCING COST				
MELTING		$ 36	38	$ 6
CORE MAKING		90	120	60
MOLDING		250	240	100
CLEANING		32	25	8
TOTAL PRODUCING COST		408	423	174
RETURNS	$ 79			
LESS REMELT VALUE	13		66	
SELLING COST 5%				50
TOTAL COST				1,381
PROFIT				
NORMAL INTEREST (5% OF INV. CAPITAL)				72
OPERATING (10% OF PRODUCING COST)				138
SELLING PRICE				
ESTIMATED				$ 1,591
COMPETITIVE				$ 1,700

Fig. 8—Cost estimate in Case No. 1 (see text).

ESTIMATE SHEET				
WEIGHT IN MOLD	30,000			
LESS GATES & SPRUES	10,000	LABOR	OVERHEAD	
GOOD CASTINGS	20,000			
METAL COST	$ 360		VARIABLE	FIXED
LESS G & S REMELTED	100		$ 260	
PRODUCING COST				
MELTING		$ 36	38	$ 6
CORE MAKING		90	120	60
MOLDING		250	240	100
CLEANING		32	25	8
TOTAL PRODUCING COST		$ 408	$ 423	$ 174
RETURNS	$ 79			
LESS REMELT VALUE	13		$ 66	
SELLING COST 5%				$ ~~50~~
TOTAL COST				$1,381
PROFIT				
NORMAL INTEREST (5% OF INV. CAPITAL)				$ ~~72~~
OPERATING (10% OF PRODUCING COST)				~~+38~~
SELLING PRICE				
ESTIMATED				$1,591
COMPETITIVE				$1,300

Fig. 9—Cost estimate in Case No. 2 (see text).

ESTIMATE SHEET				
WEIGHT IN MOLD	30,000			
LESS GATES & SPRUES	10,000	LABOR	OVERHEAD	
GOOD CASTINGS	20,000			
METAL COST	$ 360		VARIABLE	FIXED
LESS G & S REMELTED	100		$ 260	
PRODUCING COST				
MELTING		$ 36	38	$ ~~6~~
CORE MAKING		90	120	~~60~~
MOLDING		250	240	~~100~~
CLEANING		32	25	~~8~~
TOTAL PRODUCING COST		$ 408	$ 423	$ 174
RETURNS	$ 79			
LESS REMELT VALUE	13		$ 66	
SELLING COST				$ ~~50~~
TOTAL COST				$ 1,381
PROFIT				
NORMAL INTEREST (5% OF INV. CAPITAL)				$ ~~72~~
OPERATING (10% OF PRODUCING COST)				~~138~~
SELLING PRICE				
ESTIMATED				$ 1,591
COMPETITIVE				$ 1,000

Fig. 10—Cost estimate in Case No. 3 (see text).

Figure 7 shows a simple way to set up a table for use in estimating. These are loose sheets arranged in a multiple ring binder in such a manner that the unit costs of the various classes of expenses may be quickly found when needed in making estimates. The form is self explanatory.

Now that we have covered the most important principles involved in establishing normal costs, let us see how they apply to the establishment of a selling policy for an individual casting.

Figure 8 is an estimate sheet showing the various elements of costs and profits necessary to consider in estimating the selling price. This represents the cost of the complete job. However, the amounts of the various items could be shown as so much per piece or so much per pound to suit the convenience and desire of the foundryman.

Let us now consider three separate types of competition which will be met in the future:

Case No. 1—Competitive selling price at approximate ceiling,

Case No. 2—Competitive selling price under ceiling,

Case No. 3—Keen competition—Anxious for business.

In Case No. 1 (Fig. 8) we have established a selling price at $1,591.00 which covers all the cost

plus 5 per cent return in the invested capital used in producing this particular casting, and 10 per cent of producing or conversion cost. The estimated selling price being under that of our competitor, we would, of course, accept the business.

In Case No. 2 (Fig. 9) a competing company enters the market with a price of $1,300.00, which is $291.00 under the estimated selling price. By using this form of estimate sheet, with dependable cost data entered thereon, the foundryman is now in a position to ascertain what concessions in price should be made in case he is interested in bidding on this type of borderline business. He starts to cross out the items that he wishes to eliminate, as indicated by the lines across the figures shown in the last column.

He finds that he recovers all the metal, direct labor, variable overhead, and most of the fixed overhead. So, if he is operating below normal he would probably accept the business. However, if he is operating at normal he would be wise to refuse the work.

When operating below normal, as in Case No. 3 (Fig. 10), a foundryman is so anxious to secure business that he will bid far below the ceiling price established by law or any competitor's price.

This puts the foundryman in a precarious position for without a dependable cost system that will allow him to analyze the situation accurately and carefully, he may ignorantly commit the company to sales contracts that would not only pass along to the customer all the profit and fixed expenses, but some "out of pocket" money as well.

We assume that the competitor, either by pure guess work, or by computing his estimated sales prices by means of ineffective costing methods, thinks the price of $1,000.00 will recover all out of pocket money and some additional money to help pay for fixed expenses.

Observing the crossed out figures on the estimate sheet we naturally conclude that the foundryman with the dependable costing system would refuse to accept the business as he not only gives away all his profit and fixed expenses but also a portion of the variable expenses. This is just like borrowing money from the bank and giving it to the customer.

DISCUSSION

Chairman: R. L. LEE, Grede Foundries, Inc., Milwaukee.

Co-Chairman: L. M. NESSELBUSH, Falcon Bronze Co., Youngstown, Ohio.

J. F. FORAN[1]: In your discussion you took no account of foundry losses. Do you think the percentage of losses should be included in your cost?

MR. GROVER: It was shown on one of these estimate sheets, Fig. 10.

MR. FORAN: Will you amplify the item, selling cost 5 per cent?

MR. GROVER: The 5 per cent is determined by dividing the selling expenses by the conversion cost to get that percentage.

CHAIRMAN LEE: Whatever the relationship in your plant might be to selling cost.

MR. GROVER: Some companies are inclined to use total costs as a basis for establishing these percentages. These being terrific variances in material costs makes it a very dangerous thing to do. That would be particularly true in nonferrous foundries where a wide spread of material cost prevails. In fact foundries are selling their ability to convert material into shapes, sizes, and texture required by their customers.

C. B. TIBBETTS[2]: Where is your administration costs?

MR. GROVER: Administration cost is included in management expenses.

MR. TIBBETTS: Distributed against time?

MR. GROVER: Yes, as a part of Personnel Expenses shown in Fig. 2.

THOMAS LEE[3]: It would appear from the illustrations that you allocate all your overhead in proportion to your "direct labor." That is a principle which we in Great Britain are striving to eradicate. We maintain that only part of overhead should be allocated to "direct labor," and the remainder should be allocated on a "weight" basis. In a jobbing foundry many of the castings produced have a widely separated labor/weight ratio.

If for instance you get a run of 10-ton flywheels in one period it is likely the total weight produced in the period will exceed the average, although the labor force may be less, and therefore overhead would be under-recovered. In the next period you may be having to fill your foundry with intricate cored castings like cylinder heads, with a consequent increase in the labor force, accompanied by a diminution of the total weight produced and overhead would be over-recovered.

Thus the costs of cheap castings would be too low, and the dear castings would be too high. Further, the use of very expensive equipment in the way of electric cranes, and other lifting and handling plant, is much more closely related to the weight of the product than to the direct labor required.

MR. GROVER: I assume you refer to Fig. 7 showing the outline of a small iron foundry. The center section of this illustrative building houses molding floor No. 1 and cleaning operations performed by grinding, chipping, brushing, and sand blasting. The side sections are devoted to operations not requiring crane handling.

CHAIRMAN LEE: That would be the 5-ton flywheel.

MR. GROVER: They could handle those flywheels only in the center section with crane facilities.

CHAIRMAN LEE: The cylinder heads you were talking about are probably made on floors with crane facilities.

MR. GROVER: We hear much about the heavy foundry and light foundry. I have never been able to establish a line of demarkation.

CHAIRMAN LEE: It all depends on what you call "light."

MEMBER: If there is very little work going in your center floor, are you going to ignore your capital expense in estimating your side floors?

MR. GROVER: That brings up a very important point. If we are going to establish our selling prices on normal costs, then if we do not run any of these departments to normal capacity, we have an idle expense which the customer will not usually pay for.

CHAIRMAN LEE: Mr. Grover referred to obsolescence and depreciation through use. If your large floor is remaining idle, you are having obsolescence going on all the time even though you are not wearing out the machinery.

MEMBER: Are you going to ignore that total obsolescence?

MR. GROVER: I recommend measuring that obsolescence.

CHAIRMAN LEE: Well, suppose that you are building an entirely new building and you expect to get work that would be made on a certain floor and work made on another floor, and you spent $10,000 on the one and $50,000 on the other and you never got orders for normal operation. Would it be fair to charge full depreciation just because you were foolish enough to build that?

MEMBER: That is hypothetical.

CHAIRMAN LEE: That is a little extreme but it happens all the time.

THOMAS LEE: Even in the side floors you may have a run making uncored small flywheels on the machines, with a resulting high tonnage output, and yet only use the same labor as in the next period when you might be making thin shell castings, with a resulting low tonnage output.

It is, therefore, important, whatever the class of castings being produced, either jobbing or repetition work, to divide overhead between "direct labor" and "weight."

MR. NESSELBUSH: That goes back to the theory that you are not selling gray iron, you are selling hours of labor regardless of what you produce. If you are producing an ounce casting, you have to get your money back. If you are pro-

[1]Foran Foundry & Manufacturing Co., Flemington, N. J.
[2]Los Angeles Steel Casting Co., Los Angeles.
[3]Henry Hollingdrake & Son, Ltd., Lancashire, England.

ducing a 500-lb. casting, you still have to get your money back, and it is what labor will produce that is going to give it to you so the only thing that you are selling is the labor that is involved in making either an ounce or a 500-lb. casting.

MEMBER: You charge melting as a burden on direct labor according to your illustrations. I believe Mr. Nesselbush would rather apply that melting labor as a burden on material. Am I right?

MR. NESSELBUSH: Melting is a labor based on the average cost per pound, based on the equipment and the labor in that department.

MEMBER: Would you apply that directly to the material and not to the other labor?

MR. NESSELBUSH: No, that has nothing to do with material. It costs, let us say, 1½ cents per lb. to pay for your fuel oil, gas, coke, supplies, the investment in furnaces, the floor space you are using, and the taxes, all those amounting to 1½ cents per lb. That has nothing to do with material. That is a direct conversion of material into the molten state. Now that same expense he wants to take off in productive labor for molding, coremaking.

MR. GROVER: That is right.

MEMBER: I notice that you have direct labor in cleaning.

MR. GROVER: Any labor spent in actually cleaning a casting may be considered direct. It is certainly dangerous in a foundry where you have such terrific differences in labor cost per lb. to ignore that fact.

CHAIRMAN LEE: In one of our shops we make these flywheels and we make cylinder blocks. We may have a flywheel that weighs 100 lb. on which the cleaning cost is $1. On a cylinder block the cleaning cost is approximately $24. Now there is a lot of difference between $1 and $24.

MEMBER: I say that the time to establish the cleaning price on your casting is when you take the job the first time and to my mind, it makes no difference whether it is one order or a multiple order. When that job comes in and if you establish a grinding price based on the time taken, the next time the job comes in you know what the price is. In that way you can apply direct cleaning cost. It will take a little record keeping but it is well worth while.

CHAIRMAN LEE: That goes back to taking an average. You do not have an average foundry. You have to consider the individual plant.

E. W. OLSEN[4]: How do you arrive at the amount of overhead direct labor should carry? Your overhead can eat up the profits, can it not?

MR. GROVER: We covered that in the explanation of Fig. 3 showing variable expenses.

CHAIRMAN LEE: You generalize it. You have all of your charges which go against that particular department and as a matter of expediency, you say that your overhead is 100 per cent of your direct labor. That is the easy method to use but Mr. Grover does not advocate that.

In a given period, in the molding department, for instance, you spend $10,000 for direct molding labor and you spend $10,000 in overhead expense, then your overhead is 100 per cent of your direct labor. There again you are using averages.

MEMBER: I take it that you are interested in arriving at your normal cost. But in this discussion about production changes, from heavy castings to light castings, suppose that you had a foundry that happened to be set up without side floors and your product is such that you do not have much heavy work or wear and tear on cranes. Would you, in setting up your cost system, attempt to break something like that down to a point where you charge a variable depreciation?

MR. GROVER: I do not think it would be wise to go out and try to get business and charge the operation of a crane against some small work that does not require the crane operation in competition with the man who does not have that crane work.

MEMBER: But you have it there. My company is making castings and they do not have a great deal of selection as to whether or not they are going to make heavy castings. I was interested not so much in the selling or profit standpoint as in the standpoint of setting up the cost originally.

MR. GROVER: I do not put it on the weight of the casting. The depreciation is all upon the life of that particular equipment.

[4]Weber Engine Co., Kansas City, Mo.



Fig. 1—Overhead conveyor delivering rejected cylinder blocks to welding department for repair.

Cast iron repair welding has become an important part of foundry production operations and should be treated accordingly, every effort being made to develop its applications. Modern welding procedures, equipment, and materials have solved many problems of gray cast iron repair. The fabrication of cast iron by welding has tremendous possibilities. Cast iron weldments may be produced at a considerable saving in pattern, molding and core casts. Future developments depend largely upon cooperation of foundries in exchanging ideas and bringing problems to group discussion.

PRACTICAL APPLICATION OF CAST IRON WELD REPAIR METHODS

W. J. Pfander
Ford Motor Co.
Dearborn, Mich.

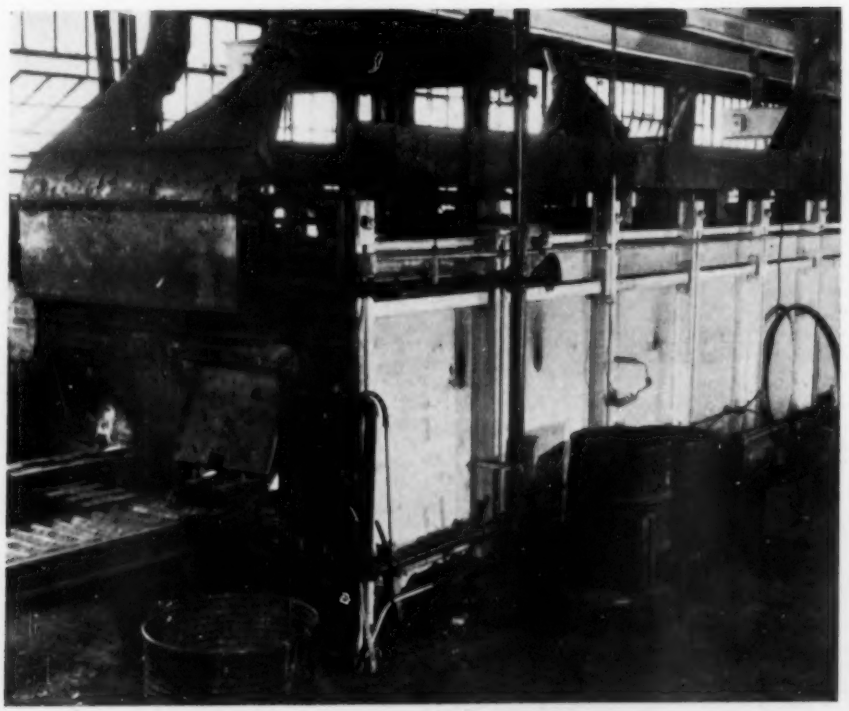


Fig. 2—View of gas-fired pusher-type furnace showing preheat exit and postheat entrance.

SOME YEARS AGO the welding of cast iron was considered impractical to the extent that its use as a repair tool was not openly admitted in the better foundry circles. Early results left much to be desired in quality and performance, and the causes of the difficulties encountered were not too generally understood.

Progress in the welding of cast iron was thus retarded and not a great deal of scientific development was achieved until relatively recent years. Today cast iron is being

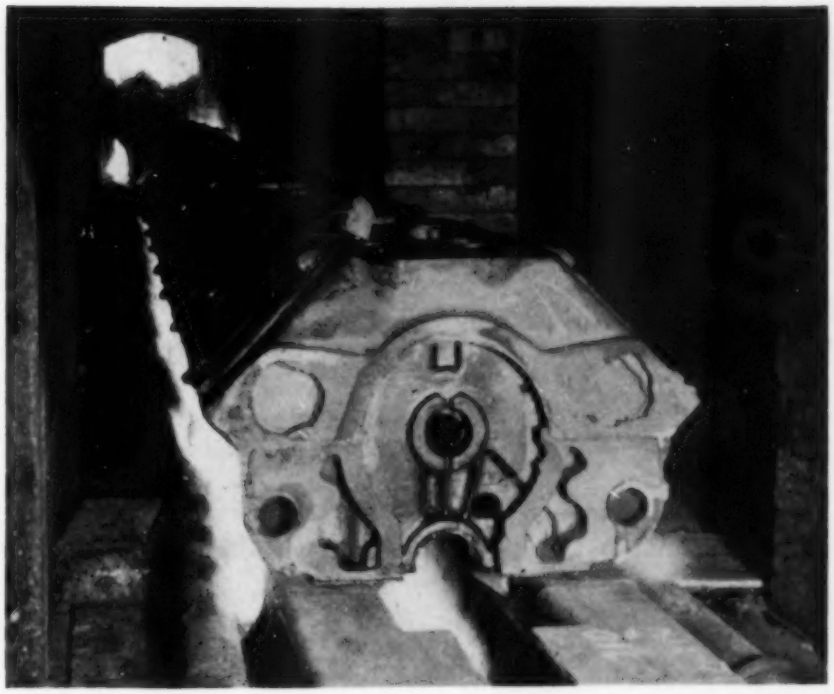


Fig. 3—Preheat passage from entrance end of furnace showing the arrangement of the burners.

welded quite generally with satisfactory results. All limitations have not been overcome but extensive progress has been made as a result of the increase in application of welding processes to cast iron.

Welding of a material possessing the chemical and physical properties of cast iron poses difficult metallurgical problems that can be solved only with the application of metallurgical knowledge and investigation. The changes in structure and differences in volume which result from heating this material above the critical temperature and subsequent cooling can be disastrous if not properly understood and controlled.

In the early days, foundries attempting this work would put a man in the darkest and dirtiest corner of the foundry or machine shop and give him the important job of reclaiming castings. Then they would wonder why the results were not satisfactory. At the company with which the author is associated this salvage work is considered an important part of production operations and, as a consequence, an attempt is made to do the job correctly and under the most favorable conditions.

In 1932, when production of "V-8" cylinder blocks was started, the losses encountered made it necessary that some means be found, in addition to acetylene welding, to salvage these castings. Up to this time little electric arc welding had been used for cast iron work because of the hardness encountered, the difficulty in getting a homogenous deposit, and non-color matching.

About this time a rod was developed consisting of about 50 per cent nickel, 30 per cent copper and 20 per cent steel, which proved to be quite satisfactory for filling small blows and defects in areas which did not require machining after welding. The deposit made by this rod could be ground flush with the casting surface and matched the color of the cast iron.

In the following years, increasing production schedules made it necessary to devise ways and means of handling larger quantities of castings and at the same time apply adequate scientific control. This was accomplished by careful consideration of all the metallurgical factors involved in the application of the different welding processes and how to best control them in production practices.

It is the purpose of this paper to demonstrate this company's present methods for making weld repairs and to disclose some of the problems encountered. Many foundrymen have had similar experiences and will find it interesting to compare their procedures with those discussed in the present paper.

Two welding processes are em-



Fig. 4—Preheat entrance of furnace showing sliding plates which lock on the rails and support block.

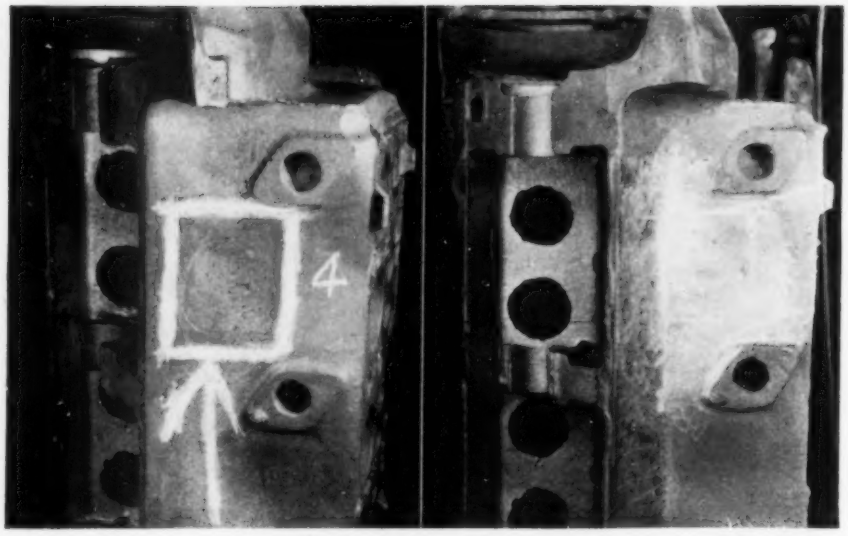


Fig. 5—Dirt defect in sidewall of cylinder block before and after repair. Furnace oxyacetylene weld.

ployed in this company's production foundry—oxyacetylene and metallic arc. Metal spray, while not a welding process, comes under the jurisdiction of the welding department and is being used in a limited amount of repair work.

The oxyacetylene welding process is the most generally used, and about 85 per cent of the welding is done by this method. It is used with two procedures, one which requires a furnace preheat and postheat and one in which local preheating is done with the torch, known as cold welding.

Acetylene Welding with Furnace Preheat and Postheat. This procedure is used on large castings where it is necessary to make repairs in areas that are restricted by the surrounding metal in their ability to contract and expand; e.g., cylinder block walls, intake port interiors, etc. Furnace preheating in repairs of this nature causes the entire casting to expand and, therefore, there is no excess concentration of stress at the area being welded.

Postheating allows the casting to cool evenly so that the highly heated areas immediately adjacent to the weld will have a chance to become almost even in temperature with the rest of the casting, thereby eliminating excessive stresses caused by uneven contraction.

Preheat temperature is held at 1100 F or just under the critical temperature, and a sufficient soaking time is allowed, depending upon the mass of the casting, to insure even distribution of temperature. Welding is started at this temperature and completed as quickly as possible; in any case, before the temperature falls below 700 F. The casting is then postheated by raising its temperature to 1050 F and slowly cooled to a maximum of 550 F before being cooled in air to room temperature.

Efficient Handling

Cylinder blocks constitute a large portion of the tonnage in this foundry and, as production may reach 6,000 blocks per day, it was found necessary to install an efficient setup for handling the percentage of these castings that must be repair welded.

Shot blasting of the cylinder blocks is done on the floor above the foundry level, and they descend on a gravity conveyor to the chipping and inspection line. Here fins and excess metal are removed by air chipping hammers and inspection made for defective blocks.

Inspectors mark for repair blocks having the following defects: surface inclusions of sand or slag, scabs, cold metal misruns, crushes, cold shuts, cracks, breaking in of gates and chipping breaks. Blocks passed by inspection move on the conveyor to the hydrostatic test stand and are tested at 60-lb. water pressure, where leaky blocks are marked for repair.

Blocks which pass the water test then enter the machine shop, while all rejected blocks are placed on an overhead conveyor which carries them to the welding department (Fig. 1). Here they are again inspected, marked and separated for furnace repair, cold welding, me-

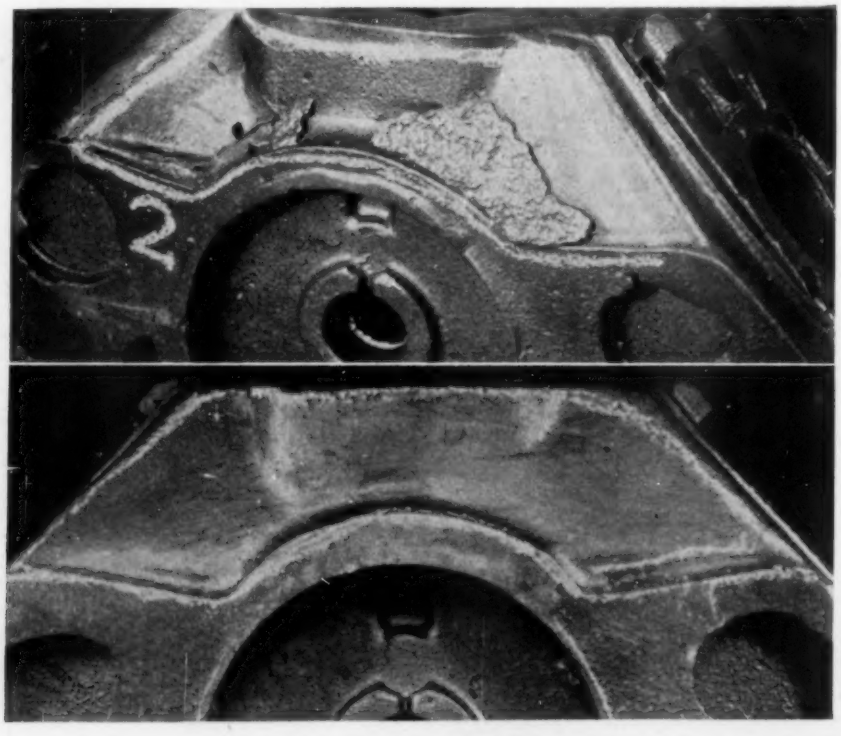


Fig. 6—Scab defect in cylinder block before and after repair. Furnace oxyacetylene weld.

tallic arc welding, or metal spray. The various repair departments are located close to the inspection area and the blocks are moved to the proper departments.

The limits as to type and size of defect it is permissible to repair are that the body or structure around the defect must be to print specification, and that no attempt be made to repair a defect which cannot be properly welded. The method of repair is governed by the location on the casting, the thickness of the metal, and other considerations consistent with good welding practice and with welding specifications which cover all the more complex castings.

Furnace repair blocks are carried down a roller-type conveyor which leads to the entrance of the preheat side of the furnace (Fig. 2). This

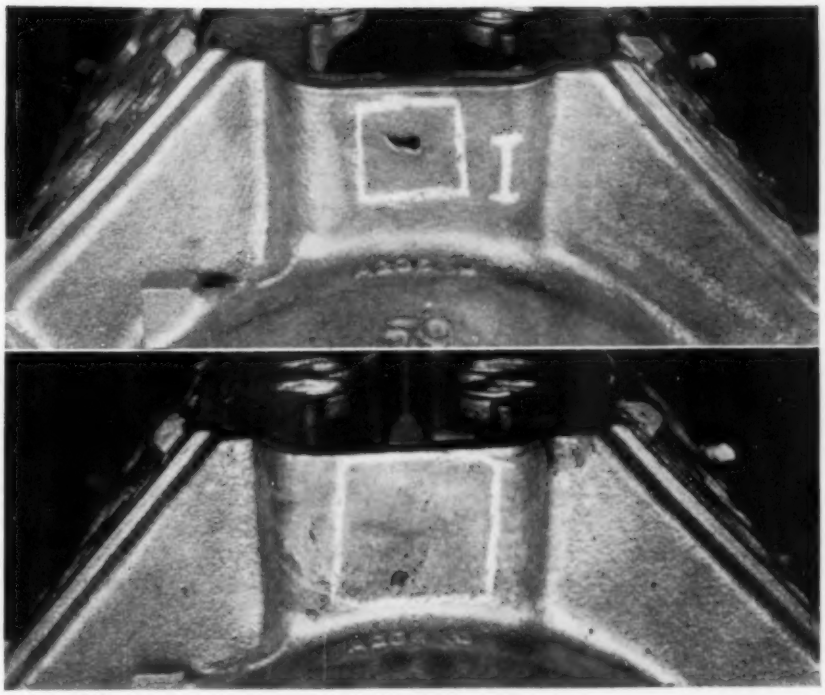


Fig. 7—Misrun in cylinder block casting before and after repair. Furnace oxyacetylene weld.

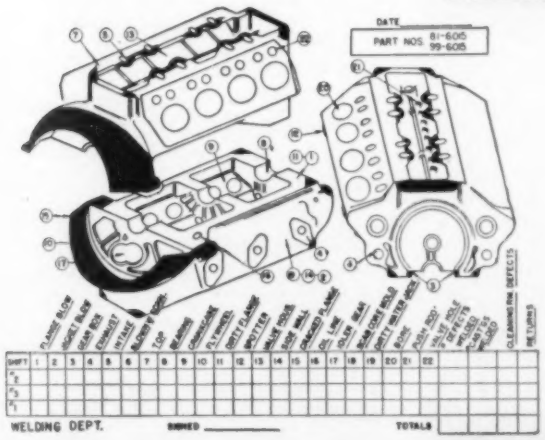
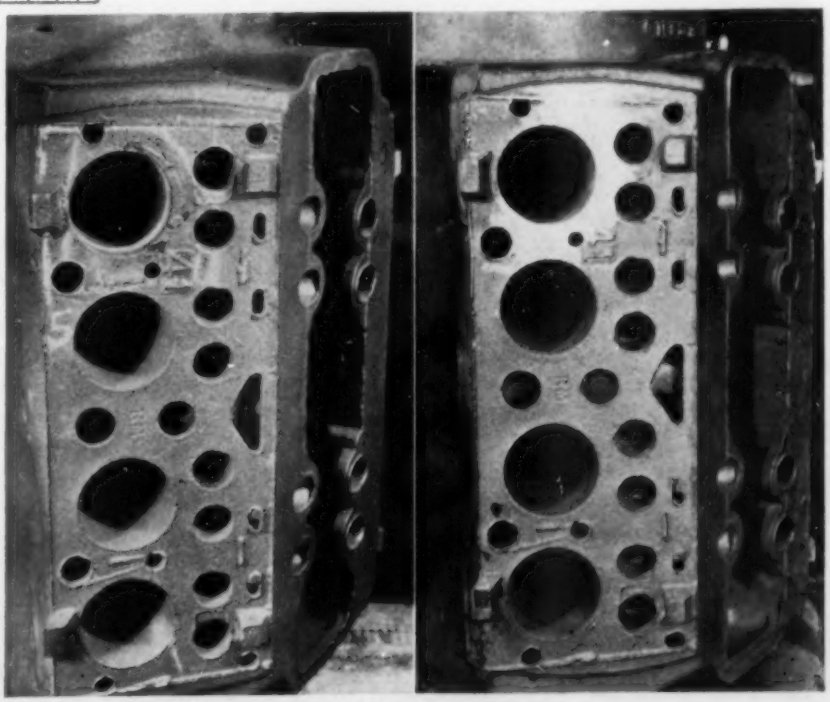


Fig. 9—Form for recording repairs. Isometric drawing of cylinder block casting. Various repair areas shown by shaded sections.

Fig. 8—Dirt defect at bore shoulder before and after repair. Furnace oxyacetylene weld.



furnace is a gas-fired pusher type using a mixture of propane and coal gas from the plant coke manufacturing operations.

It consists of two heating passages, the preheat and postheat sides which extend parallel to each other throughout the length of the furnace. Each passage is 3 ft. high, 2½ ft. wide and 40 ft. long. Burners consist of 2 in. diameter tubing with orifice holes drilled along the top of the 18-ft. lengths.

Six of these burner tubes (18-ft. lengths) are mounted horizontally and lenghtwise in the preheat side —three at the entrance end and three at the exit end. They are mounted just below the level of the rails on which the blocks slide, one between the rails and one on each side. In addition to these tube-type burners, there are two inspira-

tor-type burners mounted in the top of the furnace at the exit end of the preheat side.

The postheat side has two inspirator type burners and three burner tubes (14-ft. lengths) at the entrance end only (Fig. 3). The gas mixture is supplied to the burner tubes without previous mixture of air, and combustion depends upon the air present in the furnace. This is controlled by means of draft doors running the length of the furnace on both the preheat and postheat sides.

Regulation of these draft doors provides temperature control at various points in the furnace. The flame produced by the tube-type burner, while it plays directly on the block, is on the reducing side and has a tendency to spread as it leaves the burner orifice, thus producing a uniform heat distribution throughout the block.

Direct Heating Satisfactory

Experience with other types of furnaces in which the heat is supplied indirectly has proved the direct heating arrangement to be most satisfactory, especially in the postheat where the welding area is at a much higher temperature than the remainder of the casting and it is necessary to equalize the temperatures as quickly as possible to prevent cracking.

Sliding plates or shoes which lock on the rails are placed in position and the block is placed on top of them (Fig. 4). The block is then pushed into the furnace by a hydraulic ram. This automatically ejects a block at the exit end at full preheat temperature of 1050 F and ready for welding. Length of the passage will accommodate nineteen "V-8" blocks, but it is not necessary to have the furnace full to eject a block at the exit end as the shoes take the pushing load.

If a block is ready for exit and another is not available to go into the entrance end, then the required number of shoes is placed on the rails and pushed into the furnace. The ejected block is immediately picked up by electric hoist and placed on the welder's table.

Asbestos boards are placed in position around the block to prevent rapid or uneven cooling and the welding operation begun immediately. Preparation is done with the oxyacetylene torch using a neutral flame. Complete dirt and slag removal is assured as the operators have developed a great deal of skill in judging when the area is clean by observing the color of the flame.

A No. 20 welding torch and a No. 7 or 8 copper tip with the flame adjusted to neutral is used for welding. The torch is applied to a generous area around the defect to further minimize any abrupt temperature differential in the block and the defective area is brought up to welding heat.

The rod is heated, dipped in flux and applied to the desired spot. The rod is maintained at a higher temperature than the base material by concentrating the flame on it and allowing only enough heat to strike the base metal to keep it at welding temperature. This procedure helps to prevent hard spots which might be formed by allowing the cold rod to more or less quench the area.

The rod is held with its molten end in contact with the base metal, and as the rod flows into the weld it is allowed to build up and puddle until a sufficient amount has been deposited. The torch is then used to smooth the weld deposit and heat the general area before the block is returned to the furnace (Figs. 5, 6, 7 and 8).

Rod Specifications

Welds are made as quickly as possible to minimize the burning out of carbon and silicon and to return the block to the furnace before it cools to 700 F. The 5/16-in. diameter rod is cast in the company foundry to the specification:

Element	*Per Cent*
Carbon	3.35–3.55
Silicon	2.60–3.00
Phosphorus	0.40–0.60
Manganese	0.60
Sulphur	0.10 max.

This rod gives a relatively soft deposit ranging from 150 to 225 Vickers Brinell. The flux used is made in the company plant to the following composition:

Component	*Per Cent*
Boric acid	41½
Soda ash	41½
Ammonia sulphate	2
Powdered iron	15

Upon completion of the weld, the block is returned to the furnace and pushed through by the same procedure used on the preheat side. After exit from the furnace, it is ground and placed on a conveyor which carries it to a water test set up in the welding department, and by the time it arrives there it is almost room temperature.

After water test the block must go through all phases of inspection before being released to the foundry machine shop. This procedure is followed on all welded blocks no matter what type of repair weld has been made.

Repair Areas Recorded

Records of the location and frequency of repair areas are kept by the use of forms having an isometric drawing of the cylinder blocks with the various repair areas shaded and listed at the bottom of the sheet (Fig. 9). Compilation of these forms immediately shows the frequency of repair for the various areas and whether corrective measures are necessary in the foundry.

Cold Welding. This type of repair is applied sparingly to rough and rough-machined castings where there is no danger of affecting critical areas. It is confined to such places where expansion and contraction are not restricted; e.g., rough-machined surfaces which fail to clean up are welded by this method, while defects through water jackets are welded only with a furnace preheat.

Local preheating with the torch is applied in this type of welding and care is exercised to insure an even graduation of heat from the defect to the coolest sections of the casting. This is done by rotating the torch around the defect with decreasing radii of motion. The welding procedure after preheat is the same as is followed in furnace welding with the exception that the casting is allowed to cool in air instead of going through the furnace postheat (Figs. 10 and 11).

Minor Defect Repairs

Metallic Arc Welding. Metallic arc welding is used mainly on semi-finished blocks. Low areas, to a maximum of 1½-sq. in., which fail to clean up during machining, and small defects that do not go completely through the wall thickness, as well as cracks and broken locating lugs, are repaired by this procedure. All defects are prepared by grinding, drilling or chipping, with the excep-

tion of minor welds made for appearance only.

Welding transformers are company built with a capacity of 60-250 amperes, 24 volts alternating current. Alternating current is preferred because of its smoother operation at low current values and its less penetrating arc.

Four types of welding electrodes are employed:

1. A nickel wire which is specified to a minimum of 99 per cent nickel and used exclusively on the inside diameter of machined cylinder bores.

2. A heavy flux-coated copper-nickel alloy rod, containing about 70 per cent nickel and 30 per cent copper, which is used as a filler on low areas and locations which must subsequently be machined.

3. A nickel-core electrode composed of a central core of commercially pure nickel (99 per cent) wrapped in a commercially pure copper sheath which is, in turn, wrapped in a low carbon steel sheath, and the whole flux coated. This electrode is used for miscellaneous welding where subsequent machining is not done.

4. A copper electrode composed of a commercially pure copper core wrapped in a low carbon steel sheath and flux coated. This electrode is used for repair of water jacket leaks in conjunction with the metal spray.

Nickel Electrode. Cylinder bores in which small defects are uncovered by the machining operations are repaired in a special welding booth set up in the foundry machine shop. Arc welding in the bore is limited to the following conditions:

a. Top inch of bore—no defect greater than ⅛-in. diameter can be welded.

b. Remainder of bore—no defect greater than ¼-in. diameter can be welded.

c. No defect extending through more than one-half the machined thickness of the bore wall is welded.

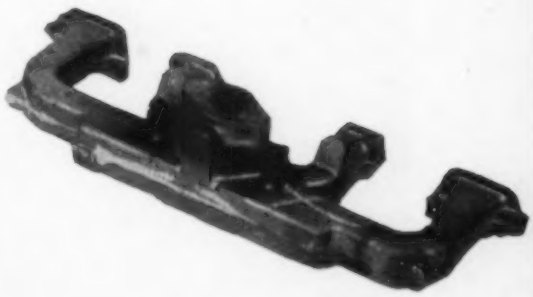
These requirements are well within the limits of the Engineering Bulletin No. 152, Revision 1, dated Sept. 12, 1944, issued by Office, Chief of Ordnance, Detroit. This bulletin was compiled through the joint efforts of the SAE War Engineering Board and the Ordnance Department.

Defects to be welded are drilled out to investigate the extent of the defect and prepare for the welding. The drilling is accomplished with a right angle hand drill adapted for this operation (Fig. 12).

The area is preheated with a gas torch to about 150 F and the welding done with the nickel electrode of ⅛-in. diameter at about 80 amperes, 24 volts. The completed weld is carefully ground with a hand grinder until it is about flush with the surface, then finished flush with a hand lapping tool and finally polished on a production lapping machine.

Copper Nickel Electrode. Cylinder blocks which require building up in areas which must be machined, e.g., broken bosses, are ground when necessary and preheated with a torch to about 600 F. The preheating is carefully done to insure an even distribution of heat and the weld made with a 5/32-in. diameter Cu-Ni electrode.

This procedure gives a weld hav-



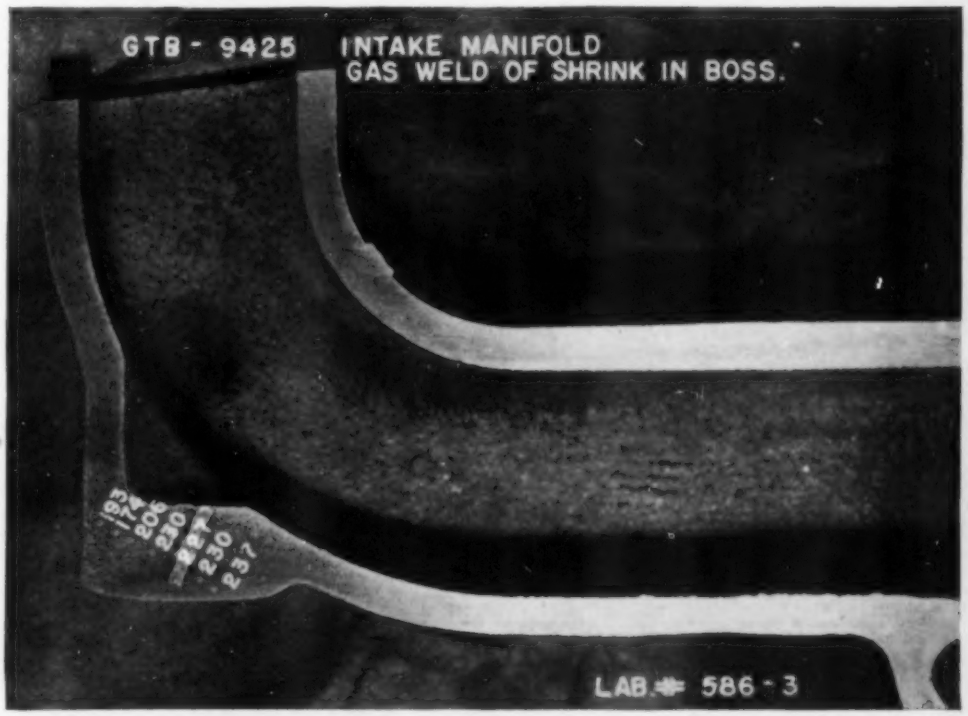




Fig. 10—(top and bottom) Repair of shrink on boss of intake manifold. Oxyacetylene cold weld.

Fig. 11—Macrosection of boss repair weld shown in Fig. 10.

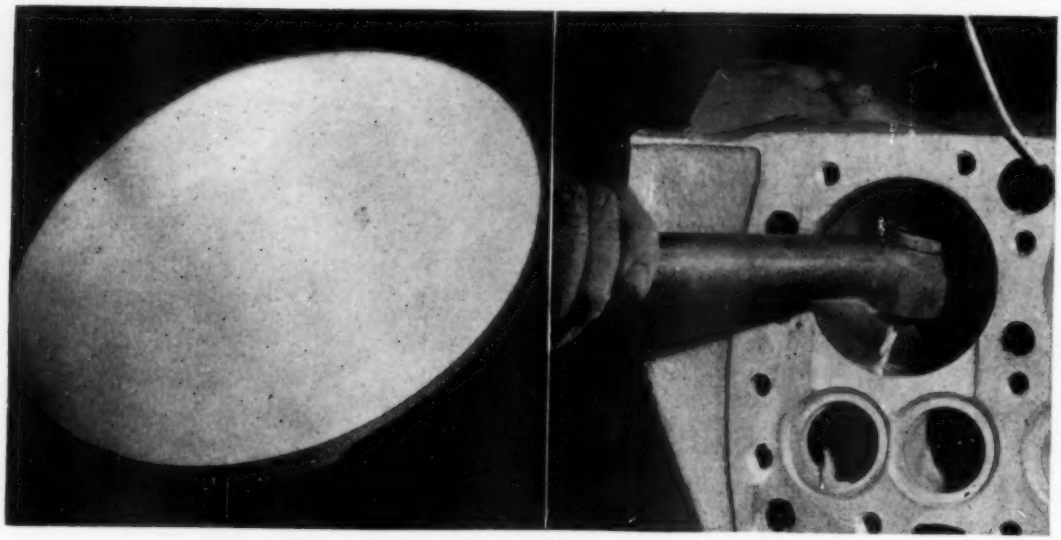


Fig. 12—Drilling out bore defect in cylinder block casting with specially adapted angle hand drill.



Fig. 13—Small slag or dirt hole in water pump housing arc welded with nickel-core electrode.

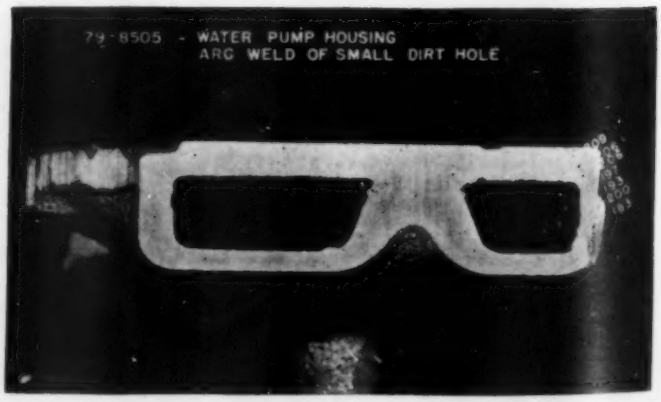


Fig. 14—Photograph of macrosection of water pump housing repair weld shown in Fig. 13.

ing no hard fusion zone and easily drilled and tapped. The deposit from this electrode will be soft and machineable when made without preheating, but the fusion line might be somewhat hard. In making large repairs and where machineability is quite critical it is better to preheat to at least 500 or 600 F to counteract expansion and contraction stresses and soften the fusion line (Figs. 13 and 14).

Preheating Small Defects

Small defects uncovered during machining are preheated to about 200 F and welded with the copper-nickel electrode. When necessary, they are drilled with a ¼-in. drill to remove dirt, and the deposit in all cases is ground or filed flush before machining.

Due to rough handling, flywheel housings which are cast to the "V-8" block will occasionally be cracked. In this case the crack runs to a free edge which must be machined. The closed end is drilled beyond the end of the crack with a ¼-in. drill and the crack ground to form a "V" trench for welding. The area is preheated to about 500 F and the nickel-core electrode used to fill the "V" up to the free edge. The free edge is welded with the copper-nickel electrode to secure a machineable deposit (Fig. 15).

Nickel-Core Electrode. The nickel-core electrode is used on surfaces which do not require machining, and is used without preheat where the defects are small and on relatively heavy sections. If the defect is large and the mass of weld deposit required might cause considerable thermal stress, then the area is preheated.

If drilling is required to remove dirt, a slight preheat generally is used. All deposits made with this electrode are ground or filed flush with the surface.

Metal Spray Procedure

Copper-Core Electrode. The copper-core electrode is used in the metal spray procedure along with the nickel-core electrode. This procedure is confined to those areas which are not machined and can be satisfactorily shot blasted, such as water jacket walls.

Seepers or leaks which would require mechanical plugging are repaired by this method. Its advantage lies in the fact that it covers a much greater area than that covered by plugging. A leaking type of defect usually is associated with spongy metal and plugging does not assure that new leaks will not develop in the adjacent area.

Seeping locations are either ground or drilled with a 7/32-in. diameter drill until a clean cavity is assured (Fig. 16). A gas torch is employed for preheating to about 150 F and an ⅛-in. diameter copper-core electrode is used to make a deposit in the bottom of the preparation. This electrode gives a dense, defect-free, water-tight deposit, effectively sealing openings.

An ⅛-in. diameter nickel-core electrode is then deposited to fill the preparation and give a color-matching repair. The deposit is then ground flush with the surface and the whole area is shotblasted (Fig. 17). The sprayed metal requires a rough surface for adherence as it bonds mechanically to the base metal.

Size G-16 grit is used in the blasting operation and provides a satisfactory surface with sharp, angular indentations into the base metal. A generous area around the defect is blasted to insure good coverage.

Spraying Two Metals

It is necessary to spray two metals on the area. The first is of commercially pure 0.090-in. diameter copper wire sprayed to a thickness of approximately 0.010-in. This is followed by a coat of 0.090-in. diameter low carbon steel wire sprayed on the copper to about 0.005-in. thickness. The copper provides a good bond to the base metal and the steel gives an acceptable color match to the casting (Figs. 18 and 19).

Water pressure testing, after repair is completed, seldom results in rejection of the casting.

All of the aforementioned welding procedures, with the exception of the metal spray, are applicable to any of the gray iron castings produced in this foundry. Restrictions on welding are determined by the casting design and the use for which it is intended; e.g., cylinder heads due to their flat shape and stressed condition may not be welded beyond 1½-in. from each end unless they are furnace preheated to temperature of 1100 F.

Preheating is done in an open furnace in which the castings are placed on racks located above an open gas flame. This type of casting is easily warped with a localized welding heat. Combustion chambers in cylinder heads are not welded as the operating conditions in this area are severe.

Severe Stress Conditions

Operating conditions of the part under consideration for weld repair should be well investigated before any welding is attempted. If these conditions are so severe that any variation in hardness or structure is liable to result in failure, then welding is prohibited.

If it is considered necessary to weld in an area of severe operating stress, then breakdown tests should be made to determine the effect on part life. Welding in this application is best done using the oxy-acetylene method with properly applied preheat and postheat.

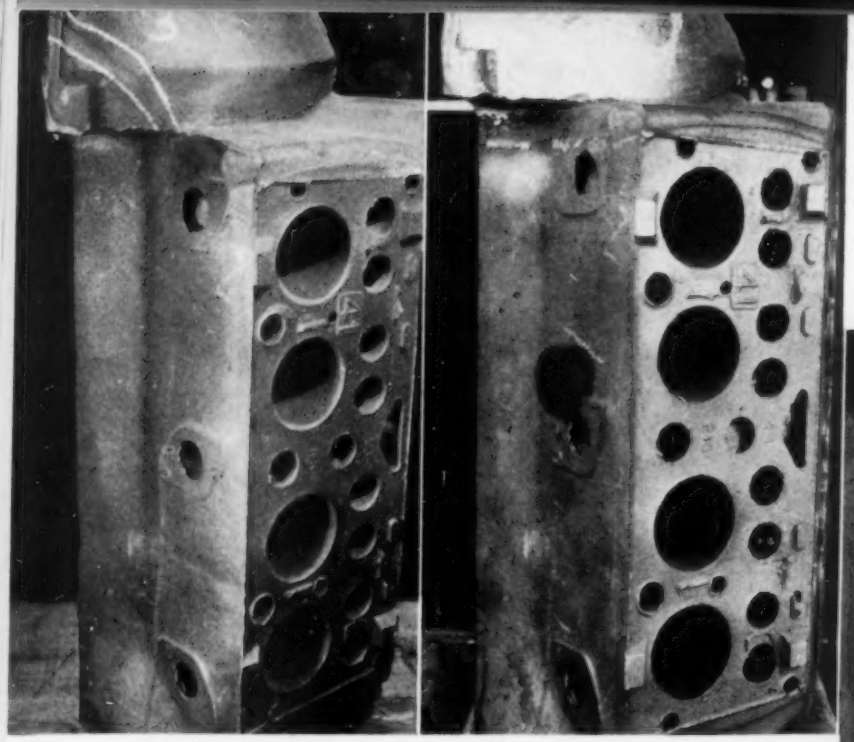
Extent and location of the defect are also factors which determine the welding procedure to be used. Generally, small defects of ¼-in. diameter in areas not requiring machining can be metallic arc welded without preheat. A slightly larger defect might require preheating to offset thermal stresses. A large defect in the same area might require a high preheat and postheat as well.

Areas requiring machining should be gas welded or metallic welded with the copper-nickel alloy rod. Extent of the defect and accessibility determine the method.

Economy Factor

In dealing with small castings a check on welding costs will sometimes reveal that it is more economical to scrap the pieces. Economy is a factor which must be closely watched in a large foundry. It is convenient to send castings for weld repair when a rather large department is available for this operation. Therefore, it is necessary for the control department to keep a close check on all castings sent for weld repair and consider them from the economy standpoint.

Today all foundries large and small must recognize that welding



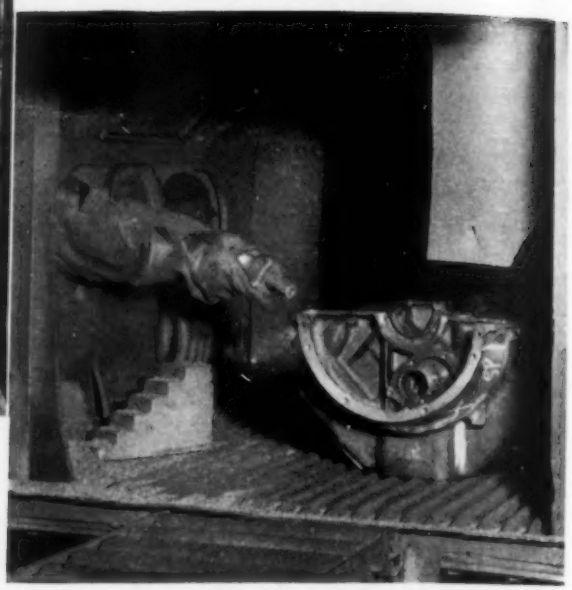




Fig. 15—(left above) Cracked flywheel housing arc welded. Nickel-core and copper-nickel electrode.

Fig. 16—(left) Grinding seepage area in cylinder block casting in preparation for metal spray.

Fig. 17—(above) Gritblasting the area shown in Figs. 15 and 16 to provide rough surface for spray.

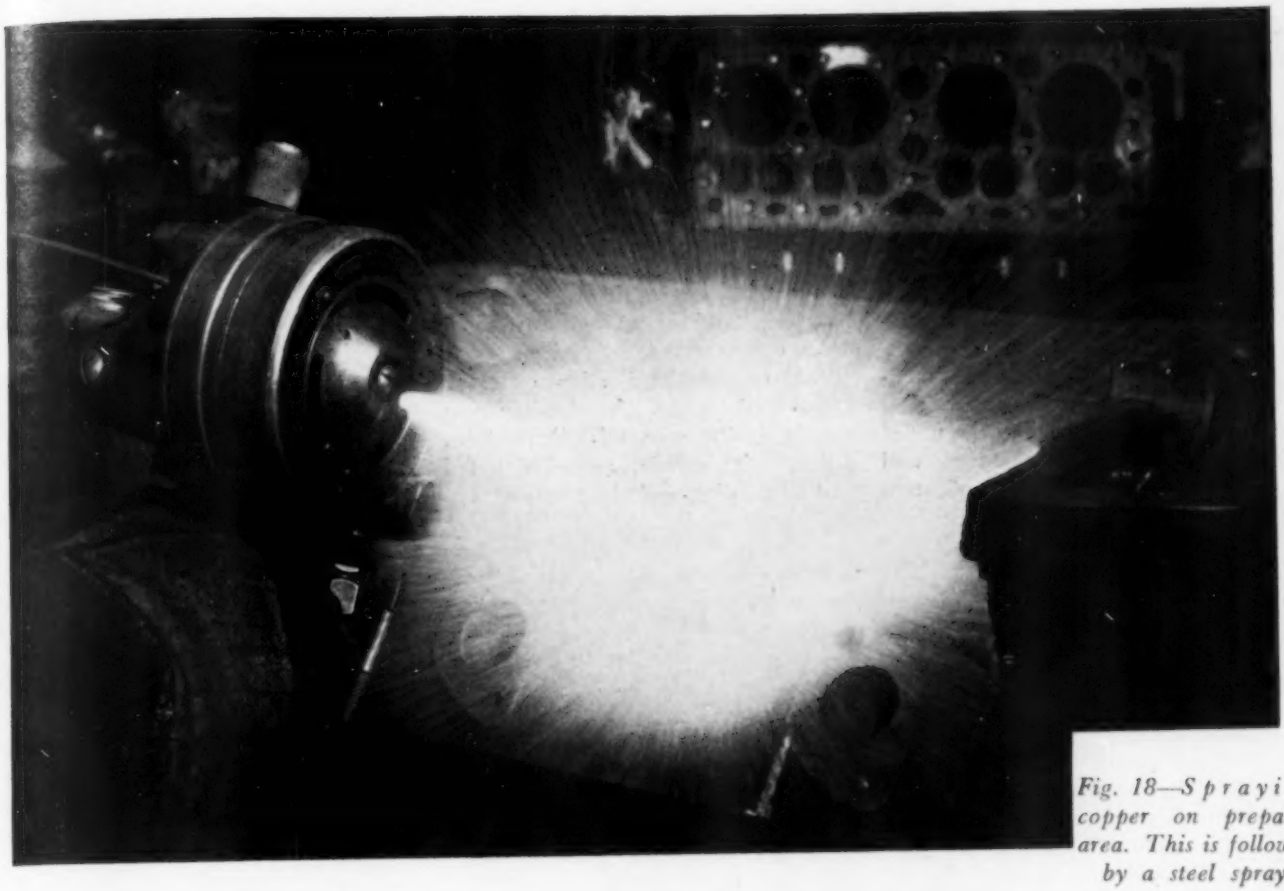


Fig. 18—Spraying copper on prepared area. This is followed by a steel spray.

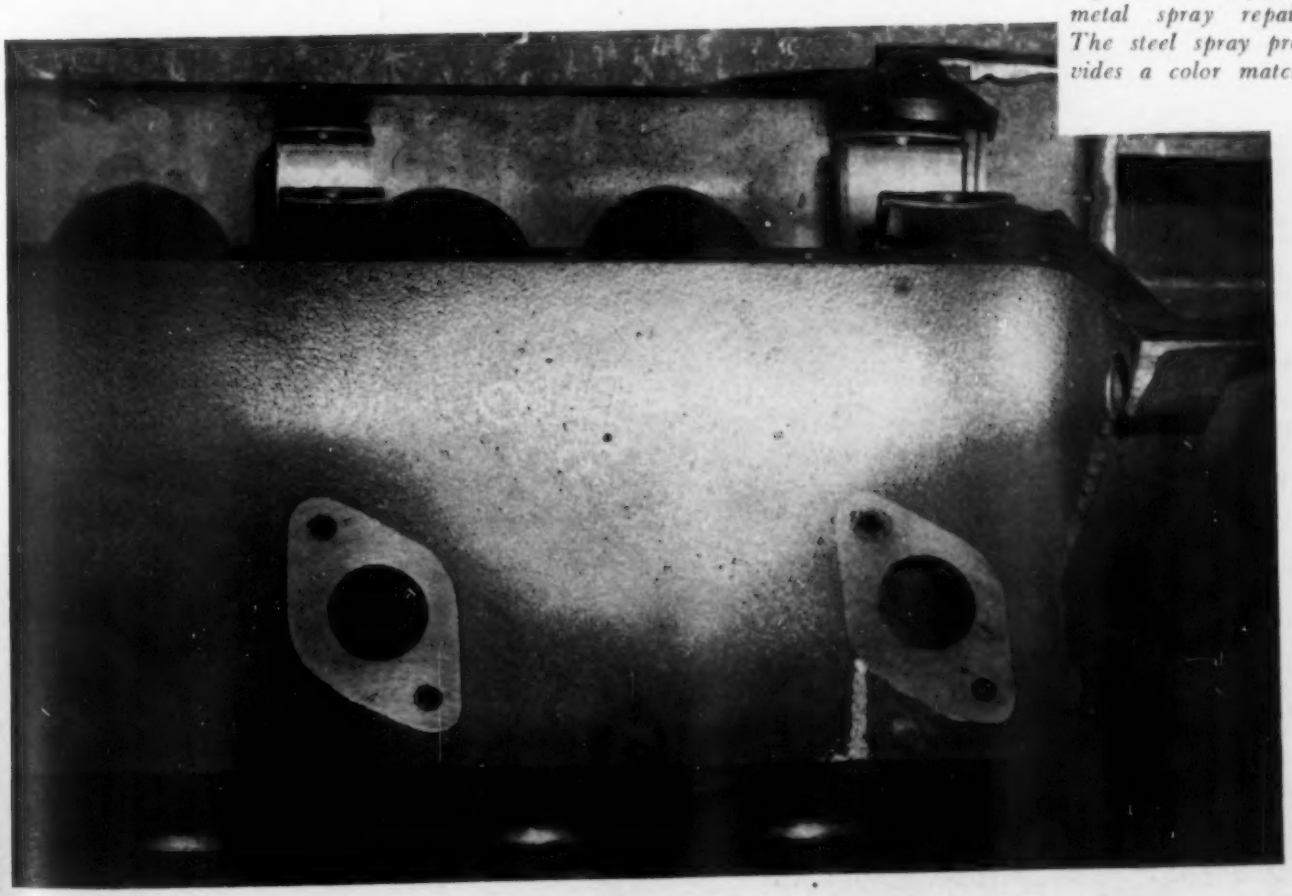


Fig. 19—Completed metal spray repair. The steel spray provides a color match.

has of necessity become an important part of production operations. It should be treated as such and every effort made to improve and develop its application.

Economy of the operation is largely determined by the investment in proper equipment, materials and personnel. If it is left to more or less shift for itself under a cheap and inefficient setup, then the results obtained will be correspondingly cheap and inefficient. Practically any gray cast iron welding problem can be solved by the use of present day welding materials and equipment and proper application of the various methods.

An indication of the progress of gray iron welding is illustrated by its use today in fabricating gray iron parts. A certain foundry making stove castings found that a particular casting was exceedingly expensive to make due to its shape and core requirements. The solution was found by casting in two separate sections and metallic arc welding them together.

A copper-nickel alloy electrode was used for the job, with one man welding 480 assemblies in an 8-hr. day. All assemblies are water tested after welding and the one per cent of leakers found are easily repaired.

DISCUSSION

Chairman: E. C. JETER, Ford Motor Co., Dearborn, Mich.

Co-Chairman: JOHN CROWE, Air Reduction Sales Co., New York.

CO-CHAIRMAN CROWE: I would like to ask what method was used in removing those defects? Was it done with a torch?

MR. PFANDER: The welding torch with the flame adjusted to neutral, was used to remove these defects. The adjacent area was preheated to reduce strain and the defective material melted out. The operator was able to judge when a sufficient amount of defective material had been removed by observation of the cavity and the yellow color of the flame as it burned out the included dirt.

GRAY CAST IRON SECTION SIZE vs. PHYSICAL PROPERTIES

V. A. Crosby
Metallurgical Engineer
Climax Molybdenum Co.
Detroit

CAST IRON is a complex iron-base alloy of much versatility. Due principally to its high degree of sensitivity to variations in alloy additions and cooling rates, its properties can be modified to meet a wide variety of uses. Within these ranges (alloy additions and cooling rates) occurs practically every reaction known in physical ferrous metallurgy. These reactions account for the wide range in physical properties possible of attainment, such as modulus of elasticity of 4 million to 26 million; tensile strength of 15,000 psi to 80,000 psi; hardness values of 100 to 600 BHN.

While all metals and alloys solidify as crystals that are influenced by rates of cooling, both as to chemical and physical heterogeneity, gray cast iron, because of the formation of graphitic carbon, exhibits this heterogeneity to the greatest extent. This apparent disadvantage in the realm of cast iron is actually an advantage over all the other members of the family of metals and alloys, for the presence of graphite reduces volume shrinkage to a degree not present in other ferrous metals or alloys. Graphite also improves the machining characteristics, damping capacity, and lubrication and antifriction properties.

The extent of graphitization, grain size, structure of the matrix and the degree of soundness are factors which determine the strength, hardness and other engineering properties. The properties of cast iron are so related to the section in which the iron is cast that one should not discuss its properties without reference to the section under consideration. Section sizes or wall thicknesses are usually considered indices of rates of cooling; but this may be inaccurate because of the influence of gate and riser locations on castings. Actually the cooling time of a given section will depend upon a number of conditions, to mention only a few:

1. Initial temperature of the iron.
2. Conductivity of the mold material.
3. Metal surface area for dissipation of the heat.

However, these are matters largely for the foundryman's consideration and usually of little concern to the designing engineer or to the user of castings.

The engineer is concerned primarily with the properties of the part in the finished condition, and he must so state his requirements that the desired properties and quality are secured. Furthermore, he must so choose the material that machining and processing from the rough casting to the finished part is feasible and that quality production is maintained without penalty.

Since, as stated above, cast iron is sensitive to the cooling rate, which in general varies with section size so that the same composition produces widely different properties in heavy and light sections, the engineer finds that the specification of chemical composition as a means of securing the properties desired is be-

Table 1

VARIATION IN STRENGTH WITH SECTION SIZE (ROTHER AND MAZURIE[1])

Tensile Strength, 1-in. Diameter Bar, As-Cast, psi.	*Percentage of Tensile Strength of 1.0-in. Diameter Bar*			
	1½-in. Diameter Bar	*2-in. Diameter Bar*	*2½-in. Diameter Bar*	*3-in. Diameter Bar*
21,300	77	72	71	67
23,800	81	69	61	53
25,900	89	74	55	46
Avg. 23,700 (soft iron)	82	72	62	55
32,800	94	84	75	73
36,300	85	83	76	76
Avg. 34,600 (machinery iron)	90	84	76	75

• Properties of cast iron are so related to the section in which the iron is cast that one should not discuss its properties without reference to the section under consideration. This paper presents a review of the studies of several investigators on the effect of mass vs. properties of gray cast iron.

Table 2

VARIATION IN STRENGTH WITH SECTION SIZE (BORNSTEIN[2])

Tensile Strength, 1.2-in. Diameter Bar, As-Cast, psi.	Percentage of Tensile Strength of 1.2-in. Diameter Bar: ¾-in. Diameter Bar	2-in. Diameter Bar	3-in. Diameter Bar
23,600	122	65	51
25,400	107	71	57
Avg. 24,500 (soft iron)	115	68	54
35,200	104	69	53
36,000	104	77	63
36,900	101	84	66
Avg. 36,000 (machinery iron)	103	77	61

Table 3

VARIATION IN STRENGTH WITH SECTION SIZE (MACPHERRAN[3])

Tensile Strength, 1.25-in. Diameter Bar, As-Cast, psi.	Percentage of Tensile Strength of 1.25-in. Diameter Bar: 13/16-in. Diameter Bar	1¾-in. Diameter Bar	2¼-in. Diameter Bar	2¾-in. Diameter Bar	3¼-in. Diameter Bar	4-in. Diameter Bar
21,100	134	80	64	—	—	—
22,000	—	77	64	66	68	64
31,000	—	86	74	71	63	61
34,600	122	81	71	—	—	—
54,000	—	94	89	87	80	71

Table 4

VARIATION IN STRENGTH WITH SECTION SIZE OF SIX CLASSES OF IRONS

Class	Tensile Strength, per cent*: ¾-in. Diameter Bar	1 1-in. Diameter Bar	1.6-in. Diameter Bar	2.0-in. Diameter Bar
20	100	79	60	53
25	100	81	73	72
30	100	78	71	67
35	100	83	68	66
40	100	85	80	67
50	100	90	79	72

*Private communication from Gray Iron Institute (Now Gray Iron Society).

Table 5

COMPARATIVE TENSILE VALUES IN PSI.

Heat No.	Tensile Strength, per cent: 1.2-in. Diameter	2.0-in. Diameter	Difference
1	83,900	60,400	−23,500
2	82,700	64,000	−18,700
58	82,280	76,550	− 5,730
3	75,550	65,400	−10,150
68	74,750	72,600	− 2,150
64	72,600	59,400	−13,200
62	71,280	55,880	−15,400
67	67,000*	69,550*	+ 2,550
35	58,900	53,450	− 5,450
32	55,500	49,600	− 5,900
65	52,450*	59,450*	+ 7,000
33	48,700	46,800	− 1,900
63	48,400	43,780	− 4,420
31	45,600	43,600	− 2,000
34	40,275	40,000	− 275

*Exceptions to the general rule.

Table 6

EFFECT OF MOLYBDENUM ON SECTION SIZE

No.	Section Size, in.: 0.8	1.2	1.6	2.2	4.8	Molybdenum, per cent
A Brinell Hardness	178	171	166	169	159	0
B Brinell Hardness	185	185	185	185	180	0.35
C Brinell Hardness	193	193	190	193	190	0.64

Fig. 1—Measurement of Brinell hardness for different thicknesses of soft gray iron.

Fig. 2—Measurement of Brinell hardness for different thicknesses of high-test gray iron.

Fig. 3—Curves showing relationship between mass and tensile properties in unalloyed gray iron in sections between 1.2 to 6 in.

Fig. 4—Effect of chromium on hardness vs. section size in irons cast into telescope castings of analyses and properties shown in Table 8.

Fig. 5—Effect of increasing nickel and section size on physical properties of cast irons of analyses shown below.

Element	No. 7 Per cent	No. 10 Per cent	No. 11
T.C.	3.26	3.36	3.35
Si	2.11	1.64	1.08
Mn	0.60	0.60	0.57
P	0.20	0.23	0.20
S	0.115	0.130	0.140
Ni	1.03	2.17	3.02

Fig. 6—Relation between tensile strength and Brinell hardness in unalloyed gray cast iron in 6-in. casting section.

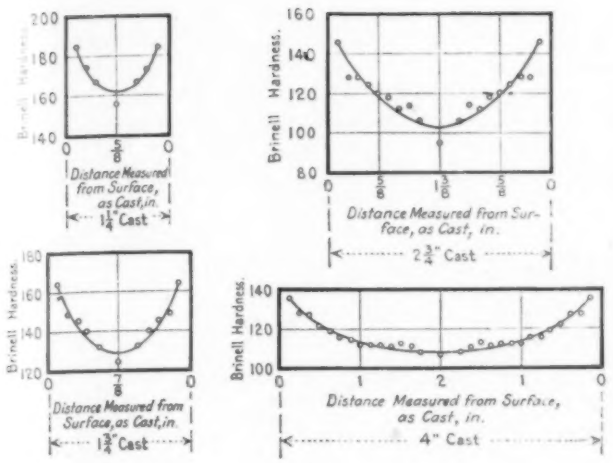


Fig. 1

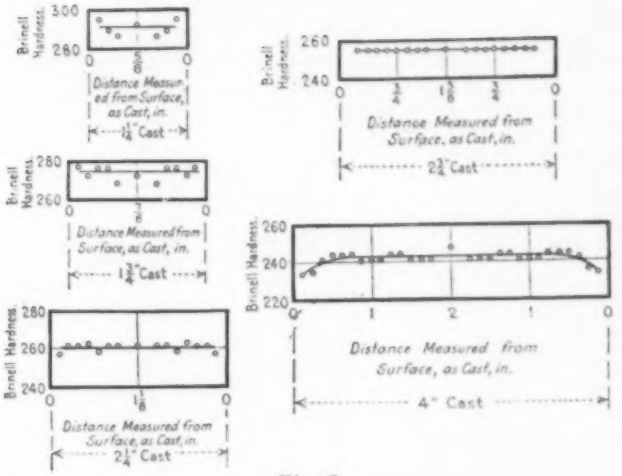


Fig. 2

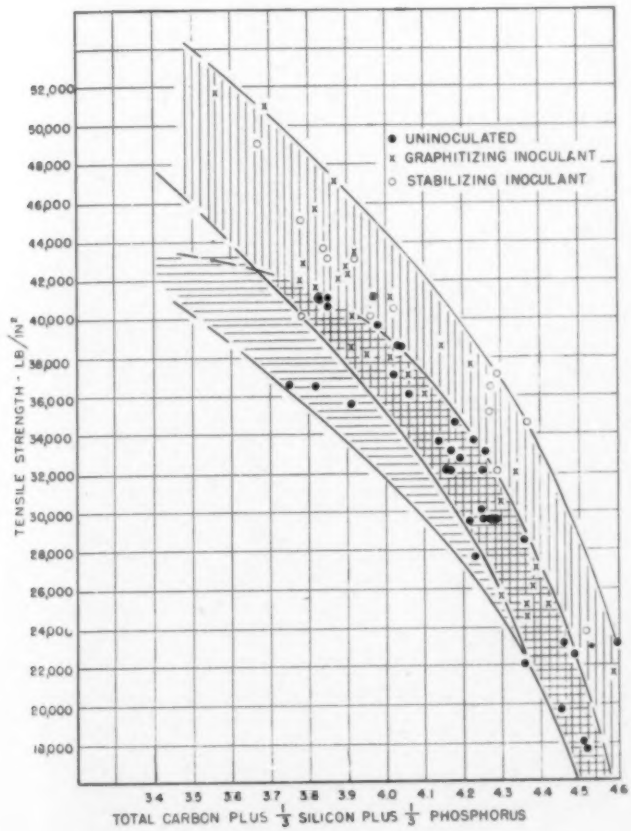


Fig. 3

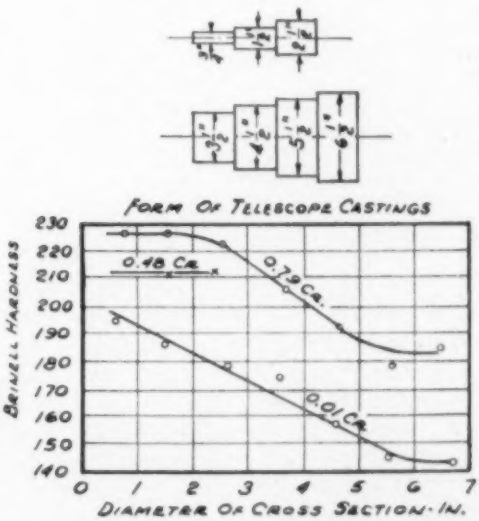


Fig. 4

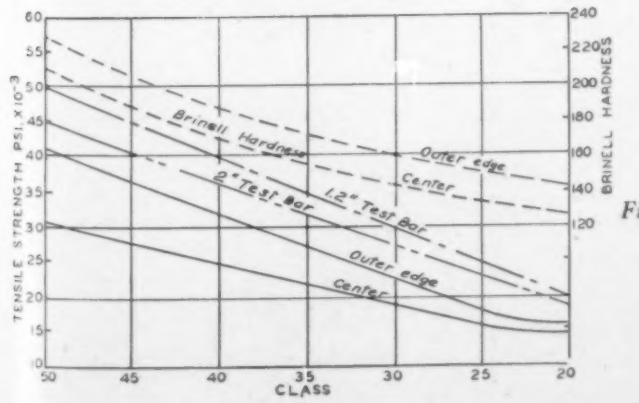


Fig. 6

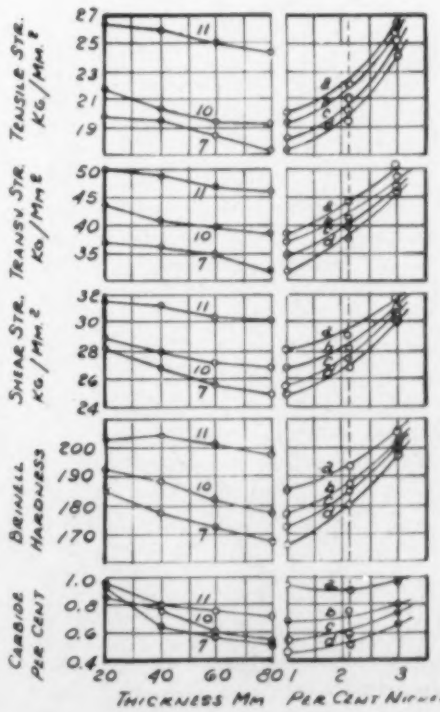


Fig. 5

coming outmoded. This is not to imply that chemical composition has little or no value; but rather to insist that it is the foundryman's tool for securing the properties desired by the engineer. Hence, it is necessary for the engineer to evaluate and specify cast iron by its mechanical properties as tested on specimens representative of the casting.

Test Bar Function

The separately cast test bar (ASTM Spec. A-48) evaluates the quality of the molten metal ready to cast. It is therefore useful to the producer as a method of control of the molten-metal quality and of eliminating "scatter" from that cause; but it does not evaluate the casting. The engineer's designing of the part to make it sound may have more to do with the strength and actual serviceability of the casting as a unit part than is indicated by the test bar physical properties.

Many investigators have studied the effect of mass vs. properties and the findings of their work are shown in the following tables. The work of Rother and Mazurie[1] is well summarized and appears in Table 1.

Bornstein[2] has also reported some carefully conducted tests showing the percentage drop in tensile strength with increase in section from ¾ to 3 in. These data are presented in Table 2.

MacPherran's[3] data on five grades of iron, ranging from 21,000 to 54,000 psi. in sections 13/16th to 4 in. indicate less drop in per cent with section size in the higher strength irons than in the lower tensile irons. These data are shown in Table 3.

The Gray Iron Founders' Society[4] has permitted use of data in the Cast Metal Handbook which confirms MacPherran's report on sections ¾ths to 2 in. using six grades of irons from 20,000 to 50,000 psi. as shown in Table 4.

MacPherran[5] has shown some interesting hardness exploration in soft irons and in high-strength irons, using sections 1¼ to 4 in. in thickness. The curves are shown in Figs. 1 and 2.

W. A. Hambley[6] has shown some curves illustrating the relationship between mass and tensile properties using unalloyed iron. These curves for sections 1.2 to 6 in. in thickness are shown in Fig. 3.

A correlation of tensile properties in 1.2-in. and 2-in. test bars is shown below in Table 5.[7] The departure in some cases from the general assumption that the heavier sections should show lower strength is accounted for metallurgically on the basis of the presence of carbide (hard spots) in the smaller sections where the cooling rate is more rapid. The foundryman's method of overcoming this difficulty is to increase the silicon content.

The effect of alloying elements on section sensitivity varies somewhat with the particular alloy used. In the case of chromium[8], it will be noted from Fig. 4 for section ¾ to 6½ in. that "chromium exerts no influence on the sensitivity to wall thickness."

The effect of increasing nickel and reducing silicon on the basis of two nickel to one silicon with increasing section size is shown by Piwowarsky[9] in Fig. 5.

The effect of molybdenum on hardness in sections 0.8 to 4.8 in. is shown in Table 6.[10]

By private communication from the Vanadium Corporation[11], it is reported that "Vanadium . . . makes for greater uniformity of strength and hardness properties in larger sections." See Table 7.

Conclusion

Since castings may be white or gray, depending upon the cooling rate; or, with cooling rate constant, castings may be white or gray, depending upon chemical composition —it is little wonder that the mechanical engineer should become confused with the metallurgist's phraseology in explaining cause and effect. Gillett[12] has aptly stated that "the design engineer does not have time actually to become a metallurgist. To do so and to keep up with modern advances in metallurgy would take all his time. What he needs to do is to work with the metallurgist and the metallurgists need to work with the engineer. Only by close cooperation of the two groups can there be insurance that all factors, both the obvious and the subtle, are given proper consideration."

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11. Data furnished by Vanadium Corporation (1944).
12. H. W. Gillett, "Metallic Materials," *Steel*, November 22, 1943-February 14, 1944.

Table 7

EFFECT OF VANADIUM ON GRAY IRON

Bar No.	Alloy	Diameter, in.	Tensile, psi.	Hardness, BHN — Surface	Hardness, BHN — Center
QQ1D	No V	3.0	48,450	262	185
QQ2D	With V	3.0	56,850	255	229

Table 8

ANALYSES AND PROPERTIES OF IRONS REPRESENTED IN FIG. 4

Analysis, per cent							Property			
T.C.	C.C.	Si	Mn	Cr	P	S	Transverse Strength, lb.	Deflection, in.	T.S., psi.	BHN
2.80	0.46	2.53	0.77	0.01	0.240	0.092	3500	0.103	36,270	200
2.85	0.52	2.52	0.70	0.48	—	—	4035	0.112	42,550	212
2.98	0.57	2.57	0.80	0.79	—	—	3750	0.103	41,650	217

KNOCK-OFF RISERS FOR NON-FERROUS CASTINGS

S. W. Brinson
and
J. A. Duma
Norfolk Naval Shipyard
Norfolk, Va.

NECKED-DOWN RISERS for castings is not a novel idea, for it was patented many years ago by Washburn. From its widespread use at the time of Washburn, the necked-down riser gradually lost popularity; then suffered neglect to the point of near oblivion.

However, the cry of the military of World War II for faster manufacture of war materials gave impetus to the making of innovations and the adoption of all manner of short-cut methods. And since Washburn's idea had something real to offer both in the way of economy and speed in the manufacture of castings, it was at once put back into use in nearly all of the nation's steel and iron foundries.

Its entrance into the non-ferrous field, however, is recent. To date, as far as the authors know, no non-ferrous foundry has applied the idea of necked-down risers to castings And this despite the fact that, due to the absence of suitable flame burning methods, there has been a long-felt need for such a riser.

In one or two respects the method of application of necked-down risers to non-ferrous castings has changed beyond recognition. Despite this, the principle involved in its use remains the same.

In a ferrous foundry, a necked-down effect is produced by the use of a thin, water-like core resembling a large washer which is placed between the casting and the riser. The core separates the riser from the casting completely except for a small (½ to 2½-in. diameter) central neck through which the riser metal feeds the casting as it solidifies.

Thin Core

It is important that the core be sufficiently thin to allow the heat from the riser metal and the casting metal to penetrate to such an extent that its effect of chilling the metal is negligible. That is, once the wafer core is heated to the temperature of the metal lying on either side of it, its presence therein appears to exert no interfering influence upon feeding.

Certain definite mass relationships are operative in connection with their employment. From data accumulated on cast iron Washburn risers, the following solid geometry relations are indicated:

a. Diameter of risers should equal or slightly exceed the thickness of the section they feed.

b. Height of risers should at least equal their diameter or thickest cross section and not exceed it by more than 50 per cent.

c. Washburn cores are uneconomical and, what is more important, functionally unreliable for lack of sufficient heat input when the size of the riser is appreciably under 3-in. diameter.

With regard to the application of the principle to non-ferrous risers, the following types of riser neck reducers are employed:

1. Perforated steel sheet screens, tinned or untinned, 0.0125-in. thick by one hundred 3/64-in. diameter holes per sq. in.

2. Refractory disk cores, ⅛-in. to 5/16-in. thick.

a. Solid graphitic carbon.

b. Silica core sand.

c. Gypsum metal casting plaster.

Of the foregoing, only two types have withstood the acid test of practical usage, namely, perforated metal screens and graphitic carbon disks (Fig. 1). This is as it should be when examined in the light of the heat properties shown in Table 1. The appraisement as to advantages, disadvantages and limitations of each is given in the following:

Without noticeable impairment of feeding efficiency or sacrifice of metal quality, carbon disks and perforated sheet metal screens are used to give risers which, in the majority of instances, fall off the castings in shake-out or are flogged off with one blow of a hand sledge. Practical methods of making and using gypsum riser sleeves to increase yields in non-ferrous castings have been developed.

Metal Screens

Advantages

1. Low initial cost (37c per 20x28-in. sheet).
2. Economical storage.
3. Easy handling.
4. Low over-all chilling effect.

Presented at a Brass and Bronze Session of the Fiftieth Annual Meeting, American Foundrymen's Association, at Cleveland, May 9, 1946, the opinions expressed in the paper are those of the authors, Master Molder and Foundry Metallurgist, respectively, and do not necessarily reflect the views of the Navy Department.

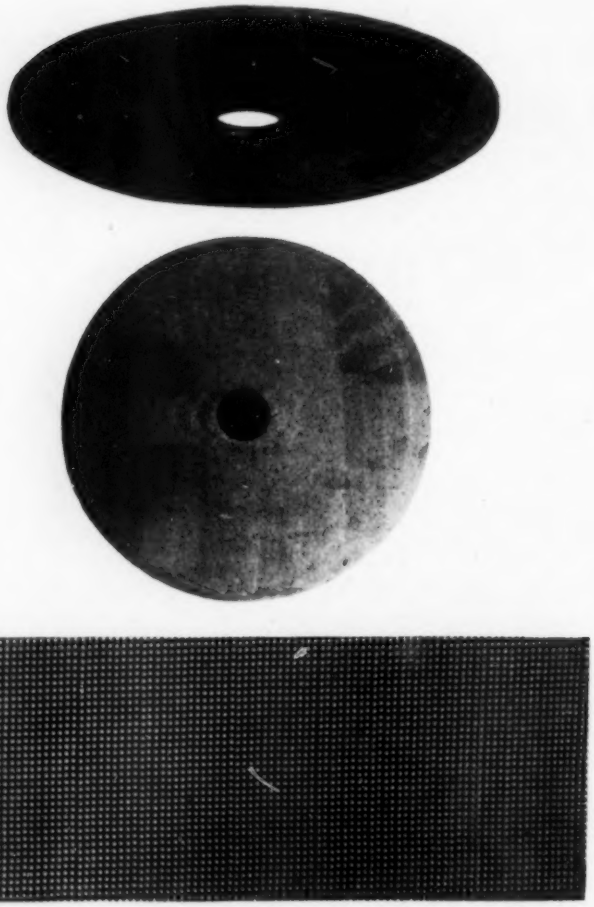
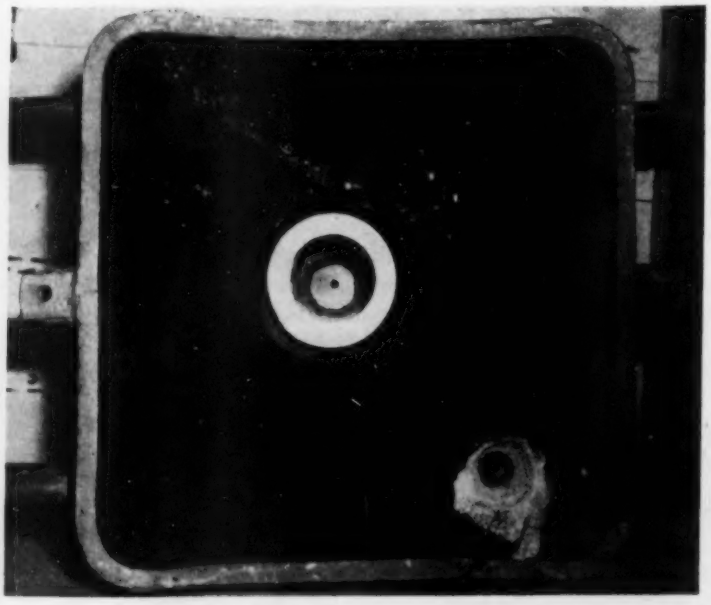


Fig. 1—Perforated metal screen and graphitic carbon disk used to make knock-off risers.

Fig. 3—Cope side of mold shown in Fig. 2, outside surface, showing screen and gypsum riser sleeve.



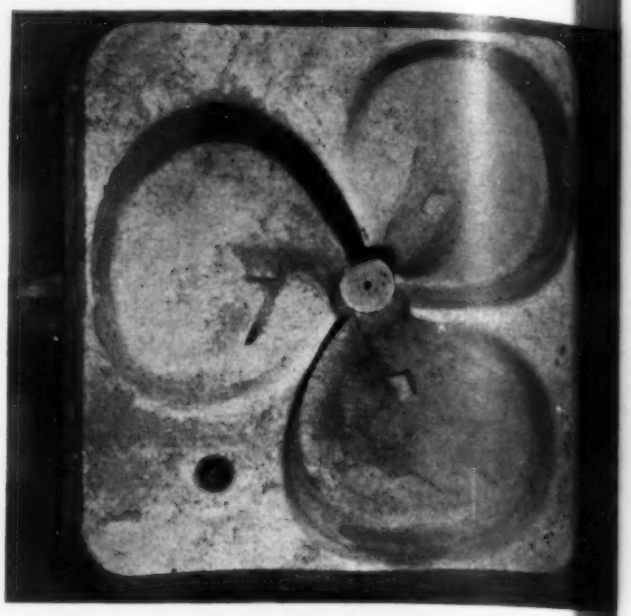
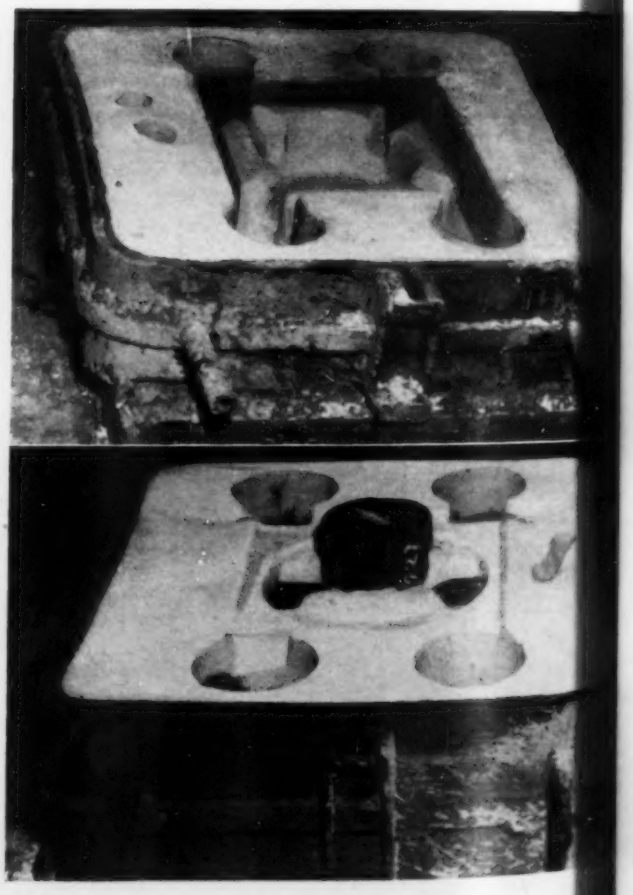


Fig. 2—Mold for 24-in. manganese bronze propeller. Inside surface, cope side, showing location of screen.

Fig. 4—Mold for manganese bronze deck socket. Note screen in drag in front of two side risers.



Disadvantages

1. Cannot be used on bronzes.
2. Limited in application to brasses and low-tin high-strength bronzes.
3. Cannot be used on large necks, i.e., in excess of 20 sq. in.

Graphite Disks

Advantages

1. Equally operative on bronzes, brasses, and monels.
2. Leaves smooth, flat surface.

Disadvantages

1. Cost slightly more than metal screens.
2. Not recommended for risers under 3½ in. in diameter. Best suited for use with large risers for economical reasons.

Sand Cores

Advantages

1. Best suited for use with steel risers.

Disadvantages

1. Cost more than graphite or metal screens.
2. Conducive to gas entrapment.
3. Warp out of shape, producing bulgy or concave surfaces.
4. Thicker pieces are required for handling than with graphite.

Gypsum

Advantages

1. Has no advantages.

Disadvantages

1. Costs more than any of the other three types.
2. Warps, producing bulgy surfaces.
3. Thicker pieces are required for handling than for any of the other three types.

A brief description of the use of perforated metal screens and carbon disks, as practiced by this foundry, is as follows:

a. Screens of the proper size and form are snipped from 20x28-in. rectangular stock. They are then inserted in their proper place atop the riser neck in the mold and secured to the sand with nails. Figures 2 to 5 are illustrative of their use in molds. The castings made in these molds, including others, are pictured in Figs. 4 to 13, inclusive.





Right—reading top to bottom:

Fig. 5—Cope and drag views of mold (Figs. 2 and 3).

Fig. 6—Propeller casting made in mold shown in Fig. 5. Knock-off riser fell off casting on shake-out.

Fig. 7—Same as Fig. 6. Note amount of flat-bottomed shrinkage in gypsum sleeve.



Fig. 8—Deck socket casting made in mold of Fig. 4 with the two knock-off risers.



Fig. 9—Keel type manganese bronze tensile test bars. Note double screen underneath riser.

Fig. 10—Manganese bronze deck socket casting showing two kinds of knock-off risers; screen and graphitic carbon (enclosed in gypsum sleeve).

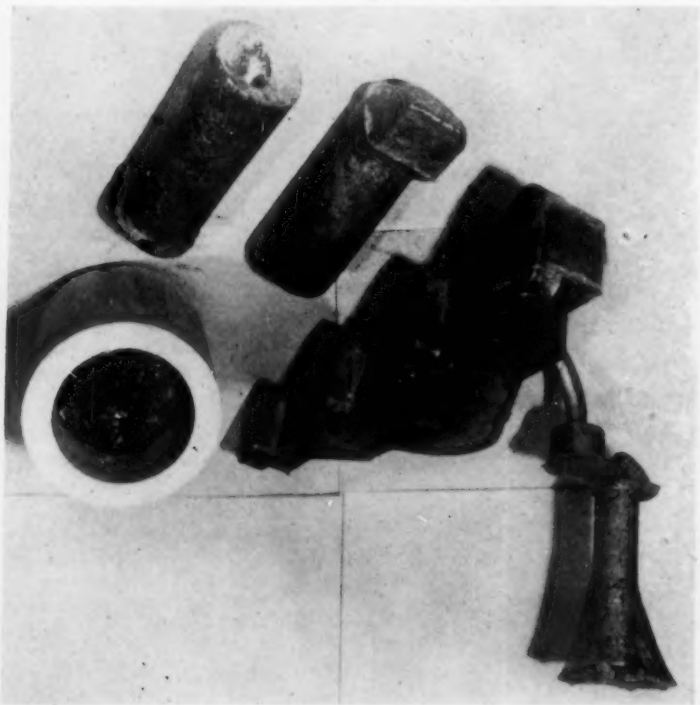


Fig. 11—Manganese bronze hoisting pad casting. Graphitic carbon and riser before knock-off.

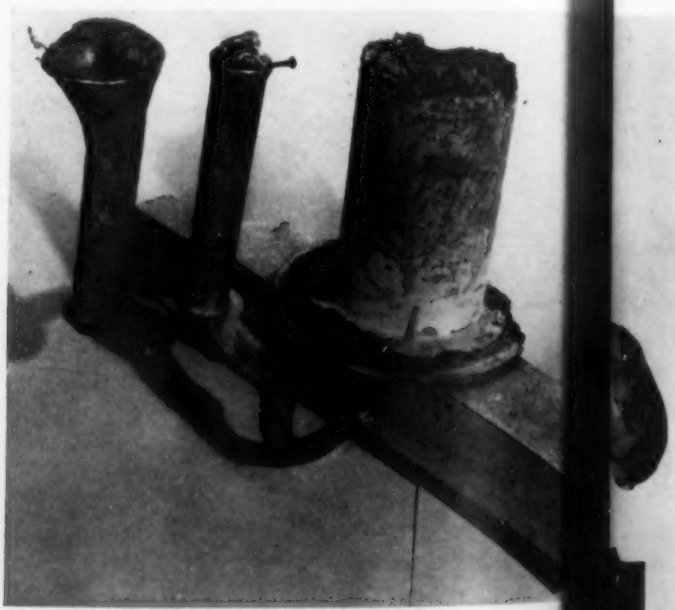




Fig. 12—Gun metal bronze casting. 1⅜-in. diameter neck slightly discernable. Made with graphitic carbon.



Fig. 13—Manganese bronze skeg shoe casting. Center riser is carbon knock-off type, while end riser is screen type knock-off riser.

Fig. 14—Gun metal billet castings made with carbon knock-off type riser. Sectioned to show solidity of metal; (left)—2-in. diameter billet, ½-in. diameter hole in carbon; (center)—4-in. diameter billet, 1-in. diameter hole in carbon; (right)—6-in. diameter billet, 1⅜-in. diameter hole in carbon.



Double sheets of screening are sometimes used with the larger risers.

b. Carbon disks are made by first boring a graphite electrode and then sawing off thin slices. The size of disk selected for any given riser is always 2 in. larger than the size of the riser, thus leaving a 1-in. circumferential shoulder for a print, upon the top surface of which is placed a gypsun sleeve.

As to the size of the central hole in the carbon, the test castings shown in Figs. 14 and 15 show that the proper hole size for any casting is equal to 20-25 per cent of the riser diameter. To cause the plane of breakage to locate slightly above the flat surface of the casting, the edge about the hole on either side of the carbon is broken with a file to give it a double-bevel face.

Occasionally, several small ⅛-in. diameter holes are drilled midway between the center and outside of the carbon to aid in the expulsion of mold gases from the mold cavity. Disks are cut ⅛-in. thick for all sizes up to 6-in. diameter, 3/16-in. thick for diameter sizes of from 6 to 10-in., ¼-in. thick for those of greater than 10-in. diameter. Castings generally are gated at the bottom-most point; pouring of metal is continued into the mold until metal comes through neck of carbon; pouring is completed by top pouring of riser.

At present, the use of both of the foregoing neck reducers is restricted to flat surfaces and generally to production set-ups involving 25 or more castings. An exception is made in the case of large castings requiring large risers where an order for one is handled as carefully as orders for large numbers.

Table 2 is offered as a guide for the use of neck reducers for the different kinds of metals and sizes of risers.

Savings effected in cleaning castings by the use of screens and carbon disks are shown in Table 3. Ease and economy are effected in the actual removal of risers by flogging. This economy rests in savings of time, of grinding and cut-off wheels, band saws, chipping chisels, melting electrodes and the attendant metal loss (sawings, chips, spatter), and the labor of sweeping up the same.

Keel block type test bars, with and without metal screens in the riser

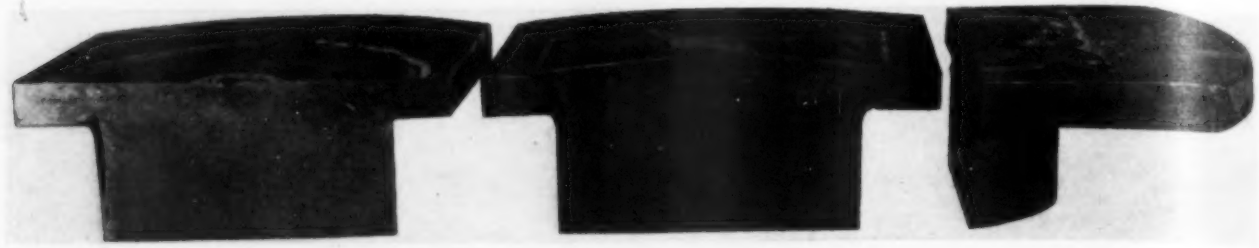




Fig. 16—The working tools for making gypsum riser sleeves.

Fig. 15—(above) Same castings shown in Fig. 12, sectioned. Note humping of one casting caused by upward bulging of gypsum core; (left)—carbon, 6½-in. diameter by 1⅜-in. diameter hole; (middle)—gypsum, 6-in. square by 1⅜-in. diameter hole.

Table 1

THERMAL PROPERTIES OF MATERIALS USED IN KNOCK-OFF RISERS

Material	*Thickness, in.*	*Diameter, in.*	*Weight, grams*	*Specific Heat at 1000° C. —Cal./G./° C.*	*Total of Calories Absorbed in Heating from 20 to 1000° C.*
Steel	0.0150	4	13.5	0.16[1]	2,140
Graphite	0.1250	4	58.6	0.29[2]	16,650
Core Sand	0.1875	4	79.1	0.26[2]	20,180
Gypsum	0.3125	4	35.9	0.26[3]	9,150

[1]*Handbook of Chemistry and Physics*, p. 1672.
[2]*Textbook of Metallurgical Calculations*, Butts, p. 377.
[3]*Handbook of Chemistry and Physics*, p. 1675.

Table 3

COMPARATIVE CLEANING COSTS

CONVENTIONAL RISERS	*3-in. Diameter Riser*	*9-in. Diameter Riser*
Sawing or Chipping	0.180	1.770
Light (Trim) Chipping	0.016	0.033
Dress Grinding	0.033	0.105
TOTAL	0.229	1.908
FLOGGING COSTS		
Screen	0.008	0.060
Carbon Disk	0.080	0.130
Labor to Knock-Off	0.020	0.030
Dress Grinding and Light Chipping	0.025	0.050
TOTAL	0.133	0.270

Table 4

COMPARATIVE PHYSICAL PROPERTIES OF CONVENTIONAL AND KNOCK-OFF RISER METAL

	Screen Riser		*Conventional Riser*	
Tensile Strength, psi	75,500	74,250	73,750	74,500
Elongation, per cent in 2 in.	30.0	30.5	31.5	31.0

Table 2

RECOMMENDED MATERIALS FOR USE ON KNOCK-OFF RISERS

Riser Size, in.	*Bronzes*[1]	*Brasses*[2]	*Monels*[3]
2½	—	Screen	Carbon
3	Carbon	Screen	Carbon
3½	Carbon	Screen	Carbon
4	Carbon	Screen	Carbon
5	Carbon	Screen or Carbon	Carbon
6	Carbon	Carbon	Carbon
8	Carbon	Carbon	Carbon
10	Carbon	Carbon	Carbon
12	Carbon	Carbon	Carbon

[1]N.D. Specifications 46B5, 46M6, 46B8, 46B9, 46M20, 46B21, 46B22, 46B23, 46B24.
[2]N.D. Specifications 46B3, 46M4, 46B10, 46B11, 46B18, 46B29.
[3]N.D. Specifications 46M1, 46N7, 46C8.

neck, were poured from the same heat of manganese bronze and tested as shown in Table 4. No significant difference in physical properties was obtained.

Inasmuch as both of the aforementioned knock-off risers generally are designed for use with gypsum sleeves, it is deemed advisable at this point to include herein a brief description of the technique set up for making and using gypsum sleeves.

The idea of gypsum sleeves for insulating risers is credited to the Naval Research Laboratory. Either regular metal-casting plaster or, preferably, gypsum type R2 are suitable for use in the basic mixture which is prepared as follows:

Gypsum	1 pt.
Cement (by vol.)	2 oz.
Water	1 pt.

This mixture is poured into sheet steel (19 gauge) cylinder molds split longitudinally and buckled together

with two clip fasteners, as shown in Fig. 16. The cylinder rests on towel paper. A tapered aluminum core is centered with respect to the cylinder mold.

Any desired combination of cylinder and core diameter sizes can be selected to make sleeves of any practical wall thickness. "Yard" practice has adopted the 1-in. wall as standard for all gypsum riser sleeves. No anti-sticking sprays are used. Five minutes after pouring, the aluminum core is withdrawn, the metal mold unbuckled, and the gypsum sleeve is air dried for 2 hr., then furnace dried for 10 hr. at 500° F.

After drying, the sleeves are either used immediately or stored in a storage rack. If stored for over 1 day, they are again dried at 400° F. for 15 min. prior to use. When used in non-standard lengths, they are sawed with a hacksaw blade to the required length by the molder.

The graph in Fig. 17 gives the thermal and physical properties of gypsum when used as gypsum sleeve material. The use of gypsum sleeves about the riser, and gypsum powder on top of the riser, results in an increase in riser yield of approximately 30 per cent.

No appreciable contamination of the sand system is encountered from the use of gypsum sleeves. They are hand-picked off the shake-out screen and discarded either whole or in large fragments by the shake-out crew.

Summary and Conclusions

In conclusion, the following observations are listed:

a. Comparing the relative efficiency of knock-off risers in both the ferrous and non-ferrous divisions of founding, it is stated that a greater number of sound castings (completely free from remnant shrinkage in necks) as well as easier knock-off of risers is obtained with non-ferrous castings.

b. Perforated metal screens and graphitic carbon disks effect distinct economies in the cleaning of brass and bronze castings.

c. Perforated metal screens and graphitic carbon disks do not disadvantageously affect the feeding of risers.

d. Perforated metal screens function best in brass risers not in excess of 5-in. diameter.

e. Graphitic carbon disks function with equal effectiveness in both brass and bronze risers, and in all sizes of risers.

f. Highest economies are realized when the number of castings is large or when the risers themselves are large.

g. For the same amount of feeding, gypsum sleeves enable the use of risers approximately 30 per cent smaller than conventional gravity-feed type risers.

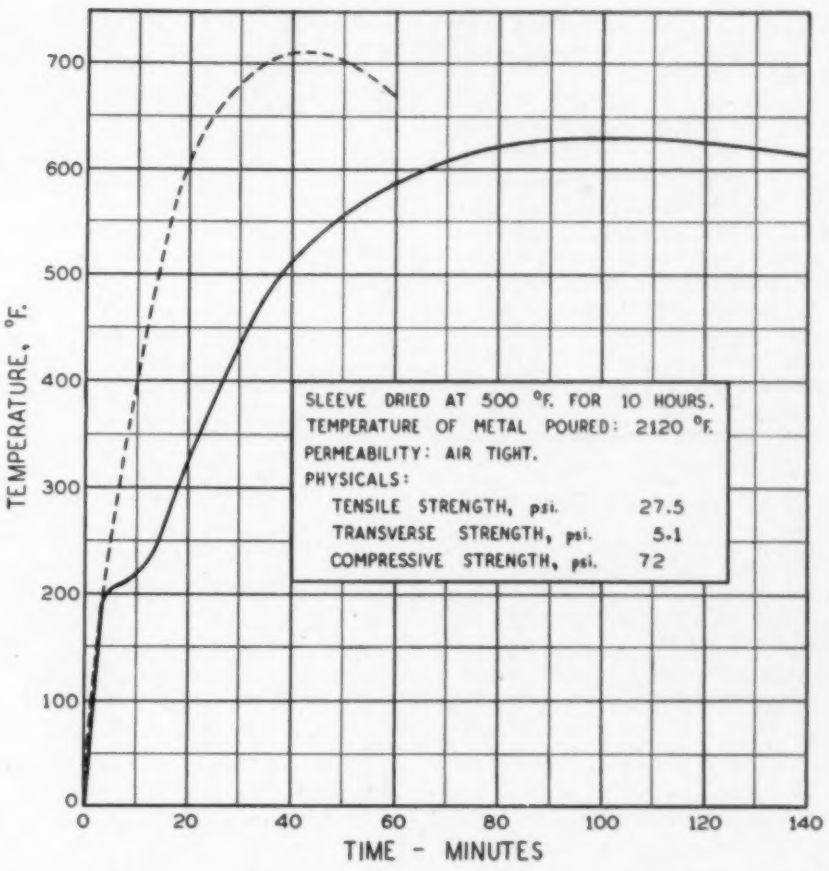


Fig. 17—Graph showing temperature rise of gypsum sleeve after pouring of casting.

DISCUSSION

Chairman: G. K. Dreher, Ampco Metal, Inc., Milwaukee.

A. Most[1]: In using the iron screen, is there any danger of iron contamination of bronzes from the molten bronzes contacting the iron screen?

Mr. Brinson: None at all. The screen is sheet steel with a tin coating. We had an experience in that connection with a steel casting at one time. In closing the mold the molder used two pieces of tinned sheet steel to stop off a fin by chilling and keeping the metal from going into the mold. He inserted them a little too far and got them into the casting cavity about an inch. This was a large casting, weighing about 25,000 lb. as poured. After it was poured, we pulled the sheet metal out of the steel casting. Volatilization of the tin formed a cushion around the sheets and kept the casting metal from contacting the tinned sheet steel.

Chairman Dreher: You do peel the sheet metal off the risers and throw it away when you are through with it. You do not let it contaminate the back scrap.

Mr. Brinson: No, we remove it immediately.

George Dalbey[2]: We tried the perforated sheet on manganese bronze and we could not knock the riser off. We believed it was alloyed with the sheet and we took some of that same sheet and detinned it with acid and tried it again and had no difficulty.

Mr. Brinson: I can not offer a reason for that. We have been using it daily for over a year and never had trouble. We wanted to show the riser before and as it was knocked off, but we could not do it because the riser was knocked off as it came out of the mold.

Chairman Dreher: I think, Mr. Dalbey, that you could use the screen without the tinning, just as an ordinary piece of steel. It has been used successfully in that manner.

Mr. Brinson: The steel comes tinned. Detinning is an added expense.

Chairman Dreher: The important point is that it seems to work both ways.

Lt. Com. B. O. Brouk[3]: I would like to ask Mr. Brinson what results he had

[1]Parkway Foundry & Machine Co., Brooklyn.

[2]U. S. Navy Yard, Mare Island, California.

[3]Bureau of Ships, Washington, D. C.

with the use of the tinned screen in connection with G-metal? Did you encounter an adverse condition?

MR. BRINSON: We were unable to knock the riser off and and that is what we were after.

MR. DUMA: The difficulty we had was that the risers stuck right to the screen. The union between screens and cast metal was so strong that no amount of hand hammering could flog the risers off. It appears that high tin (over 2 per cent) copper-tin alloys wet the surface of iron screens by tinning them and then alloy with the iron at the iron-tin interface.

MR. BRINSON: In some cases we have used two sheets, one on top of the other, and still were able to knock the riser off.

E. D. BOYLE[4]: Mr. Brinson made the statement that one investigator was working on the problem of gas pressures. That is exactly what we are doing. Within the next two years we may have sufficient data on gas contamination of risers creating pressure.

MR. BRINSON: That is one of the problems we have to solve.

H. F. TAYLOR[5]: I wonder if Mr. Myskowski, who did the pioneer development and experimental work on the uses of the screen, would comment with regard to the oxidation of the screens for use with tin bronzes?

E. T. MYSKOWSKI[6]: It works satisfactorily.

MR. TAYLOR: Did you find any improvement when you put on an oxide layer rather than a layer of tin, for example?

MR. MYSKOWSKI: It seemed to work very well; it did not tend to stick as much.

MR. TAYLOR: I think it may be tied up with a wetting condition. Whether it sticks or does not stick might depend upon whether the metal wets, or possibly alloys with, the coating.

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[4]Puget Sound Navy Yard, Bremerton, Wash.
[5]Massachusetts Institute of Technology, Cambridge, Mass.
[6]Naval Research Laboratory, Washington, D. C

CORE OVENS

Equipment and processing methods for cylinder head molds and cores and magnesium casting cores — construction and operation of baking and drying ovens — mechanical materials handling installations. Magnesium core sand mixtures identified by color system. Care and storage of cores in air conditioned storage rooms. Operation of sand reclamation system.

H. E. Linsley
Wright Aeronautical Div.
Wright Corp.
Paterson, N. J.

THE AUTHOR CANNOT, by any stretch of the imagination, be considered a foundryman. However, he can claim at least a nodding acquaintance with core ovens, having worked around gray iron, bronze, aluminum and magnesium foundries for more than 20 years.

The company with which the author is associated claims the distinction of operating the country's first aluminum and magnesium foundries designed solely for the manufacture of aircraft engine castings. While these may not be the largest foundries of this type today, they have provided a considerable fund of experience.

The heaviest production of the foundries is in cylinder heads, since 18 of these are required for each 2,200-hp. engine. They are poured in dry sand molds. Each cylinder head casting requires the preparation and baking of two half molds and six internal cores.

The cylinder head molds (Fig. 1) are made on rollover machines and are transferred directly to racks which travel on a conveyor chain between the double rows of molding machines. The molds weigh approximately 90 lb. each. At the end of the machine line the conveyor lifts sharply upward to enter the baking oven, which is carried on structural members 8 ft. above floor level.

The oven is a continuous conveyor type with an over-all length of 105 ft., and is so arranged that the conveyor chain carrying the racks must travel in multiple passes, a total distance of some 290 ft. through the preheating, baking, and cooling zones. A period of approximately 7 hr. is required for a mold to pass through the oven (Fig. 2).

Heating equipment is located at one end of the oven and consists of four pairs of external oil-fired burners. One burner of each pair operates continuously, while the other cuts in or out automatically to maintain a constant operating temperature of 450° F. Circulation is by means of induced draft fans mounted on the floor directly under the oven.

All continuous ovens are mounted overhead, thus saving floor space and allowing more comfortable working conditions by getting the heat nearer to the roof where it can be dissipated through ventilators. The open space underneath the ovens is convenient for storage, and in some cases box-type ovens are installed there.

After cooling the molds are sprayed with a fluoride solution, and then are loaded onto individual trays suspended from a conveyor chain, which carries them through a drying oven (Fig. 3). The drying oven is heated by means of an external gas-fired heater and is 16 ft. wide, 50 ft. long, and 5 ft. high. It is set on structural members, well above floor level. The oven is completely insulated with rock wool on walls, roof and floor, and is divided down its length by a sheet-metal partition to separate the two directions of conveyor travel.

Heat from the gas burners (Fig. 4) is blown through floor openings into a hot air spray duct mounted along the center line on the floor of the oven. The conveyor makes two passes through the oven in counter-flow direction and, when operating at the normal temperature of 325° F., requires about 1¾ hr. to pass a mold through.

The cores for the cylinder heads were, at one time, baked on the same equipment used for the molds, but the tremendous increase in production made it necessary to provide separate ovens for the purpose (Fig.



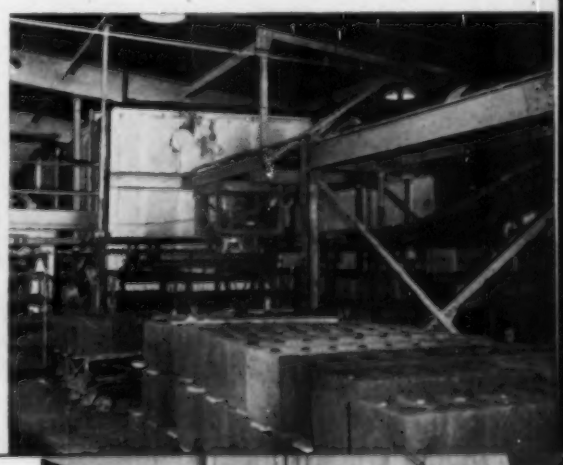
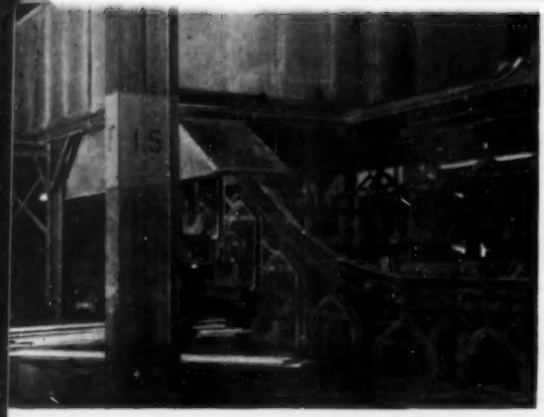


Fig. 1 (left)—Mold for aircraft engine cylinder head. Note the deep, closely spaced fins.

Fig. 2 (right)—Discharge end of oven used for baking molds shown in Fig. 1.



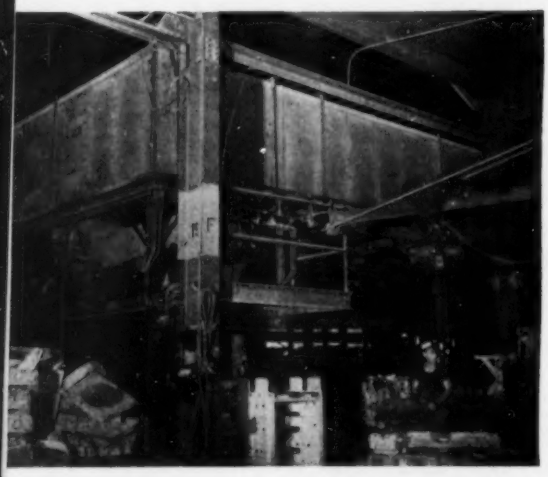


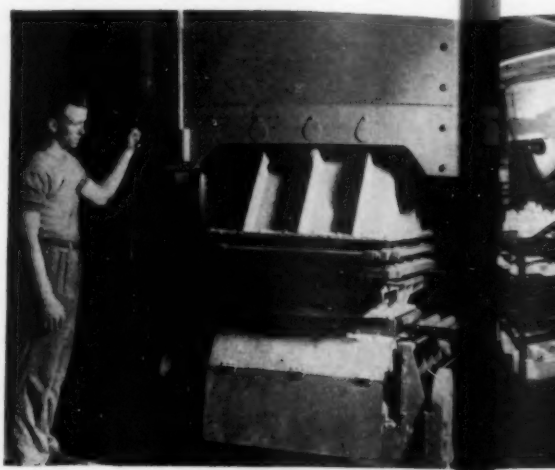
5). At the present time a tower-type oven, 10 ft. long, 10 ft. wide and 28 ft. high is being used. It has a capacity of 22 core racks, each of which has three shelves with an area of 15 sq. ft. per shelf. The operating temperature is 650° F., and production is 7.3 racks per hour.

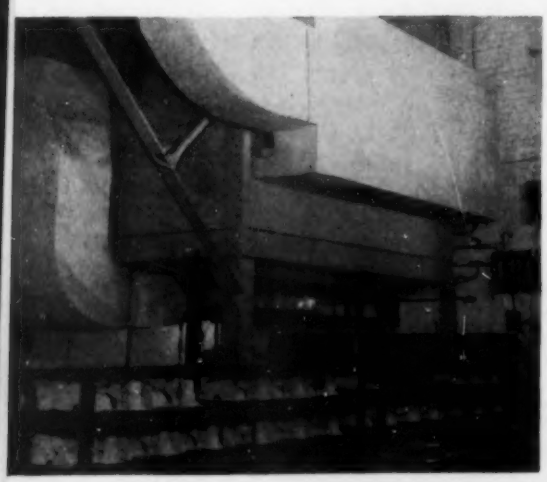
The oven is heated by means of a proportioning type oil burner which fires into a refractory-lined combustion chamber. Recirculated air is mixed with the hot gases and introduced through a series of damper outlets into the evaporation zone and suction chamber.

Parallel or crossflow is provided in the oxidation or baking zone, with a neutral draft exhaust to the stack for spilling excess gas. A recirculating fan picks up the recirculated air and gases, and returns them to the wind box. The installation is, of course, completely protected with automatic temperature and safety controls.

In addition, the oven is equipped with ten individual, single-point recorders to show the temperature at each of ten zones. It is a well-known fact that uneven baking will









Reading (left) top to bottom:

Fig. 3—Entrance to oven for drying molds after spraying with fluoride solution.

Fig. 4—Fan and burner installation for spray drying oven.

Fig. 5—Automatic, tower-type core oven used for baking cylinder head cores.

Fig. 6—Natural draft cooling tunnel for cooling cylinder head cores leaving tower oven.

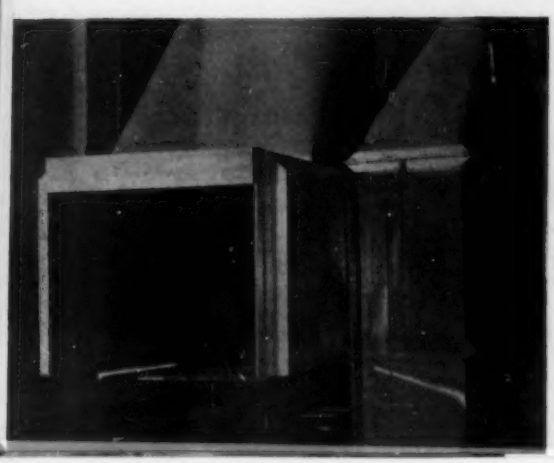
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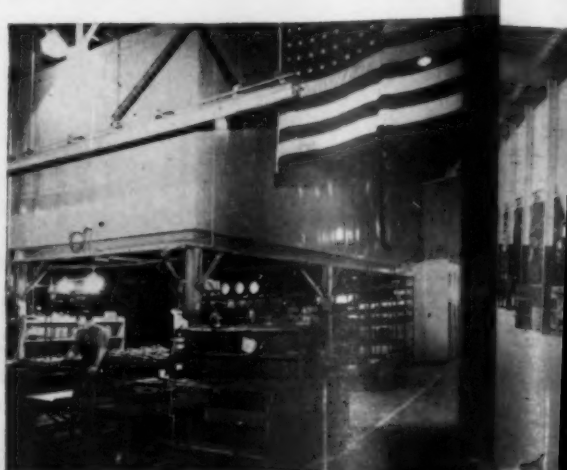
Fig. 7—Core blower used for small and medium sized cores for magnesium castings.

Fig. 8—Core blower used for large magnesium casting cores.

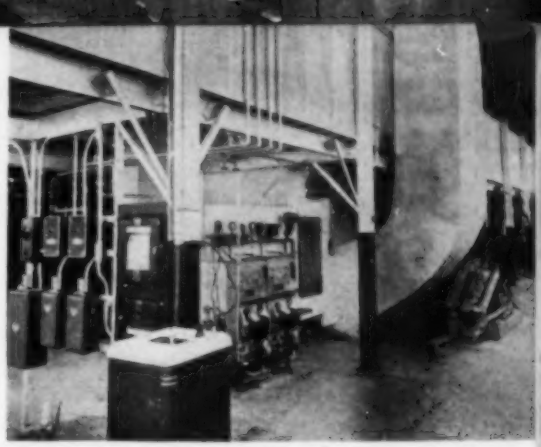
Fig. 9—Stationary ovens for fragile and experimental cores.

Fig. 10—Continuous oven for baking production magnesium casting cores.







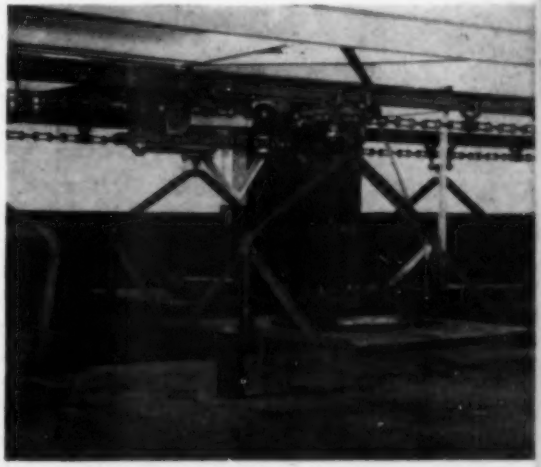


result in poor cores, and checking the temperature at just one point is not a guarantee that this temperature will be constant throughout the oven. The multiple check system enables the operators to spot any zones which are too hot or too cold and regulate the dampers accordingly.

Handling the Cores

The cores are made on core-blowing machines and are loaded directly onto the racks. As a rack is filled it is moved along by a power-operated roller conveyor which delivers it to the loading position in the front of the oven. Hydraulic jacks then raise the rack just high enough to clear the rollers and enable it to be pushed onto the carriers of the oven conveyor.





The racks travel up one side of the oven and down the other to the discharge position, where they are delivered to a second powered conveyor which carries them through a cooling chamber (Fig. 6). Cold air is drawn over the cores in this tunnel by natural draft, cooling them sufficiently to be handled with bare hands when unloading and placing on the storage shelves. The empty





Reading (left) top to bottom:

Fig. 11—Loading magnesium casting cores on a conveyor.

Fig. 12—Loaded core racks entering corebaking oven.

Fig. 13—Typical multipiece core for magnesium casting built up from a large number of separate cores.

Fig. 14—Oven designed for drying inhibitors in cores, now used as a paste drying oven.

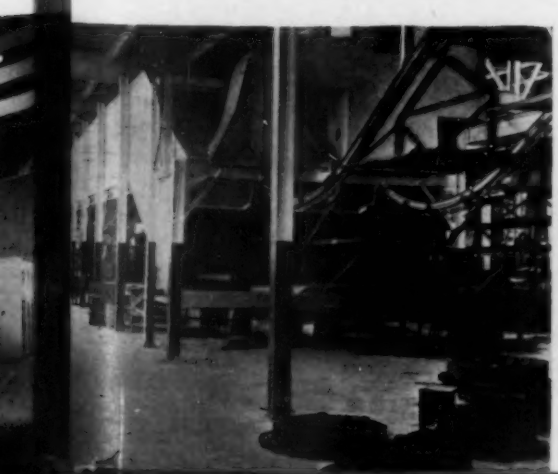
Reading (right) top to bottom:

Fig. 15—Fan and burner equipment for spray drying oven.

Fig. 16—Loading and unloading end of core drying oven. Note proximity of core storage.

Fig. 17—General view of sand reclamation installation.

Fig. 18—Inside view of furnace showing one of the ten hearths over which sand is rabbled during reclaiming process.





racks are picked up by an electric monorail hoist and carried across the room to the starting point of the loading conveyor.

The development of the engine made by this company has resulted in an increased use of magnesium, with a corresponding decrease in the amount of aluminum. Of the parts originally made from aluminum, some 25 to 30 per cent now are made from magnesium. At one time both metals were handled in the same foundry, but the present practice provides for separate foundries.

Colors for Core Sands

All of the magnesium castings are made in green sand molds, but cores are, of course, baked in the usual manner. Anyone entering the magnesium coreroom could be excused for blinking rapidly and wondering whether or not he were seeing things, because he would see the familiar tan color of the sand blossoming out into dazzling hues of pink, and blue, and green. These colors are not used because they look pretty, but as a means of quick and positive identification of the different sand mixes.

Curiously enough, when the practice of coloring the sands was first started, the production rate of the women coremakers went up about 20 per cent. A bit of trouble developed with the girl who "just couldn't stand green," and the girl who wanted to work with blue sand because "blue was her favorite color," but that was soon straightened out. The dye is added in liquid form during the first mulling and, while it does not retain its full color after baking, it is still readily identifiable.

Core Blowing

In the newest foundry of the company many of the magnesium cores are machine blown. They range from small to very large cores (Figs. 7 and 8).

Since these cast parts are for an aircraft engine in which weight must receive first consideration, many of the cores used are exceedingly complex, and necessitate the use of light cross sections. Such cores must be handled with great care, and the nevitable slight jolting of a conveyor is sufficient to fracture them when in the green state.

Such cores are baked in stationary, three-compartment ovens heated by external, gas-fired heaters (Fig. 9). Each compartment is equipped with an individual heating, recirculating, and exhaust system, and each measures 6 ft. wide, 15 ft. long, and 6 ft. high. Doors are of the vertical lift, counterweighted type, and each compartment will hold two core racks with a load of 4,000 lb. per rack.

Normally these cores are baked for 4½ hr. at a temperature of 450° F., but a wide range of control is possible on each compartment so that special jobs or experimental work on new cores and new sand mixes can be carried on without interfering with routine production in the other ovens.

Regular magnesium cores are baked in a trolley conveyor oven, (Fig. 10) equipped with four-shelf trolleys (Figs. 11 and 12). This oven is 140 ft. long 20 ft. wide, and 9½ ft. high, and is located about 8 ft. above floor level.

Trolleys are spaced at 8-ft. intervals and travel at the rate of 37 ft. per hr., giving a capacity of 3,700 lb. of cores per hr. The oven is insulated by 5 in. of aluminum foil, and is divided lengthwise by a partition set off-center so that the first section accommodates one pass of the conveyor, and the second section houses two passes.

Heaters and Controls

The fully automatic, oil-burning heaters and controls, together with the circulating and exhaust fans are located at floor level near the entering end of the oven, the heaters being rated at 4,000,000 B.T.U. per hr., and the circulating fan having a capacity of 27,000 cu. ft. per min.

The air supply is conveyed from the heaters into two distributing ducts running along opposite sides of the oven, and is blown downward through special slots in the walls. This sets up a circulatory motion and induces a large amount of secondary air. The air moves in the direction opposite to that of the conveyor, and a large part of it is recirculated from the first pass of the oven. The remainder is exhausted from the entering end by means of the exhaust fan.

The oven originally was designed for an operating temperature of 450° F. at a time when this temperature was necessary for the type of core oil used. However, since the introduction of low temperature baking oils it has been possible to operate at 350° F. with a maximum entering temperature of 475° F. in place of the original temperature of 580° F. This has resulted in a tremendous saving of fuel oil.

Complex Cores

As previously mentioned, many of the cores are very complex and are built up from a considerable number of small, separate cores (Fig. 13). These cores are assembled and pasted to form a complete unit, and then spray dried in a conveyor oven (Fig. 14). This oven is of the continuous-trolley type, 94 ft. long, 13 ft. wide, and 6 ft. high, and is built similarly to the corebaking oven except that it is less heavily insulated because of its lower operating temperature of 300° F. Carriers are spaced at 8-ft. intervals, and each carrier has a capacity of 300 lb. of cores and plates. The rate of travel is 16 carriers per hr., which means that each carrier will remain in the oven for about 1¼ hr. This is insufficient time for the core to attain the full temperature of the oven but is sufficient to dry the paste.

At one time it was the practice to spray a sulphur-boric acid inhibitor over the surfaces of the cores and to dry this when drying the paste. This additional step was necessary because the high corebaking temperature burned out the inhibitors when they were mixed with the sand. Since the introduction of the low temperature baking oils it has been possible to mix the inhibitors with the sand without danger of their burning out, and the spraying practice has been discontinued.

Heating Equipment

The heating equipment consists of a two-section direct-oil-fired air heater, a supply fan capable of delivering 19,000 cu. ft. per min. at a temperature of 350° F. into the oven, and an oven exhaust fan. This equipment is located on the floor near the return section of the oven (Fig. 15) and delivers the heated air through continuous slotted ducts situated in the upper corners of each pass of the oven.

Recirculation is from the first pass only, and dampers are supplied to regulate the amount of air in each pass and the recirculation to the heater. The burners are rated at 2,000,000 B.T.U. per hr., and the

normal operating temperature in the oven is 300° F.

Just before the entrance to the oven the conveyor chain is arched to clear a main aisle, and then makes a right-angle bend to pass the assembly benches for loading unfinished cores, and unloading finished cores into the core storage room (Fig. 16).

Core Storage

In one of the foundries is installed a completely dehumidified core storage room where, by means of batteries of silica-gel units, air heating and cooling units, etc., the moisture content of the air in the room can be brought down to a maximum of 13 grains per lb. This type of equipment is particularly important when cores are made on a production basis and held until required, since moisture in the core will react with the molten magnesium and produce hydrogen, which promotes burning of the metal and results in rejected castings.

It is not proposed to go into the details of this core storage room, which has been described in detail in the January, 1944 issue of *Heating, Piping and Air Conditioning*, but it is worth mentioning as indicating the care which should be taken in the handling of cores.

Reclaiming of Sand

Also of interest is a recently installed process for reclaiming core sand (Fig. 17). Without going into detail, this process consists of crushing the sand lumps, removing the wires and chills, and then passing the sand through a ten-hearth vertical rabble furnace where all carbon and organic matter are burned out, and clean white sand is returned to the foundries (Fig. 18).

It is estimated that this process saves the equivalent of 30 freight cars of sand a day, to say nothing of eliminating the eternal difficulty of used sand disposal.

Many foundry men are apt to assume that cores are something in which they have no proper concern except that the cores fit the molds properly and behave as they should. Granted that this is an age of specialization, with melter, pourer, coremaker, and molder each attending to his own phase of the work, the author feels that if each were to know at least a little about the other fellow's job he would find it of tremendous value in his own work.

Acknowledgments

In conclusion, the author wishes to express thanks to the various members of the Wright foundry staff for the assistance given in the preparation of this paper, and to *Industrial Heating* for permission to re-use certain material.

OPERATION AND CONTROL of the CUPOLA

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THE SUBJECT of cupola operation and control is, to say the least, a most controversial one and, in spite of the fact that there are almost as many cupola procedures as there are operators, results are, for the most part, quite satisfactory.

The discussion which follows is a brief description of the practices followed in the foundries with which the writer is connected, and are not necessarily intended to serve as "cure-alls" for the entire foundry industry.

One of the first requirements for successful cupola operation is an exact knowledge of the composition and behavior of the raw materials used. This involves chiefly the metallic components of the cupola charge and the metallurgical coke or cokes on hand or available. Needless to say, the limestone, dolomite, or whatever fluxing material is used, should be of the best quality.

Each carload of pig iron should be checked against the shipper's analysis. The analysis of a carload of purchased cast iron scrap would be an almost endless task and would be of very little value after it was completed. However, this should not discourage the use of large percentages of cast scrap in the cupola charge, as the composition of the various types of scrap available is known quite definitely.

However, the party responsible for the purchasing of scrap should insist that any given carload contain only one type of scrap and not a mixture of several kinds. Each carload should be rigidly inspected, both upon arrival in the iron yard and during the course of unloading to the storage piles or to the charging cars or buckets.

After a brief period of training in the recognition of various types of scrap, crane operators and charging crews can offer invaluable assistance to the cupola foreman, as they see practically every piece of iron used during the course of operation.

Weighing equipment for metal and coke charges must be checked frequently and kept in good repair. Above all, when metal composition must be held within very narrow limits, as is required in certain types of castings such as cylinder liners, and when making malleable iron by the duplex process, where the initial melting and superheating is done in the cupola furnace, the weighing equipment selected must be accurate, sensitive over the range in which it is used, and intelligently chosen for the job it is intended to do.

For example, 250 or 300 lb. of coke (or 25 lb. of 50 per cent lump ferrosilicon) ordinarily cannot be weighed accurately on a set of scales designed for heavy loads where the tare weight of the charging car and bucket may be 3,000 or 4,000 lb. This is not a hypothetical case, as just such a set of conditions exists in one plant of this company, and will be discussed in more detail later.

The cupola has proved itself a most versatile melting unit, capable of producing a wide variety of irons satisfactory for innumerable engineering applications. Successful cupola operation requires an exact knowledge of the composition and behavior of the raw materials used — metallic components of the cupola charge, metallurgical coke and fluxing materials.

While it may be considered rather elementary at this time to enter into the matter of "lighting-off" the cupola coke bed, there are one or two items which are worthy of note. In one of this company's foundries the kindling wood and practically all of the bed coke are put into the cupola at one time. About 6 hr. before "wind-time" the bed is "lit-off" and allowed to burn through very slowly. Before any metal is charged the bed is levelled off with cold coke, and charging then proceeds until the cupola is full.

In another foundry the bed coke is put into the cupola in three equal portions at 2-hr. intervals, while in a third department about one-half of the bed is "lit-off" only about 4 hr. ahead of "wind-time," and here it is necessary to use a gentle blast to thoroughly ignite the coke. All of these procedures are satisfactory and, if sufficient kindling wood is used, the first iron tapped will be amply hot.

The longer burning-in time is preferable, as this results in better preheating of the cupola lining and well, and when cupolas are operated on alternate days plenty of time is available. However, when it is necessary to use the same cupola every day there usually is less time available (after completion of the daily "patching" job) between light-off-time and wind-time.

In order to provide a large volume of fluid slag as soon after the start

of the heat as possible, two extra charges of limestone are placed on the bed just before charging starts, and 4 lb. of fused soda ash per ton of metal are used on each charge used to fill the cupola in addition to the regular charge of limestone.

The layer of slag thus formed acts as a protective covering over the molten iron in the cupola well, resulting in considerably higher melting temperatures at the start of the heat. This may be due to the fact that the slag blanket protects the molten iron from the cooling effect of the incoming blast, or heat losses through radiation from the surface of the molten metal are reduced.

The foundries with which the writer is connected produce sand-cast pressure pipe and fittings, a complete line of chilled plows, diesel engine cylinder jackets and liners, marine engine bearings, bases, frames and cylinders up to 70-in. bore with 48-in. stroke, automotive and agricultural implement castings, pressure valves for oil, water and gas, locomotive cylinders, and machine tool castings.

A variety of castings used by the chemical industry, such as caustic pots, ingot molds, jacketed kettles, lead pots, billet molds and the like are produced. In addition to these, the company manufactures super-deLavaud pipe ranging in size from 3 to 36 in. in diameter and 18 ft. long, cast centrifugally in water-cooled steel molds.

Figure 1 shows a low pressure marine engine cylinder casting having a 70-in. bore and 48-in. stroke. The average chemical analysis of the iron used for these cylinder castings will fall within the following ranges:

Component	*Per Cent*
Silicon	1.70-1.90
Sulphur	0.09 max.
Phosphorus	0.20 max.
Manganese	0.75-0.90
Total carbon	3.00-3.25

The silicon content of the iron from the cupola spout is somewhat lower than that found in the finished castings and, consequently, a "late addition" of 75 per cent grade ferro-silicon is made. This addition increases the silicon content from 0.20-0.30 per cent. A "late addition" of 0.3 per cent (3 lb. of alloy per 1000 lb. of molten iron) of zirconium silicide also is made to the molten iron. These additions cleanse the iron, reduce chilling tendency, and materially increase toughness.

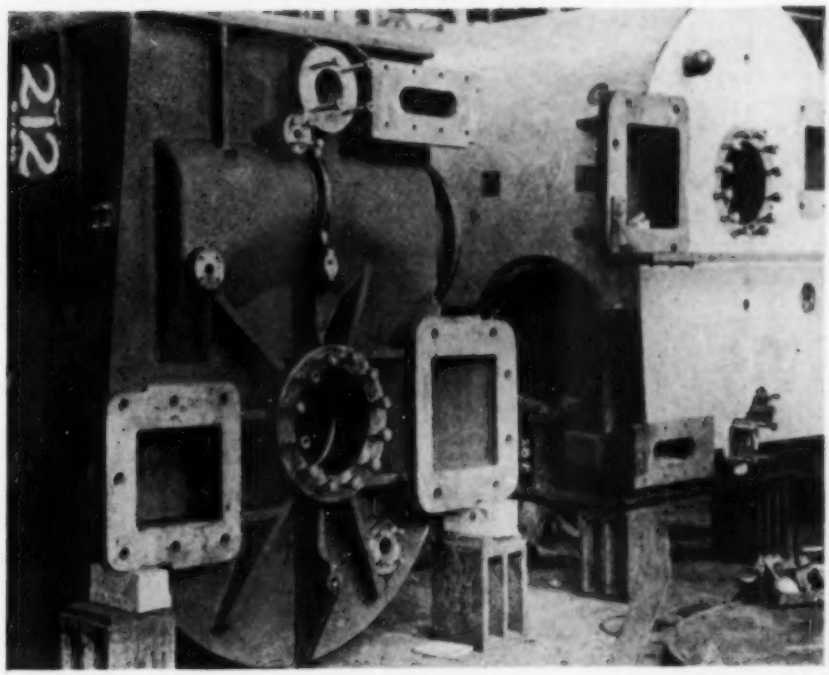


Fig. 1—Low pressure marine engine cylinder, 70-in. bore and 48-in. stroke.

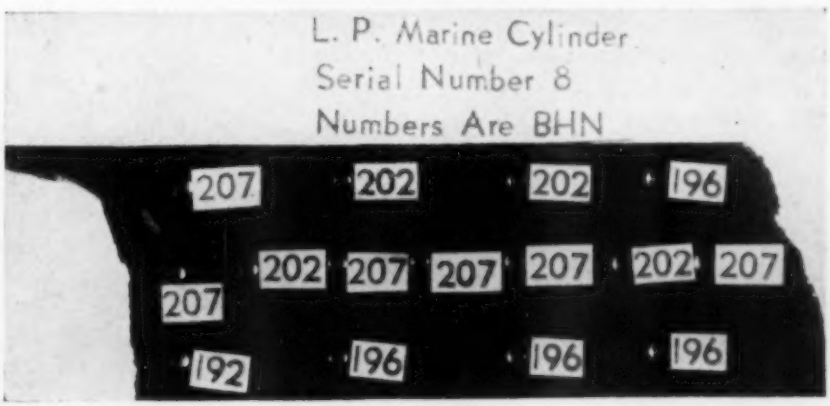


Fig. 2—Section from 2⅜-in. main wall of low pressure cylinder casting showing results of hardness tests.

The strength of tensile test specimens machined from separately cast standard arbitration bars runs from 42,000 psi. to 48,000 psi., with corresponding Brinell hardness numbers ranging from 217 to 241. On three occasions tensile test specimens have been machined from the 2⅜-in. thick main wall of cylinder castings. The results of these tests along with results obtained on corresponding specimens machined from standard arbitration bars are tabulated in Table 1.

Figure 2 shows a section from the 2⅜-in. thick main wall of a low pressure cylinder casting. Indentations are impressions made by the 10-mm. ball on the Brinell hardness testing machine, and figures indicate Brinell hardness at the various points. The uniformity in hardness is attributed to the "processing"

Table 1

PHYSICAL PROPERTIES

1.20 Bar Specimens		*Cylinder Wall Specimens*	
Tensile Strength, psi.	*BHN*	*Tensile Strength, psi.*	*BHN*
42,673	228	34,000	196
43,990	228	35,905	196
48,200	235	34,900	192

treatment to which the iron was subjected.

A cylinder liner casting (21.5 in. I.D.) just after removal from the mold is shown in Fig. 2. The liner is in casting position with sprue at right and cylindrical gate at bottom. The riser shown at upper left is 5 in. in diameter and 18 in. long. Chemical composition and physical properties of the iron used are as follows:

Component	*Per Cent*
Silicon	1.15-1.40
Sulphur	0.10 max.
Phosphorus	0.20 max.
Manganese	0.70-0.90
Total carbon	3.00-3.25
Molybdenum	0.40-0.50
Vanadium	0.12-0.18
Chromium	0.15-0.30

Tensile strength on specimens machined from Class C (2.00-in. diameter) test bars is 43,000 to 48,000 psi. The cupola mixture used for the cylinder liner casting is of the following percentage:

Malleable pig iron	30.0
Silvery pig (8 per cent Si)	7.5
Steel scrap (No. 2)	57.5
Manganese steel scrap	5.0

A hardness test specimen machined from a cylinder liner casting is shown in Fig. 4. A section of the 5-in. riser from the cylinder liner casting (Fig. 3) is shown in Fig. 5, and the machining operation is shown in Fig. 6.

Carbon Control

From the foregoing it is apparent that many "grades" of iron are necessary to produce castings of proper strength and composition to render them suitable for the applications for which they are intended. For example, the iron used in large cylinder liners (21.5-in. I.D.) will contain far too little total carbon for use in super-deLavaud pipe, and the iron used in a caustic pot would be much too high in total carbon for use in cylinder liners and so on.

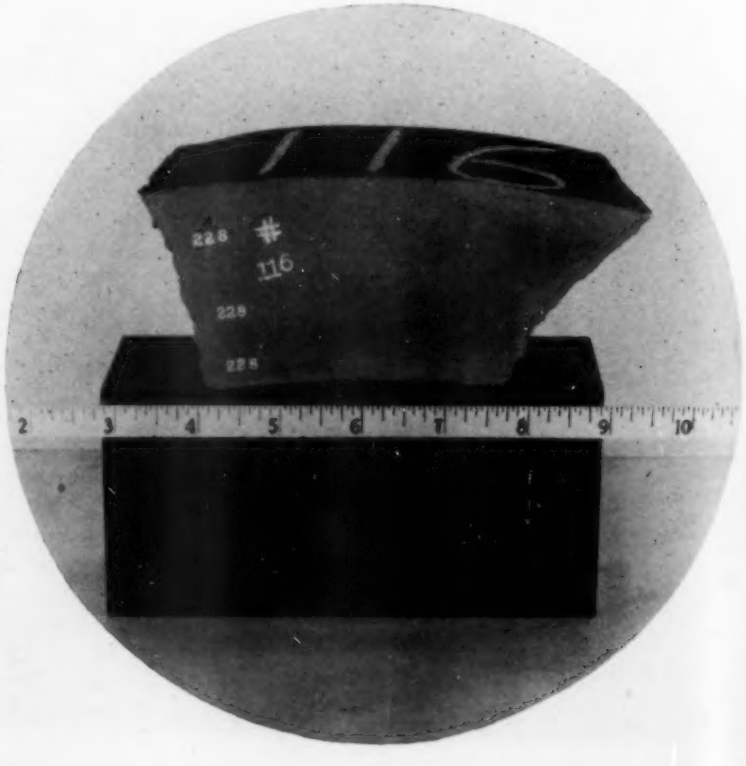
While the writer has not overlooked the importance of the other elements entering into the chemistry of the iron, it is felt that the total carbon present is of extreme importance in its influence on the properties of the finished product.

The total carbon content of the iron is, of course, influenced by factors such as the percentage of steel scrap and/or pig iron in the charge, the amount and type of coke used, blast pressure, blast temperature, melting rate, tapping technique,



Fig. 3—Cylinder lining casting (21.5-in. I. D.) after removal from mold, shown in casting position. Sprue at right and cylindrical gate at bottom. Riser shown at upper left is 5-in. diameter and 18-in. length.

Fig. 4—Hardness test specimen machined from lower end of cylinder liner similar to one shown in Fig. 3. Brinell hardness of 228 was recorded at three points across the section, which is 2⅝-in. thick.



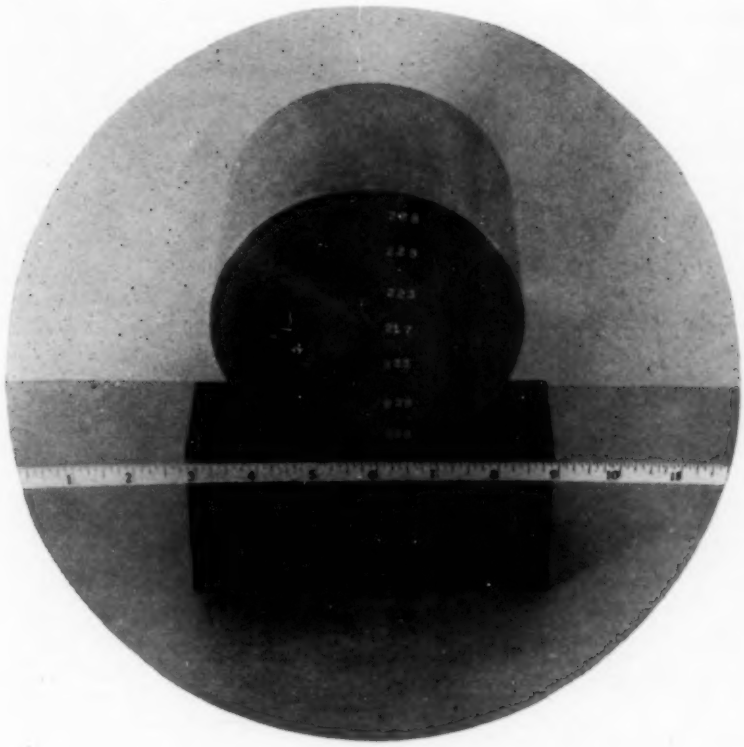
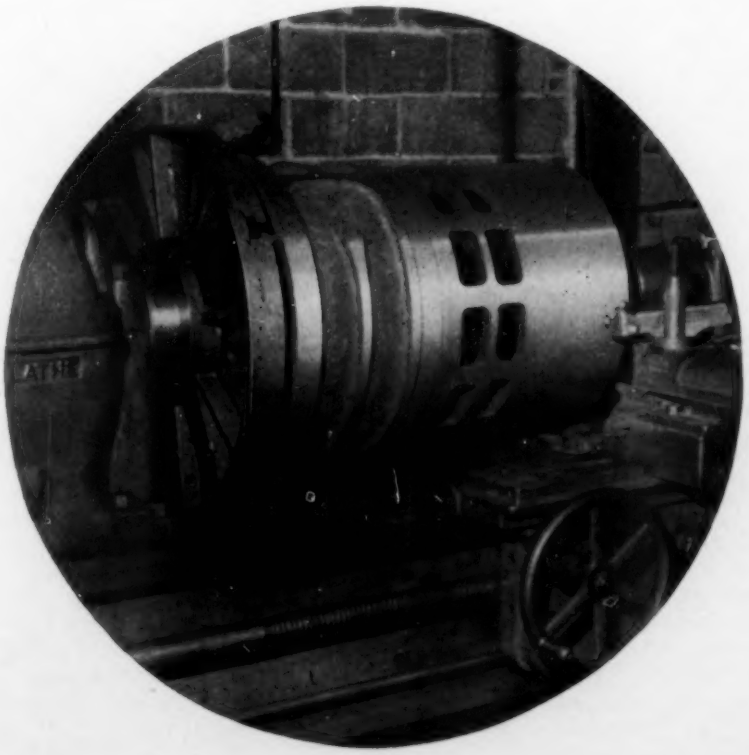


Fig. 5—Section of a 5-in. diameter riser removed from a 21.5-in. I. D. cylinder liner as shown at upper left in Fig. 3. Note Brinell hardness figures range from 217 at the center to 228 near the outer edge. This test piece and the test piece shown in Fig. 4 were from the same casting.

Fig. 6—Cylinder liner in process of machining. Eight of the 12 port openings can be seen.



melting temperature, auxiliary equipment used in connection with the cupolas, and many others.

As previously stated, a definite knowledge of the chemistry of the raw materials used is very important. For example, total carbon determinations are made of all samples of pig iron in addition to the determination of the usual elements. This, along with a knowledge of the total carbon content of the other materials making up the charge, makes it possible to calculate the carbon in the charge. This calculation is made in the same way as the calculation for silicon, for example, except that there is no deduction made for loss in melting, because in a well-run cupola there is almost without exception a carbon pickup.

Recording Cupola Mixtures

If proper records of cupola mixtures are kept, they will in time be invaluable to the operator in making up mixtures to produce irons of predetermined carbon content and, furthermore, it will enable him to make a graph or graphs showing the total carbon level to be expected in the iron at the cupola spout when the carbon in the charge is known.

All of the cupolas operated by the company with which the author is connected are of the front-slagging type which, of course, means that iron flows from the taphole continuously from start to finish of the heat. At no time is there any appreciable accumulation of molten iron in the well of the cupola, but the small amount which does collect remains at a constant level because it flows out as fast as it is melted. This type of operation results in uniform carbon pickup, which is one of the essentials of carbon control.

Intermittent Tapping

In cupolas which are tapped intermittently the amount of molten iron in the well may range from several tons (in the larger cupolas) to practically none in the few minutes required to make a tap. This means that the depth of iron within the cupola is constantly changing, and under these conditions the carbon pickup is apt to be less uniform than when the metal is removed as fast as it is melted.

It is not inferred that uniform carbon pickup in the cupola cannot be obtained with intermittent tap-

Table 2

COKE TYPES—ANALYSIS

PERCENTAGE

	By-product	*Beehive*	*Pitch*
Fixed carbon	90.0-92.0	93.0-95.0	98.0 plus
Ash	7.0- 9.0	5.0	0.50
Sulphur	0.6	0.5	0.30
Volatile matter	0.6- 0.9	0.9	1.0

ping, but it requires a little closer supervision. An exact tapping cycle must be established, and the time between taps must then be maintained by means of a stop watch.

Frequently, with intermittent tapping, a cupola tapper can see no harm in allowing the time between taps to vary 10 or 20 or 30 seconds, although actually this may be a high percentage of the established time cycle (between taps) and variations of the magnitude just mentioned may materially affect the percentage of total carbon in the iron tapped, especially in mixtures containing high percentages of steel. For the purpose of carbon control, the company with which the writer is connected has on hand three types of coke, used either singly or in combination. The approximate compositions of these cokes are shown in Table 2.

By-Product Coke

All other things being equal, the total carbon in irons melted with by-product coke is lower than that obtained when using either of the other types, and the pitch coke produces irons higher in total carbon than either of the other cokes,while melting with beehive coke produces irons of intermediate total carbon content.

From this it is obvious that there is an infinite number of possible coke combinations which could be used to help control the total carbon. As an example of the effect of different types of coke on carbon pickup, the comparison between the beehive and by-product types is shown in Table 3.

With mixture "A" the carbon pickup was 1.32 per cent, which is the difference between the average total carbon in the melted iron (3.04 per cent) and the calculated carbon in the charge (1.72 per cent). With mixture "B" the carbon pickup was 1.04 per cent, which is the difference between 3.05 per cent in the melted iron and 2.01 per cent calculated to be present in the charge.

The carbon pickup with mixture "A" may be looked upon as unusually high (as compared to results obtained with mixture "B") inasmuch as it contained 25 per cent more steel and 18.1 per cent less automotive cast scrap calculated at 3.20 per cent total carbon than mixture "B" (these percentages being based on the amounts of materials present in mixture "B"). This difference in carbon pickup is attributed to a large extent to the difference in characteristics of the types of coke used.

From previous experience it has been found that had mixture "A" been melted under identical conditions with 250 lb. of by-product coke (instead of beehive), the total carbon in the iron coming from the cupola would have been in the range of 2.75-2.85 per cent.

Pitch Coke

As an example of the effect of pitch coke on carbon pickup, the comparison shown in Table 4 is given to show the effect of replacing a portion of the beehive coke with pitch coke.

Results of a large number of determinations have shown that when using cupola mixture "C" melted with 300 lb. of beehive coke, the total carbon in the melted iron will fall in the range of 3.00-3.15 per cent, while the same mixture melted with 250 lb. of beehive coke plus 50 lb. of pitch coke will produce irons having a total carbon in the range of 3.25-3.40 per cent.

Other mixtures containing 12.5 per cent southern pig iron (No. 2) and 87.5 per cent cast scrap, when melted with 100 per cent beehive coke in the ratio of 10:1, produce irons having a total carbon in the range of 3.35 to 3.45 per cent, while the same mixtures with the same coke ratio produce irons close to 3.60 per cent total carbon when one-third of the beehive coke is replaced with pitch coke.

Mixtures containing 75 per cent southern pig iron and 25 per cent cast scrap using beehive coke in the ratio of 15:1 yield irons containing slightly above 3.70 per cent total carbon, and these same mixtures when using by-product coke in the same ratio produce irons containing about 3.60 per cent total carbon.

Mechanical Charging

As previously stated, auxiliary cupola equipment plays an important part in control and uniformity of the molten iron. All cupolas in this company's foundries are charged mechanically with the exception of one, which will be changed over as soon as installation can be made.

Mechanical charging usually improves the operation because it is easier to keep the cupola filled to the charging door level at all times during the melting operation, and the types of charging equipment used distribute both the metal and coke in the cupolas in a much more satisfactory manner than is accomplished by hand.

This is especially true with cupolas of larger inside diameters, where charging crews often are prone to place the heavier components of the charge near the charging door opening and in hot weather more of the

Table 3

COKE TYPE EFFECTS ON CARBON PICKUP

Materials	*Mixture "A"** lb.	*per cent*	*Mixture "B"* lb.	*per cent*
Silvery pig iron (8 per cent Si)	100	5	100	5
Steel scrap (No. 2)	1,000	50	800	40
Automotive cast scrap	900	45	1,100	55
Totals	2,000	100	2,000	100
Carbon in charge		1.72		2.01
Coke per charge	250 (Beehive)		250 (By-product)	
Coke ratio	8:1		8:1	
Cupola size	72 in. I.D.		72 in. I.D.	
*Total Carbon***				
Range, per cent	3.00-3.06		3.00-3.08	
Average, per cent	3.04		3.05	

*Mixture "A" contained 4 lb. Si in form of 50 per cent lump to bring Si in both charges to same level.

**In each case total carbon determinations were made on 12 consecutive ladles (2 tons each) of the melted iron.

coke charge than should be is placed near the charging door, in order to protect the "chargers" from heat inside the cupola.

Weighing Equipment. Referring again to cupola mixtures "A" and "B" mentioned in the foregoing, it is pointed out that the results with mixture "A" indicate a total carbon of 3.03 per cent plus or minus 0.03 per cent, and with mixture "B" the total carbon was 3.04 per cent plus or minus 0.04 per cent. This uniformity was accomplished in a large measure by exercising great care in weighing of all the components of the cupola charge.

The steel, an automotive cast scrap, was loaded with a crane and magnet on a 12,000-lb. capacity platform scale. The charging car and bucket weigh 4,000 lb., which, of course, has to be taken into account by the scaleman or weighmaster. Furthermore, this man looks over the materials being charged and sorts out any materials not shown on his charge sheet.

Weighing Cupola Charges

The silvery pig iron (100 lb. per charge) was weighed on a portable platform scale of 500-lb. capacity, and several piles of this material weighed in advance are shown in Fig. 7. The coke was weighed on the same set of scales as was used for the silvery pig iron. The containers are steel drums cut in two and with handles welded to each half. The gross weight is marked on each container and, for convenience and ease of handling, the coke is weighed in batches of 100 lb. or less, as shown in Fig. 8.

The 50 per cent lump ferrosilicon, totaling only 8 lb. per charge, was weighed as shown in Fig. 9 on a 24-lb. capacity spring scale of the type frequently used by housewives around the kitchen. The foregoing procedure, when properly supervised, certainly leaves very little room for doubt regarding the accuracy of weighing of the materials used and the operator is always rewarded with uniformity, not only in total carbon results but in all of the elements involved.

Cupola Blast Moisture Control

One of the variables of considerable magnitude which affects cupola operations is the moisture content of the air used for the blast. This may vary from 2 grains per cu. ft. on cold winter days to as much as 12 grains per cu. ft. on hot sultry days. The moisture content of the air may also vary within the limits referred to during the course of a single day's melting operation.

There are 7,000 grains in one lb. Now let us consider the possible variation in the amount of water carried into a cupola with the air blast when melting 200 tons per day, as is often the case at one of the plants being discussed. This operation will require about 5,200,000 cu. ft. of air, so that the water blown into the cupola will be between 1,500 lb. (180 gal.) and 9,000 lb. (1,080 gal.).

From these figures it is evident that the control or elimination of a variable of such magnitude is well worth consideration. For this reason equipment for controlling moisture in the cupola blast was installed at this plant.

The installation is designed to remove moisture from 13,000 cu. ft. of air per min. down to 4 grains per cu. ft. in the summer season, and also is capable, in dry weather, of introducing as much as 7 grains per cu. ft.

The principle of operation lies in the fact that the system contains a solution of lithium chloride which, under the proper conditions, has a large capacity for absorption of moisture from the air. However, this absorption capacity can be controlled and predetermined by regulating the temperature of the lithium chloride solution.

Moist air from the outside is taken in through a filter to remove dust and dirt. It then passes through the "contactor cells" in intimate contact with a spray of lithium chloride solution, where moisture is removed (or added) to a predetermined amount within the design limits of the system.

To prevent loss of lithium chloride solution due to carry-over, the air passes next through a series of glass-wool eliminators, and finally is drawn through a set of heating coils and thence to the cupola blower intake. The heating coils maintain the cupola blast at a temperature of 105° F.

When a cupola is operating so as to require 10,000 cu. ft. of air per min. taken into the moisture control system, with 8 grains of moisture per cu. ft. and leaving with only 4 grains per cu. ft., the lithium chloride solution is taking on about 40 gal. of water per hr. To maintain efficient operation of the system, this excess water must be removed continuously and as rapidly as it is absorbed.

This is accomplished in the regenerator unit, where a part of the diluted lithium chloride solution, after being heated to a temperature of 235° F. in the solution heater, is sprayed over a relatively small contactor surface where a small stream of outside air passes over it. At the elevated temperature (235° F.) referred to the solution gives up the moisture which was previously removed from the cupola blast.

The lower part of the regenerator unit serves as a storage space for the solution when the system is not in operation, and it also houses a series of filters which constantly remove foreign material from the solution when the system is in operation.

Table 4

Effect of Pitch Coke on Carbon Pickup

Materials	*Mixture "C"* *lb.*	*per cent*	*Mixture "C"* *lb.*	*per cent*
Malleable pig iron	700	35	700	35
Steel scrap (No. 2)	1,100	55	1,100	55
Duriron scrap	100	5	100	5
Manganese steel	100	5	100	5
Totals	2,000	100	2,000	100
Carbon in charge		1.66		1.66
Beehive coke per charge	300		250	
Pitch coke per charge	none		50	
Total coke per charge	300		300	
Coke ratio	6.7:1		6.7:1	
Cupola size	72 in. I.D.		72 in. I.D.	
Total carbon, per cent	3.00-3.15		3.25-3.40	

In the winter season it is necessary to add moisture to the cupola blast in order to maintain a year-'round constant control. This is accomplished by means of a small heating coil where the solution is heated with steam to a temperature at which it will give up moisture to the cupola blast in just the amount required to maintain proper control. The principle of operation of the heating coil is somewhat the same as that of the regenerator.

While it so happens that the company with which the author is connected has installed a moisture control system for maintaining a year-'round uniform moisture content of 4 grains per cu. ft. of cupola air blast, this figure in itself is not particularly significant. Whether the system be designed for 2 grains or 5 grains or any other figure, the significant feature is that variations in moisture content of the cupola blast have been eliminated.

It is well to point out at this time that even though the moisture control system is fully automatic in its operation, it requires daily inspection and setting of some of the control valves and thermostats. This is taken care of by a laboratory technician, who spends an average of not more than 30 min. per day to make all adjustments.

Air Weight Control Blower. As previously stated, about 5,200,000 cu. ft. of air are required to melt 200 tons of iron in the cupola. Air at 14.7 lb. psi. atmospheric pressure and 60° F. weighs 0.0764 lb. per cu. ft. It is evident from these figures that for every lb. of iron melted, practically one lb. of air is required. Every effort is made to weigh the metal charges accurately and, to insure uniformity of operations, the air should be weighed accurately.

To this end an automatic air weight control blower was installed. This blower is operated by means of a 150-hp. motor (D.C.) and is capable of delivering 13,000 cu. ft. of air per min. at a pressure of 2 lb. The amount of air delivered to any one of the four cupolas lined to 72-in. I. D. is measured in lb. per min. so that the blower reading in lb. per min. will be close to the melting rate of the cupolas expressed in lb. of metal per min.

The older blowers used by the company are of the positive displacement type. Each revolution delivers a definite volume of air to the cupolas, so that running at a constant rpm. will deliver a fairly constant volume of air.

However, to deliver a constant weight of air to the cupolas with the older blowers would require constant adjustment in speed corresponding to the slightest changes in atmospheric temperature and pressure. For this reason it is obvious that the constant-air-weight-control blower eliminates the variables of constantly changing atmospheric conditions.

Balanced Blast Cupolas. Two of the four cupolas at one of the company's plants and two cupolas at another plant are of the balanced-blast type, each equipped with one row of main tuyeres and two rows of auxiliary tuyeres. The main tuyeres are provided with slide valves so that they may be opened or closed at will during the course of the heat.

Two of the main tuyeres diametrically opposite are closed at the same time so that the area directly in front of them is heated to a temperature sufficiently high to melt off all overhanging slag. After 8 or 10 min. these are opened and two others are closed. This procedure is continued throughout the duration of the heat and results in the prevention of bridging and blocking up of the main tuyere openings.

The openings of the auxiliary tuyeres are adjustable, but the settings cannot be changed without shutting off the cupola blast. The upper, or auxiliary tuyeres, bring about more complete combustion of the coke and thereby make possible a more efficient cupola operation, and in conjunction with the controllable main tuyeres the balanced-blast cupolas melt somewhat faster and at higher temperatures with the same coke ratios than the cupolas provided with tuyeres of the conventional type.

Processing the Iron. As previously stated, all of the cupolas operated by the company are of the front-slagging type. Molten iron from the cupolas runs directly into forehearths or teapot ladles, where an amount equivalent to several cupola charges

Fig. 7—Silvery pig iron (100 lb. per charge in this case) being weighed in advance on portable platform scales. Note several piles at right ready to be placed in charging buckets.

Fig. 8—Coke charges are weighed on same platform scales as shown in Fig. 7. Gross weight is marked on each container.



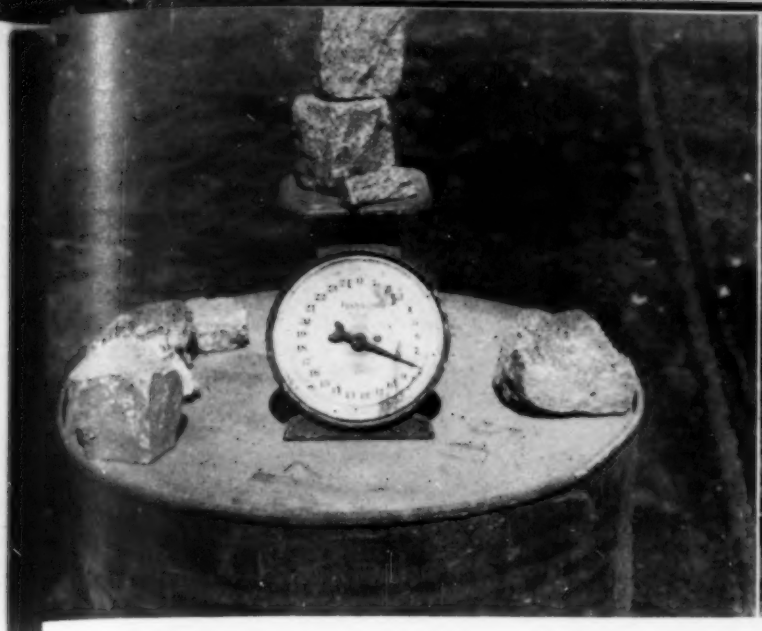


Fig. 9—Small 24-lb. capacity spring scale is used to weigh 50 per cent lump ferrosilicon when used in small amounts (8 lb. in this case).



Fig. 10—Caustic pot made in accordance with the process described in the text. Inside diameter, 10 ft. 3¾-in.; height, 8 ft. 3 in.; weight, 30,000 lb. Metal section ranges from 2½-in. near top to 3½-in. in bottom. Chemical composition in percentages: silicon, 1.04; sulphur, 0.054; manganese, 0.90; phosphorus, 0.12; total carbon, 3.81.

is stored and mixed, resulting in greater uniformity of metal composition than would be obtained when tapping directly from the cupola into the pouring ladles.

This is especially true when the capacity of the pouring ladles is small (1,000 lb. or less). The size of the forehearth ladle is such that it will hold as much molten iron as the cupola melts in 10 to 15 min. From this it is obvious that different sized forehearth ladles are used with different sized cupolas.

While the forehearth is being filled with molten iron and after each ladle is taken out, the metal (in the forehearth or teapot ladle) is treated with fused soda ash for the purpose of reducing the sulphur content, removing entrained silicates, increasing fluidity, and for deoxidation and grain refinement.

As the metal is transferred from the forehearth or teapot ladle to the pouring ladles, it is inoculated in all cases with a minimum of 0.2 per cent (2 lb. per 1,000 lb. of molten iron) of zirconium silicide, which is an alloy containing 35-40 per cent zirconium and 50 per cent silicon. This is an excellent scavenger and deoxidizer.

In addition to this, 75 per cent grade ferrosilicon also is added, as all of the mixtures are calculated with the silicon content in the iron at the cupola spout somewhat (0.20-0.40 per cent) below that desired in the final product, thus deliberately leaving room for a "late" addition of ferrosilicon. This type of operation has a very distinct advantage in that it makes it possible to pour castings varying widely in section size from the same "base" iron.

This is done by adjusting the cupola mixture to "fit" the castings having the heavier metal sections. For lighter castings varying amounts of ferrosilicon are added, and for the heaviest castings very little ferrosilicon and varying amounts of ferrochromium or one of the stabilizing types of inoculants may be added. In addition to those just referred to, other alloys and inoculants are added to the molten iron when necessary.

Inoculant Types

These may be either the graphitizing type or the stabilizing type. When producing high-strength irons for use in thin-walled castings, double inoculation often is resorted to, the purpose being to facilitate machining. Double inoculation consists in making additions of both the graphitizing and stabilizing types, and has been a very successful procedure especially in cases where low-carbon-equivalent irons are used.

All of the ordinary alloying elements such as nickel, chromium, molybdenum, copper, titanium and vanadium may be used separately or in combination, depending upon the class of iron to be produced, the particular application, or the customer's specifications. The purpose of the use of inoculants is to control the structure of the iron in the castings, that is, to produce irons having a pearlitic matrix, free of ferrite, with graphite flakes in random distribution.

From each ladle of iron a chill-test specimen is poured which is used as a guide and quick checkup for the cupola operator. The size, or dimensions, of the chill-test specimen will depend upon the composition of the metal, and this in turn is governed by the size and type of castings to be poured. All chill-test specimens are poured against water-cooled, copper chill blocks so that chill depth is not affected by variations in the temperature of the chill block.

As a further aid to the cupola foreman, strict laboratory control is maintained at all times. When the cupolas melt one type or "grade" of iron for several hours in succession, samples for complete chemical analysis go to the laboratory hourly.

Results of total carbon determinations are returned to the cupola foreman within 10 min. and the silicon, sulphur, manganese, and phosphorus are reported as soon as they have been completed. This usually requires 45 min. to one hour. Two sets of test bars (2 per set) are poured daily from each type of "regular" iron.

When producing high-strength irons, certain irons made to customers' specifications, and practically all special castings, test bars and samples

for laboratory analysis are poured from each ladle. The bars are handled in the physical testing laboratory, where transverse strength, deflection, and Brinell hardness are determined.

Tensile-test specimens are machined, on a lathe set up in the laboratory especially for this purpose, from the halves of broken test bars, and tensile strengths are determined in the usual manner. The laboratory is of ample size and has adequate facilities for handling large numbers of samples for complete chemical analysis, and is equipped to make all of the physical tests required to furnish complete data on all castings produced.

Likewise, there is a complete microscopic laboratory with darkroom facilities. These are used constantly for visual examination of the structure of the irons, and in certain instances photographic records are kept.

Producing High Quality Cast Iron. A series of custom-made, high quality cast irons (patented) is produced by the foundries with which the author is connected. The method of manufacture is based upon the use of synthetic pig iron containing a maximum of 0.20 per cent of phosphorus. This pig iron is made by melting down a cupola charge consisting of 100 per cent steel scrap or other ferrous materials containing relatively small percentages of silicon and carbon, such as malleable scrap or any combinations of these materials.

During the melting process, in the cupola, the charges absorb upward of 3.00 per cent of carbon, and thus the iron tapped out will have a composition generally falling within the following limits:

Component	*Per Cent*
Silicon	0.10-0.40
Sulphur	0.10 max.
Manganese	0.20-0.40
Phosphorus	0.20 max.
Total carbon	3.10-3.50

This iron is collected in a 10-ton capacity teapot ladle where several of the cupola charges are mixed in order to obtain uniformity throughout the bath of metal. From time to time, during the melting process, ladles of iron are taken out of the teapot ladle and poured into ingot molds. During this transfer of molten iron from the teapot ladle to the pouring ladle, additions of silicon, manganese, zirconium or any other elements ordinarily used may be made.

However, this is optional and is governed by the end product for which the metal is to be used. Upon solidification the ingots are removed from the molds and taken to the iron yard, where they are broken into pieces of suitable size for recharging into the cupolas for the second and final melting operation.

Brief examination of the analysis of the iron as tabulated in the foregoing will make evident the fact that this metal remelting can, by means of proper additions to the cupola charge, be made suitable in composition for use in castings of any metal section whatsoever.

Cupola Charge Additions

Where relatively low-silicon-high-carbon irons are required, relatively small additions of materials high in silicon will be added to the charge. No "raw" steel will be used, so the final product will be relatively high in total carbon content, that is, in the range of 3.45 to 3.90 per cent.

The manganese can be held at any desired level, and in some instances has been deliberately kept as low as 0.20 per cent. This was in iron for a number of large caustic pots (Fig. 10) where manganese in the normal range of 0.70-1.00 per cent was found to impart an objectionable color to the finished product.

Ordinarily, irons of low-silicon and high-carbon contents are used, as previously stated, for caustic pots, kettles, jackets, covers, autoclaves, ingot molds, and a variety of other shapes used in chemical industry. This type of iron is soft and resistant to heat shock as well as to grain growth in applications where castings are subjected to elevated temperatures (below the critical) for long periods of time.

When iron made by the foregoing methods is used for castings requiring a high-strength iron, the cupola charges will contain synthetic pig iron (previously described) and the percentage of steel deemed necessary for the class of iron being produced.

From the foregoing description of the process it is evident that any class of high quality iron can be produced, ranging in tensile strength from 25,000 to 60,000 psi. All operations relating to the cupola practice are carefully controlled and supervised as previously described.

One of the chief differences between irons produced by this method and the other irons produced lies in the fact that up to 1.5 per cent (30 lb. per ton) of zirconium silicide frequently is used as a ladle inoculant and, while this alloy ordinarily is looked upon as a graphitizer, additions even in very large quantities do not result in the formation of ferrite in the microstructure. Any of the alloys ordinarily used in cast iron may be used in iron produced by this method if and when the occasion demands.

DISCUSSION

Chairman: T. E. EAGAN, Cooper-Bessemer Corp., Grove City, Pa.

H. LOUETTE[1] *(written discussion):* If the "lighting-off" of the cupola has not been performed in a most satisfactory manner, we have found from experience in our foundry that about the third hour bridging occurs, effectively reducing the melting rate and producing metal with an apparent "spot" at the spout. When this condition prevails what advantage would the Balanced Blast Cupola have over the conventional type?

Secondly, is it possible to have a "spot" on the iron when the total carbon is in the range 3.45 to 3.50 per cent? We have found that this phenomenon occurs very frequently when the total carbon is in the lower range 3.20 to 3.30 per cent and entirely absent in the former.

MR. LEVI *(reply to Mr. Louette's written discussion):* I believe that the balanced-blast cupola would materially help in preventing the formation of the bridge which you state occurs after about three hours' operation. You also state that an apparent "spot" is visible on the iron at the spout. This "spot" as you know, appears while the iron is cooling off and usually starts to appear at about 2550° F.

Between 2550° F. and 2500° F. the surface of the iron will be made up of a combination of dark glossy (mirror-like) areas and brighter matte surfaces. These matte surfaces are what I believe you are referring to when you mention "spots." Below about 2500° F. the entire surface of the iron would have the matte appearance; just one big "spot," so to speak. Above 2550° F. the surface of the metal should be entirely free of "spots" and, as previously stated, would have a mirror-like surface.

The reduction in melting rate caused by the bridging of the cupola can also be indirectly, at least, a cause for the appearance of the "spots," especially if you tap your cupola intermittently. With this type of operation you probably allow the well of the cupola to fill up before tapping.

As the melting rate falls off, due to bridging, it requires progressively longer time intervals to fill the cupola well, and inasmuch as the molten iron in the well is losing temperature it is quite possible that you are tapping at a temperature

below that at which the "spots" start to form on the surface.

Even with the balanced-blast cupola it is possible to form a bridge if the materials charged are too long, or if insufficient coke and limestone are used. "Scaffolding" which can take place at any level in the cupola stack could not be prevented by use of the balanced-blast type of cupola.

In reply to the second paragraph of Mr. Louette's discussion I would say it is possible to have a "spot" on the surface of the iron when the total carbon is in the range of 3.45 to 3.50 per cent. We handle many tons of iron daily with total carbon in this range (and higher) and all of it is poured at such a low temperature that the "spot" is present on the surface.

I believe that the "spot" starts to form at higher temperatures as the total carbon (or at least the carbon equivalent) falls off and if both irons are melted at approximately the same temperature the "spot" will probably appear more frequently on the surface of the metal having the lower total carbon content.

Mr. Louette: We would like to have you answer whether a balanced blast cupola would have an advantage?

Mr. Levi: The balanced blast cupola does help keep any bridge or any overhanging slag and droplets of iron which freeze in front of the tuyeres melted off. Closing up the tuyeres creates a hot spot in front of the tuyeres. I think that you would probably have some advantage, because your bridging, of course, is just above the tuyeres. The balanced-blast cupola would not necessarily prevent "hanging-up" or "scaffolding" because this can take place at any level in the cupola stack.

H. E. Gravlin[2]: Did you have an experience with sizes of coke as well as coke types?

Mr. Levi: We have received some fairly small coke and found that the smaller coke gives us lower carbon pick-up. We have not had too much experience with it.

Mr. Gravlin: Do you put more emphasis on the size of your regular coke?

Mr. Levi: No, we take it as it comes.

Member: I was interested in the kindling tuyeres. How far below the regular tuyeres are they placed and just what are they?

Mr. Levi: I can not say how far below the regular tuyeres they are placed because I do not know how far your tuyeres are above the sand bottom, but they open right at the top of the sand bottom.

Member: What is to prevent the iron from flowing out?

Mr. Levi: We plug them up before we start the blast. Ours happen to be 4 in. dia. pipe which open right at the top of the sand bottom. We allow the draft to go through them until shortly before the blast is turned on, then we put a ball of mud into each kindling tuyere and ram it flush with the inside of the lining of the cupola. We have a cover which fits over the tuyere to hold the mud ball in place.

Member: You use 4 lb. of soda ash, is that per ton?

Mr. Levi: Yes, per ton of metal charge. If you put, say 6 one-ton charges in your cupola, you would use 24 lb. of soda ash.

Member: That is used with each charge?

Mr. Levi: That is right, put it right in with the limestone or whatever flux you are using.

J. A. Bowers[3]: I would like to know if there is much variation of carbon content in the same grade of pig iron received at different times?

Mr. Levi: We find some variation in our silvery pig. In 8 per cent silvery, we ordinarily figure about 2.6 per cent carbon. However, we have found it to run as low as 2.4. I do not believe it will have too serious an effect, especially if the percentage present in the charge is relatively small.

We do find some variations in the carbon on our regular pig iron. I figure the carbon in malleable grade of pig iron with 2.20 per cent silicon to be about 4 per cent. It may be a little under or over, but we do have some variations.

I have some figures on southern pig iron. In a carload of pig iron with 1.13 per cent silicon and 0.71 per cent phosphorus, we found 3.80 per cent carbon. The one in this particular group which is closest to that is 1.17 per cent silicon, 0.69 per cent phosphorus with 4 per cent carbon. That is the widest variation I have here. I do think, in running a total carbon on pig iron, that it may not be too reliable because there is so much graphite present, but we run them and it at least gives us an idea where we stand.

Member: In weighing your coke, is it under cover or is it out in the open?

Mr. Levi: We have it both ways.

Member: How do you figure it when it gets wet?

Mr. Levi: We do not make any extra additions. We use too much as it is.

T. R. Walker[4]: Have you used any graphite briquettes and what is your experience?

Mr. Levi: We have not used graphite briquettes so I can not tell you what the effect is.

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MALLEABLE FOUNDRY CONTROL

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PRODUCTION OF MALLEABLE IRON castings cannot be classified under any standard operations. Processes used in the industry vary greatly in all departments, including melting, core making, molding, cleaning, annealing and engineering. This paper mentions factors which directly affect the quality of the castings, and result in the elimination of unnecessary scrap, which is generally caused by improperly controlled manufacturing methods.

Melting

Pig Iron

The first requisite in purchasing pig iron or any other material is to establish reliable sources to be familiar with certain characteristics. The important factors to check on pig iron are: chemical composition, including harmful residual alloys; microstructure; and nature of manufacture.

Chemical Composition—Under chemical composition can be included analysis of carbon, manganese, silicon, sulphur, phosphorus, chromium, copper, molybdenum, and any of the other elements.

Microstructure—The microstructure of malleable pig iron has not been stressed much. The size and shape of the graphite particles should be of interest. We definitely have proven that the size and shape of graphite in silvery iron have a direct relationship to the carbon nodules in both malleable and pearlitic malleable iron. The appearance of abnormal amounts of phosphides in the silvery pig is detrimental to annealability of the iron.

Nature of Manufacture—Manufacturing variation from one year to another is apt to change the characteristics of the pig iron; therefore, affect the properties of the molten iron. It is well to check the blast furnace operations. Table 1 is a form which can be used for this check.

Briefly review this form, on quality of iron as affected by Blast Furnace Practice.

Steel Scrap

All steel scrap should be graded according to analysis, size and type, and piled so as to avoid danger of mixing.

Size and Shape—Good scrap should be free from excessive rust, any plated metal, and be of the proper size, shape and weight. Gray iron scrap is objectionable due to its effect on graphitization.

Chemical Analysis—In connection with the furnace charge, it is important that the scrap steel should be carefully examined for possible presence of alloys. Care should be exercised to avoid excessive amounts of residual elements, which will affect the graphitization of the metal in the hard iron and in the annealed state. Most important of these are: chromium, tin, and aluminum; of which chromium is the most liable to cause trouble, due to the fact that it occurs in many steels.

This paper covers what the author considers the best practice to control the quality of malleable iron castings, and to produce them at the lowest possible price consistent with good quality.

Some foundries have discovered that additions up to 0.002 per cent boron is effective in counteracting the carbide stabilizers of residual alloys, especially chromium.

With further reference to chromium in drop forge flashings or other shapes—if it is known that certain shapes are apt to contain chromium, they must be removed and segregated from the charge.

Fuel

Coke and Limestone—Coke and limestone for cupola operation should be purchased to conform to definite and uniform specifications. They should be checked, not only as to the analysis as in the case of coke, but to size, strength, and cleanliness.

Coke

It is recommended that when a satisfactory cupola coke has been obtained, a standard analysis be established, and all coke checked against that analysis.

It is also good practice to work close with the coke oven in regards coals used and coking time.

The following form is suggested for coke analysis:

Fixed Carbon
Volatile
Sulphur
Ash
Percentage Cellular Space
Moisture
B.t.u.
Weight per cu. ft. (Wet)
Screen Test—33.5% retained on 3½ in. mesh
85.0% retained on 2 in. mesh
Shatter Test—50 lb. from a height of 6 ft. four times; 36 lbs. (72%) shall be retained on a 2 in. screen

NOTE: The cellular space to be determined according to the following formula:

1. *Apparent specific gravity* × 100 = % cellular space true specific gravity

Table 1

QUALITY OF IRON AS AFFECTED BY BLAST FURNACE PRACTICE

PIG IRON—BLAST FURNACE DATA

1. Furnace No.....Location.....
2. Superintendent.....Date Produced.....
3. Composition of Ores:

Name					
Fe	.....	.....	.....	.....	.....
Phos.	.....	.....	.....	.....	.....
Si O_2	.....	.....	.....	.....	.....
$Al_2 O_3$	.....	.....	.....	.....	.....
CaO	.....	.....	.....	.....	.....
MgO	.....	.....	.....	.....	.....
Mn	.....	.....	.....	.....	.....
Per cent	.....	.....	.....	.....	.....
Pounds	.....	.....	.....	.....	.....

4. Coke Charge.....lbs. Coal Charge.....lbs.
5. Rounds per day..... Analysis:
6. Kinds of Coke: Fixed Carbon.....
   1. Shatter Test..... Vol. Hydro C.....
   2. True Sp. Gr..... Ash .....
   3. App. Sp. Gr..... Sulphur .....
   4. Porosity ..... Water .....
7. Burning Characteristics of Coke:
   1. Av. Temp. Rise.....
   2. Av. Time for 250° Temp. Rise.....
   3. Ignition Temp.....
8. Kind of Coal..... Analysis:
   Fixed Carbon.....
   Vol. Hydro C.....
   Ash .....
   Sulphur .....
9. Limestone Source..... Water .....
   Analysis:
   $CaCO_3$ ..... CaO ..... Lbs. of limestone per round..
   $MgCO_3$ ..... MgO..... Lbs. of limestone per ton.....
   Oxide .....
   SiO_2 .....
   .....
10. Blast Temp..... Max..... Min..... Cycle.....
11. Blast Pressure..... Av. lbs. per sq. in..... Max..... Min...
12. Air per minute.....cu. ft. .....lbs.
13. Fuel burned per hour.... lbs. ....lbs. per sq. in area at max. dia.....
14. Pounds of air per pound of fuel.... Air. per lb. of carbon....
15. Blast velocity through tuyeres .....ft. per min.
16. Composition of top gases....% CO....% CO_2....% H....%
17. Temp. of top gases.....°F.
18. Burden per hour....lbs.....lbs. per sq. ft. area at max. dia.....
19. Composition of slag.....(flushing slag)
    Si O_2 ....
    $Al_2 O_3$ .... Appearance of slag.....
20. Weight of iron produced per hour.....lbs.
    Weight per sq. ft. area at max. dia. per hour..... lbs.
21. Depth of metal in hearth at tapping..... ft.
22. Casting Temp.....max.
    .....min.
    .....av.
23. Time between casts.....hrs.
24. Dimension of furnace:

1. Stock line to tuyeres	.....ft.	.....in.
2. Tuyere level to hearth	.....ft.	.....in.
3. Size and number of tuyeres	.....	
4. Iron notch to conder notch	.....ft.	.....in.
5. Dia. of hearth	.....ft.	.....in.
6. Dia. at top of both	.....ft.	.....in.
7. Cubical contents of furnace	.....cu. ft.	

25. Number of stoves..... In use..... Heating up.....
26. Chemical Analysis:

AVERAGE OF FIVE PIGS

T.C.	C.C.	Si.	Mn.	P.	S.
.....	.....	.....	.....	.....	.....

27. Atmospheric temperature degs. F. .....max.
    .....min.
    .....av.
28. Barometric press.....in. Hg.
29. Humidity..... .....grs. per cu. ft.

NOTE: Analysis of ash:
Silica—per cent
Iron Oxide—per cent
Aluminum Oxide—per cent
Lime—per cent
Magnesia—per cent
Fusion point of ash—°F.

Limestone

The limestone shall be of such size that 90 per cent will be retained on a 1 in. screen, and no particle shall be in excess of 2 in.

Chemical analysis shall be as follows, computed on air-dry basis:

Calcium Carbonate.....	—% Min.
Magnesium Carbonate....	—% Max.
Silica	—% Max.
Iron and Aluminum Oxides.....	—% Max.
Sulphur	—% Max.
Phosphorus	—% Max

Ferro Alloys

These should be purchased on analysis, or from sources known to conform to specifications. Under this caption can be included: Ferro-silicon, ferro-manganese, silicon, and manganese briquettes.

Refractories—These should be purchased on a performance basis, and from reputable firms.

Molding Practice

Molding Sand

The nature of the molding sand will determine the necessary tests to maintain the properties of the sand. All incoming raw material used in the molding sand preparation should be carefully checked. It is advisable to keep close control of new shipments of sand, especially as to grain size and distribution. It is important that incoming sand be uniform from day to day.

Other materials on which controls should be maintained are: seacoal and rebounding clay. In general, coal for seacoal should not contain over 6 per cent ash and a minimum of 30 per cent volatile combustible matter. The sulphur content should be under 0.50 per cent.

Rebonding clays of various types can be controlled by using a predetermined per cent of clay with the optimum amount of water, and using A.F.A. 50-70 mesh standard sand.

Core Sand

Sand—In many plants the incoming core sand is the same as that used for molding sands. This is especially true in a mechanized plant using conveyor molding. Therefore, it is important that the core sand is carefully controlled. Cores in malleable foundries probably are the cause of as much trouble as any other single item, so uniform core sand is a prime requisite. Screen analysis and grain distribution must be checked periodically as incoming shipments.

Core Oil—All incoming core oil should be checked by any or all of the standard tests. The chemical laboratory should check the specific gravity, saponification number and viscosity. The sand laboratory should

run baked tensile and permeability tests on standard sand, mixed in the laboratory and baked in the laboratory oven. Also check the core mix by baking in the production ovens. Conditions of baking time and temperature should be standardized in the laboratory test. Other tests, such as: gas evolution, stickiness, effect on skin, etc., should also be made.

Flour—Incoming corn flour, or any other flour, should be checked for density, and for green compressive strength developed in an emperical mixture with standard sand. Another characteristic of flour is that, if the cores are blown the core mixture have excellent "blowability". Our experience has been that lighter flour will blow better than the heavier ones. Some arbitrary figure for density of the flour should be set up. The author uses a maximum of 13 lb. per cu. ft.

All other material used in the core room, besides those mentioned should be subjected to test. These include bentonite, iron oxide, etc.

Annealing

Heat Treating

The raw materials used in the heat treating of malleable castings include—annealing pots or containers, refractories and fuel.

Annealing Pots—There are generally two tests sufficient for annealing pot control—a chemical analysis for the desired elements and an oxidation and growth test of the metal. The chemical analysis of an iron pot will vary with different sources of supply. The oxidation test can be conducted on a 3x6x½ in. plate of the pot material, weighed before and after each trip. A standard maximum percentage loss in weight can be used for control. A growth test can be conducted on a 1 in. square, 1 ft. long bar accurately measured before and after one or more trips through the oven.

Refractory Blocks—Refractory tests are often made if annealing car blocks are produced from certain mixtures at the foundry where they are used. A minimum strength test is specified.

Fuel—Fuel for annealing, if oil or coal is used, can be analyzed to advantage to maintain uniform operation.

Operation Control of Processes

Melting

Irrespective of what method of melting is employed, some limits of chemical analysis must be set up for control purposes only.

The two principal conditions to meet in the production of a salable iron are: (1) the ability of the iron to have the proper strength, and (2) good machinability after heat-treatment. To pass this test, it is usually necessary for a test bar produced from the heat to show certain minimum physical properties, and a test sprue of given dimensions to show little or no primary graphite.

Fracture Testing in Hot Iron—This should produce a fracture in the casting comparable to the test sprue. A test sprue about 8 to 9 in. long; about 1¾ in. at one end and 2 in. at the other should be satisfactory for medium heavy castings.

It is true, that there are other conditions besides the nature of the iron, which can produce mottled iron castings. The cooling rate of the casting in the mold directly effects the mottling condition. It is necessary to control pouring conditions, iron temperature and the many other variables which directly effect the conditions of the iron

While test bars are a valuable adjunct in the control of quality, the author knows of no better test than to periodically break castings under average conditions and inspect for mottle and soundness of the casting.

Chemical Composition—For close control, preliminary chemical analysis should be taken for carbon, silicon and manganese. The time and conditions under which the tests are poured is a matter to be taken up in each case, depending upon whether the operation is continuous or intermittent. Sufficient tests should be taken to guarantee a uniform iron going into the mold.

Final analysis of each heat, or at certain intervals throughout the day, should be run on the regular elements. Since chromium is considered an undesirable element—once a day, or more often, this element should be run and maintained at a predetermined maximum.

Temperature—Temperature control should be maintained at the melting unit and throughout the pouring operations. Good pyrometers are available, and there is no reason for not installing adequate temperature control equipment. It will pay big dividends.

Blast—Blast control of both volume and pressure is necessary in the successful operation of a cupola in duplexing. In air furnace melting—to maintain proper furnace atmosphere—the atomizing or airing of the fuel and air calls for certain standardizing and control of the mixture.

Humidity and Hot Blast—In cupola duplexing, both hot blast and humidity control are a valuable addition to the operation to insure maximum uniformity.

As you increase the number of controls necessary to produce the best iron, an increasing personnel is required to carry out the operations, but this is usually justified.

Repair of Cupola—When the cupola is used in the manufacture of malleable iron, either as a single unit or in combination with air or electric furnaces, the control of the cupola operations has a major effect on the quality of the final product. The following factors are important to control: Construction of the cupola, which includes—ramming of the bottom; building of the well; construction of the tap hole and slag hole; tuyere opening; and proper relining of the melting zone. It is important that the height of the coke bed and the order and nature of the charging be standardized according to a predetermined practice. Not enough emphasis is placed on this part of the operation.

Air Furnace—In the operation of an air furnace, there are numerous important rules to follow to produce uniform metal. The repair of the furnace must be properly made and maintained. This includes the bottom, tap hole, side walls, roof, bridge wall, and the throat of the furnace leading to the stack.

The charging procedure should be standardized and maintained according to predetermined good practice. Fuel consumption should be checked, and be under constant control. Other controls include: the standardizing of melting down; superheating; tapping; together with proper transferring of the iron.

Electric Furnace—The use of the electric furnace is generally as a

holding furnace. Following are some of the details to be controlled: the bath level should be kept constant and either slag free or under controlled slag; the electrical equipment must be in perfect order to hold carbon pick-up to a minimum; records should be kept of Kwh per ton, electrode consumption, refractory life and any other factors affecting the operation of the electric furnaces.

Ladle Practice—All pouring and transfer ladles, and forehearths, should be of standardized design and construction, in order to obtain the longest life, and produce the highest quality iron. The following details should be considered: (1) mixing of ladle lining material, (2) ramming of ladle lining material, (3) thorough drying and preheating to avoid the presence of moisture.

Molding Practice

Molding Operations—It is important to produce consistently a mold of the same hardness. Therefore, automatic jolting is one factor which should be advantageous, to eliminate the variation in the number of jolts. The proper amount and duration of the squeeze is another variable which should be controlled.

Molding Sand—Sand should be checked periodically for moisture, permeability and bond strength. The temperature of molding sand is important in order to prevent the sand from sticking to the pattern and producing dirty castings. The temperature of the sand should be held to a minimum. Other control tests to be made daily, or as often as necessary, are: dry compressive strength; green tensile strength; sand distribution curves; and per cent of seacoal and clay.

Equipment Control—On a mechanized molding unit, it is important that, all operating details are closely controlled. Some of these can be listed as follows:

1. Condition of muller and muller blades
2. Clay and seacoal feed arrangements
3. Regulating the amount of spill sand
4. Condition of revivifier
5. Condition of screening equipment — also magnetic equipment

Seacoal Preparation Unit—If seacoal preparation is carried out in the foundry, it is important that the seacoal be crushed, so that the fineness meets a given specification. A coarsely ground seacoal has many detrimental effects.

Core Making

Once a satisfactory core mixture has been worked out for a particular job, the physical properties should be determined through standard methods, and the process checked regularly.

Dried Sand—The first step in maintaining a uniform mixture is to start with a dry sand. If the company operates its own drier, a certain amount of attention will be required to maintain the proper drying control.

Operation of Mixers—In the operation of the mixers, it is important to weigh carefully or measure all ingredients. If this is properly done, the duplication of a given core sand mixture calls for standardized mixing practice. This can be accomplished by controlling the sequence of additions to the mixer, and total mixing time.

Properties of Mixtures—The properties of the core sand mixtures should be determined in the green and baked condition. Usually the batches should be tested for moisture, green compressive strength and baked tensile strength. The baked tensile strength test should be carried out in the standard laboratory oven, and at least once a day in each production oven. This will give a check on both mixtures and baking conditions.

Operation of Core Ovens—The operation of the core ovens should be constantly checked as to the time of heat, temperature, loading and other operating details.

Heat Treating

Temperature Control—There are many good heat-treating practices. They all vary to a considerable extent, depending upon the type of iron used, and the heat-treating equipment available. The temperature throughout the charge must be uniform for good annealing.

Good pyrometry equipment is necessary, and frequent checking of thermocouples against tested standards must be constantly carried out. Once a satisfactory annealing cycle is used, it should be maintained.

Combustion and Atmosphere—In operating radiant-tube atmosphere control ovens, it is necessary to run gas analyses in order to maintain proper atmosphere. Gas analyses should be periodically taken to maintain proper combustion of the fuel, used in heating the tubes.

Life of Pots, Trays, Stools—The material cost for this part of the operation runs high and demands constant checking and control to hold cost per ton to a minimum. Accurate records of the material in use, and new materials on test, should be kept.

Control of Loading—In the operation of controlled atmosphere ovens of the continuous type, uniform loading by weight of the trays is important. To a lesser extent this is also true of the car type ovens.

Test Lugs and Test Bars—Some customers demand test lugs to be placed on castings. In this case, it is well that the size and shape of the test lugs be standardized for various casting sections.

Use of the regular round tensile bar is principally for controlling the uniformity of the metal from day to day. For control of annealing operations, some plants also use a ¾ to 1 in. square bar, 6 to 8 in. long. Others use a wedge-shaped bar. Some bar should be used to indicate by bend or fracture, the nature of the anneal.

Processing

Tools—There is a certain amount of control required to maintain proper equipment in the processing of the castings. If shear knives, milling cutters or broaches are used in the removal of gates, the quality of these tools should be checked.

Cleaning—The wheelabrating and milling of castings should be studied, for the purpose of obtaining a satisfactory, cleaned casting in the least time. The control of these operations is a function of the surface condition of the castings, together with the type of equipment available.

Welding

Some plants make a practice of welding defects on annealed castings. All welded castings must be either completely re-annealed or given an 8-hr. temper at 1350° F.

The control of this operation is important, so as not to allow hard welds to be put in service.

Engineering

The first requisite in the production of a sound casting is to engineer and gate the job properly. Once this is established it should be adhered to, and no changes made unless the job shows excessive scrap, such as: cracks, dirt, etc. A record control of the original layout, and any changes made, should be kept, so that if the job is out for a period of time, it can be replaced with complete information at hand.

Some plants take photographs or make sketches of all new jobs as soon as a satisfactory gating has been developed. Cores and chills are also sketched and photographed. A case history card of each job, with the following data, serves as valuable information for the particular job:

Data on molding sand mixtures.
Data on core sand mixtures, including properties.
Analysis of scrap, as to cause and percentage.
Weight of iron in mold.
Weight of rough casting.
Yield of good casting.
Changes made in pattern equipment on gates and risers.

Inspection

Inspection control can be augmented by the use of x-ray or radium. Both can be used to develop proper gating. Numerous plants use magnetic particle methods to inspect castings for cracks. Cutting up castings in both the hard and soft state and etching the pieces in acid, is probably the most satisfactory method of checking the soundness of a casting.

DISCUSSION

Chairman: C. F. JOSEPH, Saginaw Malleable Iron Div., General Motors Corp, Saginaw, Mich.

R. A. LODER[1]: Is it economical to use test lugs?

CHAIRMAN JOSEPH: During the last ten years in our plant we have eliminated test lugs, due to the fact that we have felt it better to control the metal than to control the casting. Test lugs are unnecessary, if you have accurate control of the raw material, melting process, and the annealing process.

We do not have test lugs on our castings. Most plants that make large castings, do put test lugs on castings. If they put test lugs on the castings, those test lugs should be of the proper shape and design. Very often the round test lug is put on the casting or is put in a place where it does no good.

Years ago, when we used test lugs, they were put on certain parts of a casting where they did not mean very much, and they were the wrong shape such as a little round test lug in place of a long slended test lug. Then we used to make the test lug so it would snap off easily, rather than take a bend. There was a sharp corner at the bottom of the test lug rather than a fillet.

I think it is a matter of policy whether to put test lugs on for the control of the product. The better your control in processing the metal, the less necessity there is for test lugs.

E. M. STRICK[2]: On castings of certain design do the test lugs give a true picture of the product?

CHAIRMAN JOSEPH: I do not say that a test lug should not be used on certain castings. We do not run test lugs for a number of reasons. They cost money to break off for one thing, and if you avoid anything that costs money in the making of a casting, it is one step in the right direction. If your practice is what we call sloppy and lacks accurate control, then I would think test lugs are very important. A lot of companies specify test lugs unless you can arrange with them to omit the test lug.

I have seen a lot of test lugs broken off in a foundry, where the man just knocked the test lug off and there was no visual examination made of the test lug. In that particular case, the test lug is useless, unless it actually is examined to evaluate the fracture.

MR. STRICK: You want the test lug to bend to denote the ductility of the metal, and if it was not struck in correct manner it sheared off leaving a lustrous, steel-like fracture that was useless as far as visual examination was concerned?

CHAIRMAN JOSEPH: That is right. Also if a test lug is broken hot it will appear satisfactory. If it is broken cold, it will not appear so good. So there are a lot of factors that govern the proper breaking off of a test lug.

DAVID TAMOR[3]: Have you ever seen test lugs break white where the fracture of the casting is normal?

CHAIRMAN JOSEPH: Yes, I have seen the test lug break white when the fracture of the casting was normal in the days when we had periodic box-type annealing ovens.

MR. TAMOR: Then to what do you attribute that white fracture?

CHAIRMAN JOSEPH: I think generally a thin section will break whiter than a heavier section. A heavy section will show a tendency to break white on the compressive side of the break.

J. E. REHDER[4]: Speaking of control in operations, I understand it is sometimes the practice in an electric furnace to keep the doors open or not to keep them open. Do you classify that as atmosphere control over the iron?

CHAIRMAN JOSEPH: It is true that in the operation of the electric furnace, whether it is for malleable iron or whether it is for gray iron, if you have a reducing atmosphere over the bath, you are more apt to mottle the iron, or it will gray easier, at a lower total carbon and silicon content. If you leave the doors open and have an oxidizing atmosphere, you can run a higher total silicon and carbon and not have primary graphite show in your casting. It is a matter of oxidation, a matter of the atmosphere over the bath.

For instance, in malleable practice we leave the door on the electric furnace open, and in the manufacture of a gray iron casting you should keep the furnace door closed or keep a slag on the bath. If you keep a slag on the bath in the making of malleable iron, you will have to run your total silicon and carbon at a lower total than if you keep the bath clean.

MR. STRICK: In controlling analysis of iron to minimize primary graphite on a heavy section casting in the cupola melting practice for malleable iron is it necessary at times to ladle alloy the iron?

CHAIRMAN JOSEPH: Do you mean where you make straight cupola or cupola electric iron?

MR. STRICK: I mean the cupola electric iron.

CHAIRMAN JOSEPH: We have not used alloy for years. There was a time when we did use a small amount of tellurium on a test we ran on one or two jobs, but we do not use alloy for producing white fracture in heavy sections. We produce iron to make the heaviest section that we fabricate. If we have to run the carbon, silicon or total carbon and silicon low, we just have to do that, but we have a certain maximum cross section which we can produce satisfactorily.

It is not always the cross section of the casting that gives the mottle. It is the cooling rate in the mold. It is not that you standardize on a 1½-in. section or a 1-in. section. You naturally must examine the casting that you are producing; fracture and correlate that casting with your test sprue results.

MR. STRICK: In cupola electric practice or cupola air-furnace practice, it is possible then to pour any section of malleable casting within standards?

CHAIRMAN JOSEPH: I would not say you could pour any section. You would naturally have to standardize on certain jobs. For instance, if you should get a job into your foundry that would graphitize, i.e., give you primary graphite in a heavy section of that casting, you would either have to use a chill on that casting, put on mottling fins, or take some other steps to make that casting, unless you wanted to change your iron analysis.

For instance, in our particular plant we would not want to change the iron just to take on a heavy job. If we had a heavy job come in that would graphitize in the mold, we would try to make that some way or have it redesigned or do something about it before we would do anything to the iron. The iron should be the same from week to week or month to month, unless you are taking on a job which is going to run for a long time.

[1]Erie Malleable Iron Co., Erie, Pa.
[2]Erie Malleable Iron Co., Erie, Pa.
[3]American Chain & Cable Co., York, Pa.
[4]Grinnell Co., Ltd., Toronto, Canada.
[5]St. Ayme M. Remy, Herstag, Belgium.

Then you might want to make some drastic change. Even then it would be critical to jeopardize the entire run of iron for just one job.

MARCEL REMY[5]: Would tellurium help in this particular case?

CHAIRMAN JOSEPH: It would. I believe some have used tellurium to a great extent in heavy sections.

MR. TAMOR: We have used it but not with good results. Tellurium interferes with the graphitization of white iron so we do not use it.

CHAIRMAN JOSEPH: Of course, we did use tellurium with the addition of a certain amount of silicon, but even then we did not believe we eliminated the primary graphite. We always had some minute primary graphite form. The fractures did not show white. They showed a grayish cast, and you could not see the large graphite particles, but they were still there under microscopic examination. If you wanted to cut a section and look at it, it probably showed some small particles of graphite.

MR. TAMOR: Would you like to add something in the way of slag control? I am thinking particularly of duplex malleable, where we are melting lower toward the tuyeres than for gray iron practice.

CHAIRMAN JOSEPH: Do you mean slag control in the cupola?

MR. TABOR: I mean observing the slag in the cupola at ½-hr. intervals for color viscosity, and whatever controls are necessary.

CHAIRMAN JOSEPH: Naturally the most important thing in the slag is to have the slag flow properly from the cupola. That is usually controlled by the flux that you use, the amount of flux, and the raw material condition that enters the cupola.

We have run over a period of time slag control methods by color, by analysis, but it is not necessary, if you properly control material going into the cupola, fluxes, and the various melting operations. I do not know of anybody definitely that is running viscosity tests on slag. We have run some chemical analysis tests for iron oxide, magnesium oxide, and SiO_2. We have also established certain colors that we like to have, but those colors vary.

I do not know of anybody that is controlling cupola slag day after day. Periodically they may, to see what their slag looks like or to get some idea of the analysis of the slag which they desire to maintain.

APPLICATION OF HARDENABILITY REQUIREMENTS IN THE STEEL FOUNDRY

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THE REMARKABLE CONTRIBUTIONS of the steel foundry industry during the war in solving many of the critical problems arising from the acute shortage of manufacturing facilities in many fields of industry, in particular in the forging industry is, indeed, a credit to the steel foundry metallurgist. The successful manufacture of castings calling for tensile requirement of 150,000 to 200,000 psi. as substitutes for forgings and fabricated assemblies is proof.

In analyzing the fundamentals underlying this progress, we believe it to be the result of the extensive research into the proper utilization of alloys and the application of heat treating methods, including liquid quenching.

Although considerable basic metallurgical knowledge in this field was available as a result of work carried out in the wrought steel industry, its application in the foundry has presented many problems. As the greatest advantage to the engineer in using castings lies in the almost unlimited possibilities in realizing the ideal design, castings frequently are of much more intricate shape than forgings and, therefore, more susceptible to excessive quench-stresses which may lead to cracking.

The nature of a cast metal in itself also tends to exaggerate this condition because of its inherent dendritic structure producing planes of weakness, in particular where abrupt changes of section occur.

Presented at a Steel Session of the Fiftieth Annual Meeting of the American Foundrymen's Association at Cleveland, May 10, 1946.

In spite of these difficulties, the foundry industry proved itself fully capable of fulfilling the extreme demand placed upon it by a nation at war. Wartime conditions with their steady call for ever-increasing tonnage, regardless of cost, are, however, a far cry from the rigid competitive basis which forms the foundation of our peacetime manufacture and to which we now must return, if not immediately, certainly within a few years time.

If the foundry industry is to retain the gains made, and to further increase the field of application of castings for highly stressed parts, in particular in the automotive and allied industries, during the postwar period, consideration must be given to certain requirements of these consumers.

Of, perhaps, the greatest concern to these industries today is the present method of the foundries in evaluating the properties of castings.

Whereas the wrought steel manufacturers have standardized their products by grading the steels according to their chemical composition and supplemented this by pertinent information relative to the hardenability of each grade, thereby guaranteeing reproducible properties with a given heat treatment, the casting manufacturers are offering their products on the basis of physical properties only as determined by tensile test of a 0.505 sq. in. specimen prepared from a coupon of about 1 to 1¼ sq. in. area.

The author defines hardenability and discusses the Jominy end-quench hardenability test and the mechanism of hardenability. He also discusses factors affecting transformation of austenite to martensite, hence hardenability of the steel. Since hardenability requirements are being incorporated in steel casting specifications foundrymen should become more familiar with this property.

The consumer, therefore, has the choice of selecting a forging or a fabricated structure for which a material of known chemistry and hardenability characteristics can be specified, or a cast design which, although it may suit the particular part better, only can be purchased on a basis of physical properties as determined above with the chemistry left to the option of the foundry and the hardenability characteristics unknown, or at least uncontrolled.

Specifying Castings

The method of specifying steel castings on the basis of physical properties obtained on 0.505 test bar may be entirely satisfactory when it represents a casting requiring only annealing or possibly normalizing. In the case of a casting, however, that must be transformed into martensite, either through liquid quenching or if highly alloyed by some less drastic heat treatment, in order to produce the properties required, it is evident that, although the 0.505 bar may exhibit the required minimum physical properties, there is no guarantee that these are reproduced in the casting.

Because of the design the casting may have a cooling rate much slow-

er than that of the test coupon and entirely insufficient even to produce martensite.

The test bar results alone are, therefore, not sufficient to judge properly the castings which they represent, although they serve as an excellent indicator of the quality of the particular batch of metal and its response to a predetermined heat treatment.

The reluctance of the foundries to accept the chemical composition as part of the specification is quite understandable when we consider, first of all, the lack of standardization in melting equipment prevailing in the foundry industry as compared with the wrought steel industry, and, also, the fact that many foundries specialize in a particular class of castings and have, after years of experimentation, worked out compositions most suitable to their individual type of work.

The result of the conditions described is that most firms in the automotive and allied industries have set up their own specifications, usually based on both chemical analysis and physical properties, thereby limiting the number of sources available as many foundries which may be perfectly capable of producing the part in question, but are not prepared to add a new composition in their shop.

Proposed S.A.E. Steel Casting Specification. In an attempt to arrive at a specification which will serve the needs of these industries, and, at the same time, be acceptable to the foundry industry, the S.A.E. about a year ago organized a committee composed of men representing both the consumers and the producers to study the problem. A short review of the recommendations made by this committee may be of interest in view of the fact that hardenability has been incorporated in the specification of high strength steels.

A survey made of present consumer specifications indicates the complete lack of standardization existing, with many of the requirements almost identical and frequently overlapping. This survey contained data accumulated from sixteen of the largest users of steel castings in the automotive and related industries.

As a specification broad enough to cover each and every requirement shown would be both cumbersome and impractical, it seemed evident that the best approach would be to divide the steels into the

Table 1

GRADES OF STEELS AND REQUIREMENTS OF PLAIN CARBON AND LOW ALLOY CAST STEEL

SAE Grade	Approx. Equiv. ASTM	QQS-681b	Description	Chemical Analysis C	Mn max.	Si max.	P max.	S max.	Residual Alloy Limitations	*3 Min. Tensile Properties T.S. psi.	Y.P. psi.	Elon. in 2 in. per cent	R.A. per cent	B.H. *4	Heat Treat *5
0022	A-27 Rev.		Low carbon cast steel suitable for carburizing.	.12/.22	.60 *1	.60	.05	.06	Residual alloys shall be limited by maximum core hardness required subject to specific agreement between producer and purchaser in inquiry and order.	Not Required					A or N or N-T
0030	A-27 Rev.	1	General purpose cast steel; welding grade.	.30	.70 *1	.60	.05	.06	See *2	65000	35000	24	35	131	A or N or N-T or Q-T
0050		3	Medium high carbon cast steel suitable for castings requiring high surface hardness.	.40/.50	.50/.90	.20/.60	.05	.06	To be specified by max. hardenability values when residual alloy limitations are required.	85000	45000	16	24	170	N or N-T
										100000	70000	10	15	207	Q-T
080	A-27 Rev.	3	Carbon or alloy cast steel (usually of lower C content than type 0050)	Not Required			.05	.06		80000	40000	18	30	163	A or N or N-T or Q-T
090	A-148 Rev.	4D2	Alloy cast steel.	Not Required			.05	.06		90000	60000	20	40	187	N-T or N-Q-T

*1 For each reduction of .01% carbon below the max. specified, an increase of .04% Mn. above the max. specified will be permitted to a max. of 1% Mn.
*2 Residual alloys, the determination of which shall be matter of agreement between the manufacturer and purchaser, shall not exceed the following limits: Copper, max. per cent—0.50, Nickel, max. per cent—0.50, Chromium plus Molybdenum, max. per cent—0.25, Tungsten, max. per cent—0.10 —Total content of these unspecified elements, max. per cent—1.0. For each 0.1 per cent below this specified max. alloy content of 1.0 per cent an increase of 0.02 per cent chromium plus molybdenum content and 0.06 per cent nickel and copper content above the specified max. will be permitted. No change is recommended on the other chemistry or the carbon-manganese relationship.
*3 These values to be obtained from coupons of 1-1¼ in. cross sections. Tempering temperatures for the castings to be adjusted for the required Brinell Hardness to compensate for variations of mass between casting and test bar.
*4 These values are applicable to casting sections not over 3 in.
*5 Explanation of abbreviations: A—Anneal, N—Normalize, N-T—Normalize-Temper, Q-T—Quench-Temper.

Table 2

HIGH STRENGTH STEEL CASTINGS FOR STRUCTURAL PURPOSES

SAE Grade	Approx. Equiv. ASTM	QQS-681b	Description	Chemical Analysis C	Mn	Si	P max.	S max.	Hardenability Factor H	*3 Min. Tensile Properties T.S. psi.	Y.P. psi.	Elon. in 2 in. per cent	R.A. per cent	B.H. *1	Heat Treat *2
0105	A-148 Rev.	402	Alloy Cast Steel	Not Required			.05	.06	Consult Table 3 for H values	105000	85000	17	35	217	N-Q-T
0120	A-148 Rev.	403	Alloy Cast Steel	Not Required			.05	.06	Consult Table 3 for H values	120000	100000	14	30	248	N-Q-T
0150	A-148 Rev.	404	Alloy Cast Steel	Not Required			.05	.06	Consult Table 3 for H values	150000	125000	9	22	311	N-Q-T
0175	..	..	Alloy Cast Steel	Not Required			.05	.06	Consult Table 3 for H values	175000	145000	6	12	363	N-Q-T

*1 These values are applicable to casting sections not over 3 in.
*2 Explanation of abbreviations: N-Q-T—Normalize-Quench-Temper.
*3 These values to be obtained from coupons of 1-1¼ in. cross sections. Tempering temperatures for the castings to be adjusted for the equivalent Brinell Hardness.

minimum number of basic groups, each containing the essential classes needed to permit maximum degree of standardization throughout the industry. Accordingly, the following method of grouping was adopted:

1. Plain carbon steels which, except for the carburizing grades, are specified by both chemical composition and minimum tensile properties necessitated by their specific applications,

2. High strength steels specified by minimum tensile properties only, for miscellaneous uses where requirements do not justify hardenability control,

3. High strength steels specified by minimum tensile properties and hardenability requirements.

Tables 1 and 2 show the different grades composing each group and their requirements. Although the majority of present consumer specifications for high strength steel castings include both chemical and physical properties, it was felt that the chemistry could be left out of this specification by introducing minimum hardenability requirements.

Accordingly these steels were divided into the two groups, one to cover the field in general and limited by minimum physical properties only, and one for critical castings as well as for castings machined and heat treated by the consumer and covered by physical properties coupled with hardenability.

Fiber Test

The manufacture of castings to hardenability requirements is not new to the industry, at least not to those foundries who during the war were engaged in the manufacture of armor castings. It here took the form of the so-called fiber test which was introduced by the Army to insure that the castings were fully quenched out.

For those that are not familiar with this test, the same consists of hardening and tempering to the required Brinell hardness a sample representing the average cross-section of the casting and fracturing the same by impact. This test is based on the knowledge that a fully quenched out steel tempered to within a certain hardness range exhibits a fibrous fracture when broken by impact; whereas a steel not fully quenched out exhibits a crystalline fracture caused by precipitation of ferrite.

The hardenability requirements in the proposed specification are based on the Jominy end-quench method, and thereby permit the consumer to specify the degree of hardenability desired as will be discussed later.

It is fully realized that the incorporation of hardenability requirements in the foundry where the parts made attain their final shape right in the mold, presents certain problems not encountered in the steel mill. In the case of an off-heat in the foundry, little can be done but to scrap it; whereas in the steel mill where the metal is poured into ingot form, an off-heat can be diverted to some less important part prior to rolling. With careful control of all pertinent factors, however, the well regulated foundry will not be unduly handicapped by these requirements.

Before proceeding with our discussion of the application of hardenability requirements in the steel foundry, it may be in order to review some of the fundamentals underlying these inherent properties in steel termed hardenability.

Mechanism of Hardenability

Hardenability may be defined as that property of a steel which enables it to resist transformation to the stable soft end products, ferrite and carbide, upon cooling from the austenitic stage. The degree of hardenability is measured by the percentage of the intermediate hard product, martensite, formed with a given cooling rate, regardless of just what maximum hardness this martensite may achieve.

Although there are many applications, as will be discussed later, for shallow hardening steels in which a partially hardened or soft interior is an advantage, it is essential to have sufficiently high hardenability to produce martensite whenever optimum strength and ductility throughout a section are to be developed. The size of section which can be uniformly heat treated to initial high hardness and the depth of hardening in larger sections are, therefore, controlled by the minimum cooling rate which will produce martensite.

A. Factors affecting hardenability in plain carbon steels.

The formation of martensite, however, depends upon the steel being cooled rapidly enough to prevent the austenite to ferrite-carbide reaction, which reaction in turn is a matter of the nucleation rate for the formation of ferrite and carbide nuclei. Any change which retards this nucleation rate contributes to slower critical quenching speeds and, in turn, to increased hardenability. The nucleation capacity is, therefore, the actual controlling factor in hardenability.

As the ferrite and carbide nuclei are largely located around the austenitic grain boundaries, it is evident that the size of the austenitic grains play an important roll, and that the smaller the grains the larger the total boundary surface and, therefore, the greater the number of nuclei. As a result, a fine grained steel requires a greater quenching speed in order to produce the same per cent martensite as a coarse grained steel of the same chemical composition.

Hardenability Factors

From this it follows that the chemical composition alone does not entirely control the hardenability, but that other factors which influence the austenitic grain size, also must be considered.

Of particular importance to the foundryman are the effects of various deoxidizers used in making the steel. Frequently aluminum is added as a deoxidizer, and many foundrymen think of aluminum solely as a control for porosity and neglect its profound effect in increasing nucleation. The presence of a few hundredths of a per cent retained aluminum will greatly lower the hardenability.

Austenite, like most metals, is subject to grain growth and coarsens with higher temperature, with the result that the nucleation rate decreases and the hardenability increases. Identical samples heat treated to varying temperatures for austenitizing followed by quenching will, therefore, display different hardenability depending upon the temperature and the resultant austenitic grain sizes.

B. Alloy steels.

The increase of hardenability

through the transformation of a coarse austenitic grain, although important, cannot be applied in practice to any great extent. It, instead, becomes necessary to change the composition of the austenite through the addition of alloying elements which all, with the exception of cobalt, will retard the transformation and, therefore, contribute to the hardenability. The degree differs with the element in question, some having greater effect than others.

The extent to which the alloys are uniformly dissolved in the austenite, that is the homogeniety of the austenite, is of great importance. In particular is this the case when sluggish alloys, notably chromium and other carbide formers, are used which require considerable soaking time and high temperature in order to fully dissolve.

Carbide-Forming Elements

If sufficient time or temperature is not allowed, these alloys will be only partially dissolved with the resultant decrease in hardenability under that anticipated, aside from the economic waste. The carbide forming elements are, furthermore, natural grain growth restrainers as long as some portion of the fine carbide particles remain.

These facts are of great importance to the foundryman, especially when he utilizes alloy steels containing such elements, and care must be exercised in the selection of suitable heat treat temperatures and soaking time.

In addition to the alloys added purposely, most steels contain small amounts of residual elements such as boron, arsenic, tin, nitrogen, etc., all of which affect the hardenability. These elements, with the possible exception of nitrogen, are present in the charge and are particularly difficult to control with the current scrap situation.

Non-metallic inclusions formed through the addition of deoxidizers such as manganese, silicon and aluminum and which are retained finely dispersed in the steel, also adversely influence hardenability in that they restrain grain growth.

The foregoing discussion has dealt with the factors affecting the transformation to martensite and the depth of hardening. It is to be remembered, however, that martensite is a decomposition product of austenite and that its physical characteristics are largely dependent upon its carbon content, that is to say, that the intrinsic hardness of the martensite formed is primarily a function of the carbon present; although the composition as a result of alloys has a certain limited influence, for all practical purposes the above is the case.

The ultimate tensile properties obtainable, being largely a function of the hardness, are, therefore, controlled mainly by the carbon content, and the depth to which the steel can be transformed to produce these properties is a function of the alloys plus residual elements.

Hardenability Requirements

The primary object of introducing hardenability for high strength steel castings is, therefore, to insure that the castings are fully quenched out to the required depth, in other words, transformed to martensite. Only when this assurance is at hand do the test bars become representative of the castings, thereby allowing proper interpretation of the same.

The ultimate object, however, goes further than this and concerns itself with the need for some means of guaranteeing reproducible results in identical castings made from different chemistries and heat treated alike. Frequently a consumer is forced to engage two or more suppliers in the manufacture of a given casting with the result that the heat treat department may receive identical castings made from two or three different chemistries.

The confusion that this is apt to lead to is readily understandable when we consider that although the same quenching temperature may be applied, each chemistry usually requires a different tempering temperature in order to produce the same properties. The necessity, therefore, of keeping the castings segregated according to their chemistry throughout the various stages prior to heat treatment makes such an operation undesirable. This condition, undoubtedly, has also contributed to the inclusion of chemistry in present consumer specifications.

Complete reproducability may be difficult to obtain because of the influence of varying alloying elements upon resistance to softening with given tempering temperatures. In particular do steels containing strong carbide-forming elements such as chromium and molybdenum require higher tempering temperature or longer tempering time to soften to a given Brinell hardness than, for instance, steels containing manganese where the alloy is mainly dissolved in the iron and steels containing nickel which has relatively little effect. Additional factors such as melting and deoxidation practice as well as segregations, etc., are also significant.

Effect of Alloys

Evidence is, however, at hand to show that a combination of various alloys in cast steels have little if any effect upon the mechanical properties providing that the steels are fully quenched out. One may, therefore, conclude that various types of alloy steels of similar hardenability and heat treated to the same hardness will exhibit similar properties. As the question, however, is one of reproducing the properties with similar heat treatment, not only the type of alloys present but also the carbon content must be considered.

It may, therefore, in the final end be necessary to grade the steels according to carbon content and possibly according to the type of alloys, namely carbide forming or non-carbide forming. The effect of the latter, however, can probably be minimized if care is exercised in selecting the proper time necessary for tempering.

Because of the radical departure in the proposed S.A.E. specification from previous such writings, it was deemed advisable for the present to introduce only the basic hardenability requirements necessary to enable the consumer to specify the degree of hardenability desired for a given casting thereby assuring that not only the test bar but also the casting will meet physical requirements.

At some later date when the foundries have had the necessary time to thoroughly familiarize themselves with operations to hardenability requirements and much additional data not yet available has been collected and evaluated, it is hoped that a hardenability specifica-

tion for steel castings equal or more comprehensive in its application than the specification at the present time used in the wrought steel industry can be written.

Although the simplest solution in introducing hardenability requirements for steel castings may on the surface appear to be the adaptation of some of present wrought steel hardenability bands, the following considerations make it evident that such is not the case. These bands are the result of an evolution which started with the construction of bands representing the maximum and minimum limits of given chemistry ranges.

The field obtained in such a manner was divided into three sections and the middle third adopted in actual operation. It was, however, soon found that about twenty per cent of the heats made fell outside this range. As a result, the bands were widened by increasing the carbon by 0.02 per cent over the upper middle third limit and lowering the carbon by 0.02 per cent under the lower middle third limit. This modification resulted in approximately 97 per cent of all heat tabulated falling within the limits. The bands, as finally adopted, were set a shade wider than the above.

When we therefore consider the fact that the present bands are based on modifications of given maximum and minimum chemistry limits and, as a result, only indicate the hardenability to be expected with these chemistries, the adoption of the same for castings can only lead to approximate values which may or may not be applicable.

Hardenability in Foundry

When it is furthermore considered that the major reason for introducing hardenability in the foundry is to insure that the casting is fully quenched out to a predetermined depth, it appears simpler to base the hardenability requirements on a minimum as quenched Rockwell hardness representing a certain percentage martensite that must be obtained in a given section thickness than to attempt an adaptation of a series of bands. In particular is this so as the chemistry to be used is to be left at the option of the foundry.

In order to establish the minimum as quenched hardness, however, it is necessary to decide upon the percentage martensite which shall represent a quenched out structure. Although it is generally conceded that full quench hardness is required in order to obtain maximum strength and ductility, this in practice frequently leads to serious difficulties because of the magnitude of internal stresses produced.

Grossman in his pioneering work established a 50 per cent martensite limit at the center of a given cross section as the minimum requirement for full hardening of a steel. Although the 50 per cent martensite structure is sufficient to insure so-called "full hardening" for carbon and low alloy steel, considerable difference of opinion exists as to the adequacy of this percentage for higher hardenability steels.

With the knowledge at hand supplied by such investigators as Boegehold and others, coupled with the experience obtained in a number of foundries working to hardenability requirements during the war, it was felt that a minimum of 80 per cent martensite constitutes a workable basis.

To guard against over alloying which may result in excessive hardenability leading to cracking, maximum limits should rightfully be included along with minimum limits. The question arises, however, as to what would constitute the proper basis for maximum limits.

Susceptiblity to Cracking

This is particularly so as little data is available as yet relative to the relationship between hardenability and susceptibility to cracking. Lacking this necessary data is was believed advisable to exclude maximum limits in the foundry until such time that more information on crack sensitivity is at hand.

Specific cases may, however, arise where the casting forms a part of a composite structure requiring welding, thereby necessitating a maximum limitation which usually is then best procured by limiting the carbon content.

In view of the above considerations, it was decided to base the proposed specification on minimum hardenability values only, leaving the analyses to the option of the foundry, except that the carbon limits must be declared to the consumer in order that the Rockwell hardness values for the as quenched structure can be established.

Depending upon the stresses to which the castings will be subjected in service, the hardenability requirements will differ. In the case of tensile and impact stresses for instance, as has already been stated, it is generally conceded that maximum hadenability is essential in order to obtain maximum properties; whereas in the case of torsion or bending stresses a relatively shallow hardening steel may suffice. This latter statement is, however, still debatable and two schools of thought exist on this subject.

In order to meet these varying requirements, it was desirable to write the specification sufficiently flexible to allow for varying degrees of hardenability for similar cross sections. With this in mind Table 3 has been proposed.

Table 3 shows the minimum as quenched Rockwell hardness for varying carbon contents necessary to obtain an 80 per cent martensitic structure and the equivalent distance on the Jominy bar for varying cross sections hardened to the center, as well as one-fourth and one-eighth of the section thickness.

Critical Section Thickness

In selecting the hardenability factor for any given casting or group of castings, both the so-called "critical section thickness" of the casting, by which is here meant the thickness of the most highly stressed portion, and also the type of stresses to which the casting will be subjected in service, as already mentioned, must be carefully considered. The application of the table is probably best shown through the following example. See Table 3.

In the case of a casting with a critical section thickness of one inch (as specified by the consumer) and required to meet a minimum of 120,000 psi., in other words type S.A.E. 0120, and subjected in service to stresses requiring full hardenability within the limits of the specification, the following procedure is applied.

Consulting Table 3 we find at the top that one inch thickness corresponds to factor H-2, and by following the vertical line to the bottom of the table to C that in order to harden to the center a harden-

ability equal to a distance of 5/16 in. on the Jominy end-quench bar is required.

Although the chemistry does not form a part of the specification, the foundry must declare to the consumer the minimum carbon limit for the steel to be used. In the case of 0.30 per cent carbon steel, as indicated on the left side of the table, it will be found by following the horizontal line that the corresponding as quenched hardness must be Rockwell C 42 minimum for 80 per cent martensite. This specification will, therefore, read S.A.E. 0120 H2C and indicates that a steel of sufficient hardenability to produce a minimum of Rockwell C 42 at a distance of 5/16 in. on the Jominy end-quench bar must be selected.

Similarly, castings subjected to stresses requiring only shallow hardening or which are to be used for parts requiring high surface hardness, can be specified as to required hardenability by following the procedure outlined.

Application of Hardenability in the Foundry. In order for the foundry metallurgist to decide what type of steel to apply to a given specification, it becomes necessary to establish the relationship between chemistry and hardenability. This is best done by the method worked out by Field which is shown in Fig. 1 to 5 and Tables 4 and 5. The writer shall cite an example to indicate its application.

Having a steel of the analyses as shown in Table 4, the approximate grain size must first be decided upon. In the writer's experience a grain size seven has generally been found satisfactory when aluminum is used as a deoxidizer. From Fig. 1 we find that a not alloyed steel of 0.32 per cent carbon with a grain size seven has an ideal critical diameter of 0.19. From Fig. 2 the multiplying factors for the balance of the elements present will be obtained. The ideal critical diameter for the steel in question is then found by multiplying the factors of all these elements as shown in Table 4.

Having established the ideal critical diameter, the initial hardness of Rockwell C-51 for the given carbon content is obtained from Fig. 3. By now applying Fig. 4 the ratio between initial hardness and distance hardness values are obtained for the given critical diameter, and the Rockwell hardness at various distances on the Jominy curve determined as shown in Table 5. The construction of the Jominy curve follows as indicated in Fig. 5. Close correlation between calculated and actual hardenability is representative of a large number of such tests carried out in the Ford Motor Company steel foundry (Fig 6).

Table 3

CRITICAL SECTION THICKNESS OF CASTING

Class	H1	H2	H3	H4	H5	Approx. Per Cent Martensite
Thickness, in.	½	1	1½	2	2½	
.45						80
.40						50
.35						48
.30						45
.25						42
.20						39
						36
A—	1½	2½	3½	4½	5½	
B—	2	4	6	8	10	
C—	3	5	8	11	14	
D—	4	8	11	14	18	

CARBON CONTENT

MINIMUM AS QUENCHED ROCKWELL HARDNESS "C"

Equivalent Jominy Distance in Sixteenths of an Inch

A—For Desired Quench Hardness to a Depth of ⅛ of Critical Section Thickness—Agitated Water Quench.

B—For Desired Quench Hardness to a Depth of ¼ of Critical Section Thickness—Agitated Water Quench.

C—For Desired Quench Hardness at Center of Critical Section Thickness—Agitated Water Quench.

D—For Desired Quench Hardness at Center of Critical Section Thickness—Agitated Oil Quench.

NOTE: The values in this table were obtained from plate sections. For rounds see S.A.E. handbook, page 318.

Residual Alloys

High residual alloys may upset the calculations somewhat but as this usually is in the direction of increased hardenability no appreciable ill effect will result.

This method of calculation not only is of value in determining the original chemistry required, but it has also the evident advantage that the hardenability can be calculated before the heat is tapped, providing rapid preliminary analyses are obtained and adjustments made in the furnace if necessary.

For the actual determination of hardenability in the foundry a standard Jominy end-quench bar is cast in a permanent mold, eliminating all machining except for cutting off and the grinding of parallel flats for the Rockwell readings. A typical

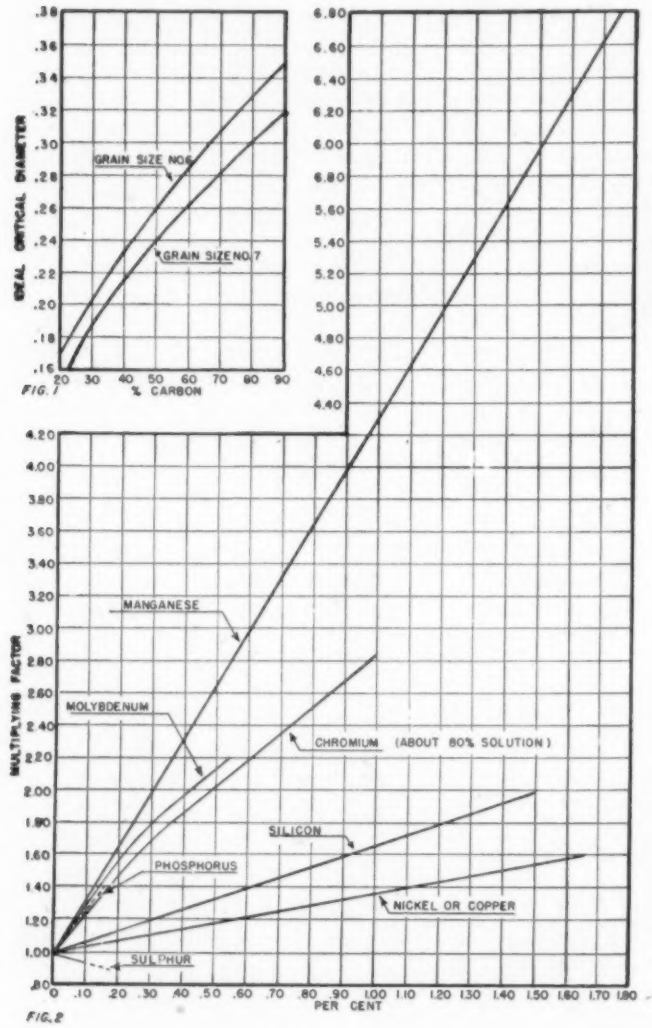


FIG. 1

FIG. 2

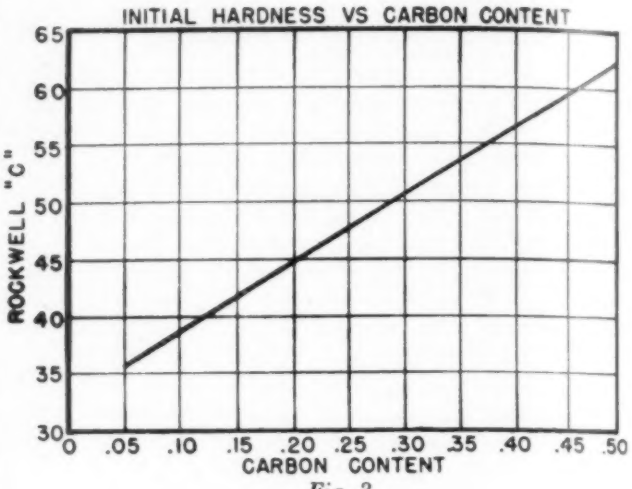


Fig. 3

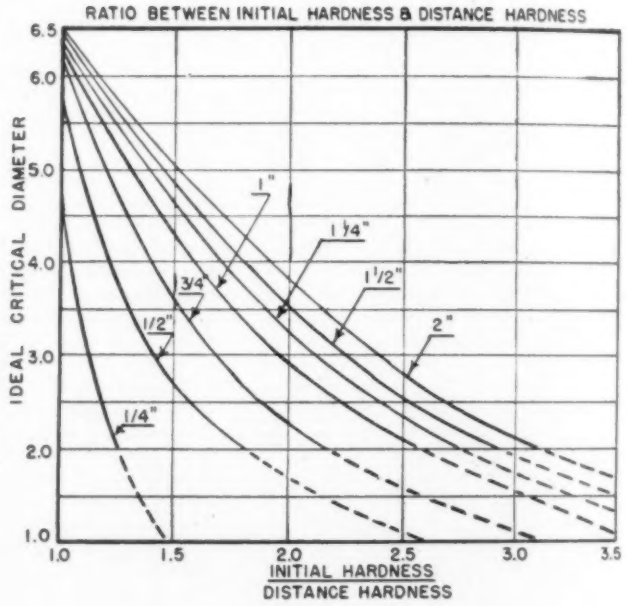


Fig. 4

Table 4—TO CALCULATE JOMINY HARDENABILITY OF A STEEL OF ANALYSIS SHOWN PROCEED AS FOLLOWS:

Item	*Amount*	*Factor*
Grain Size 7	.....	.....
Carbon (Fig. 1)	.32	.193
Manganese (Fig. 2)	.95	4.16
Silicon	.27	1.18
Phosphorus	.030	1.02
Sulphur	.015	.99
Nickel	.59	1.21
Chromium	.66	2.28
Molybdenum	.40	1.94

Ideal critical diameter = D_I = .193 x 4.16 x 1.18 x 1.02 x .99 x 1.21 x 2.28 x 1.94 = 5.15

Table 5—HARDENABILITY CALCULATION

For carbon content .32, the initial hardness is 51 R "C" (Fig. 3); from Fig. 4, hardness from quenching end

		R "C"
At	¼ in. = 51/1.0	= 51
	½ in. = 51/1.06	= 48
	¾ in. = 51/4.16	= 43.5
	1 in. = 51/1.27	= 40
	1¼ in. = 51/1.35	= 37.8
	1½ in. = 51/1.40	= 36.4
	2 in. = 51/1.46	= 35

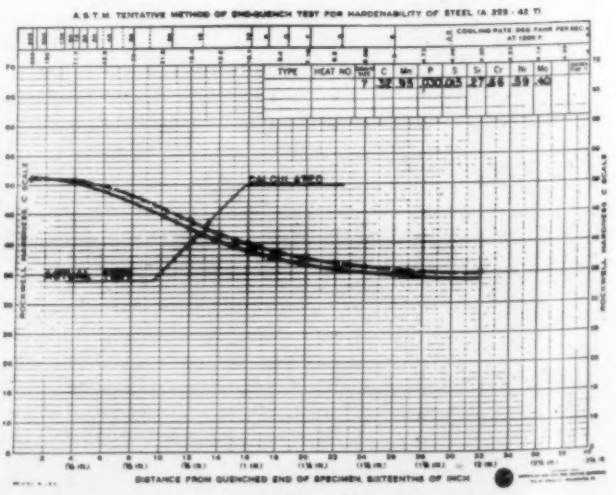


Fig. 5

CURVES AND TABLES USED TO ILLUSTRATE CALCULATIONS OF JOMINY HARDENABILITY IN TEXT

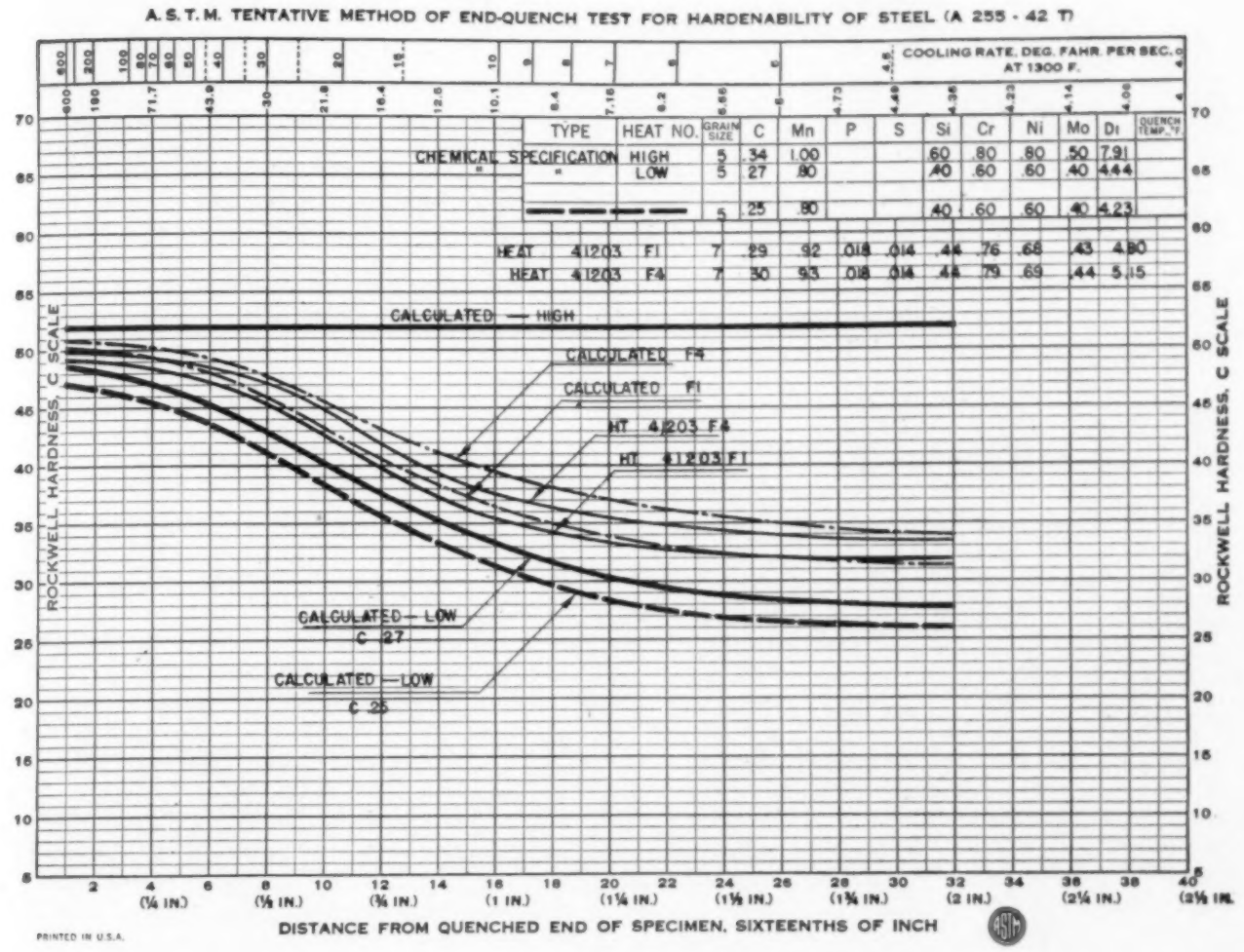
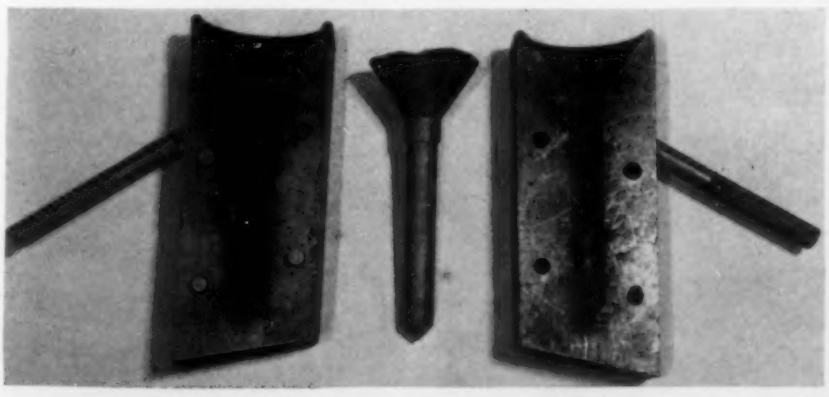
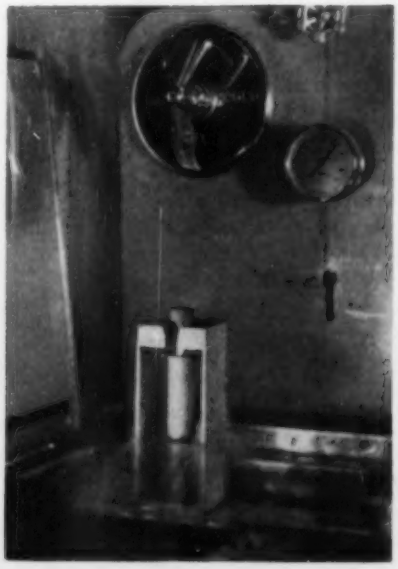


Fig. 6—Curves show close correlation between actual and calculated hardenability in a large number of tests performed at the author's plant.

Fig. 7a (Below)—Permanent mold for casting a standard Jominy end-quench bar. Fig. 7b (Right)—Standard Jominy bar being quenched in the quenching fixture under standard conditions.





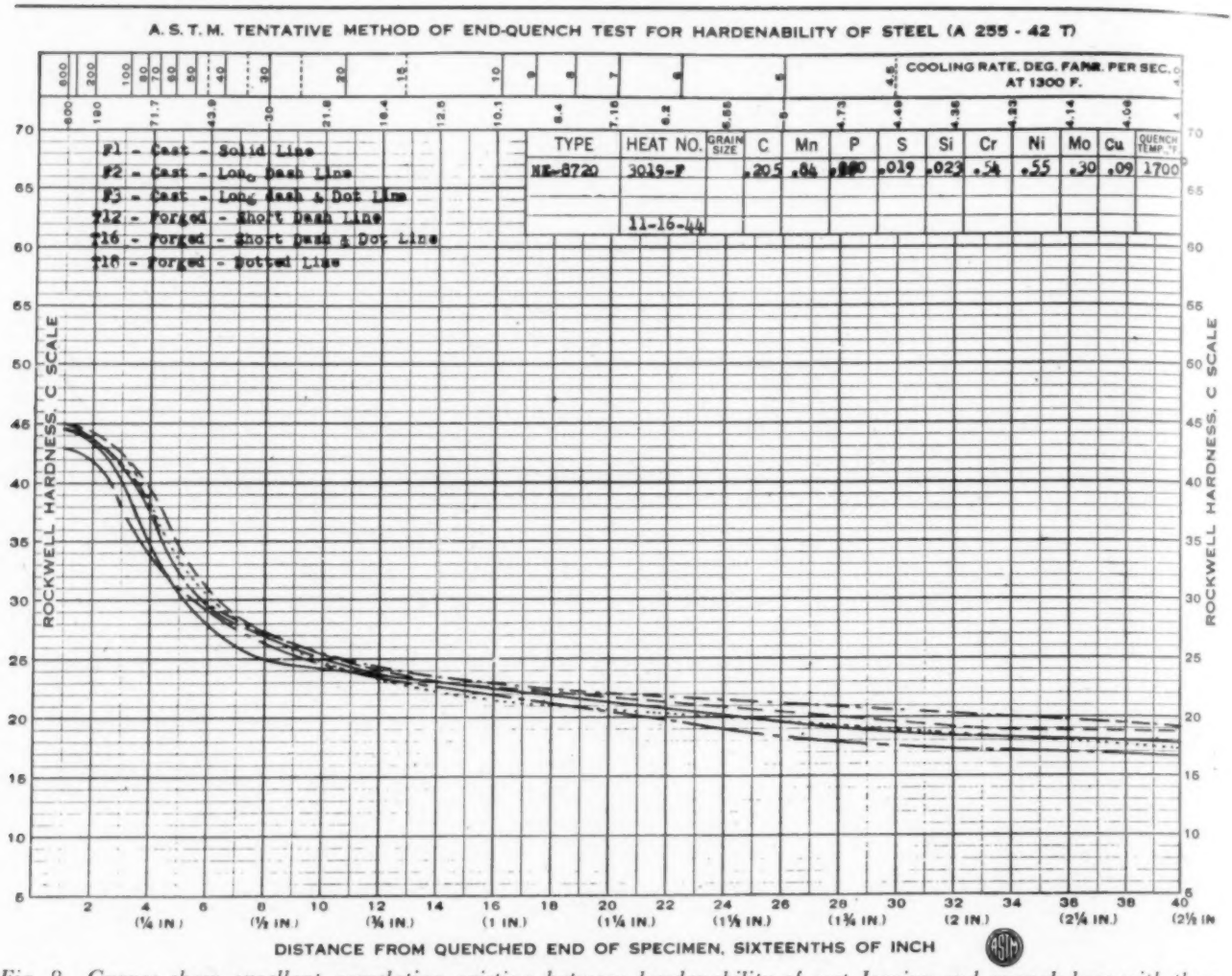
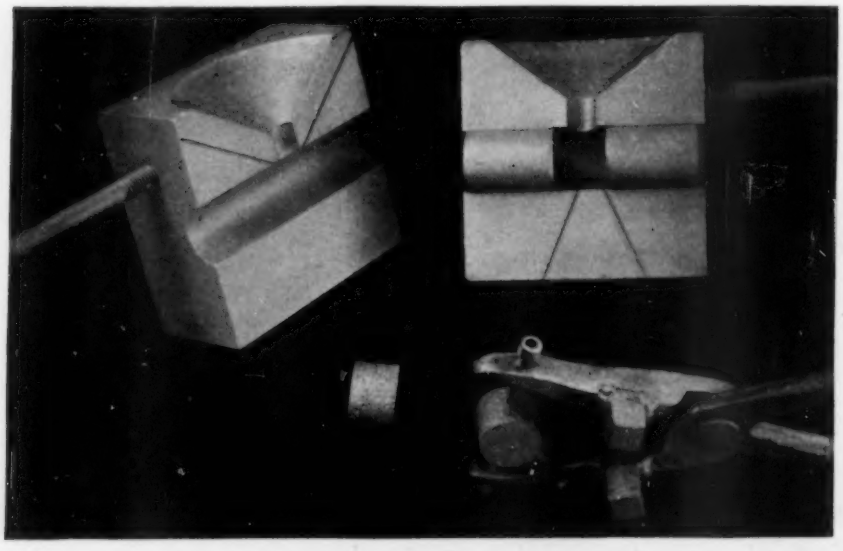
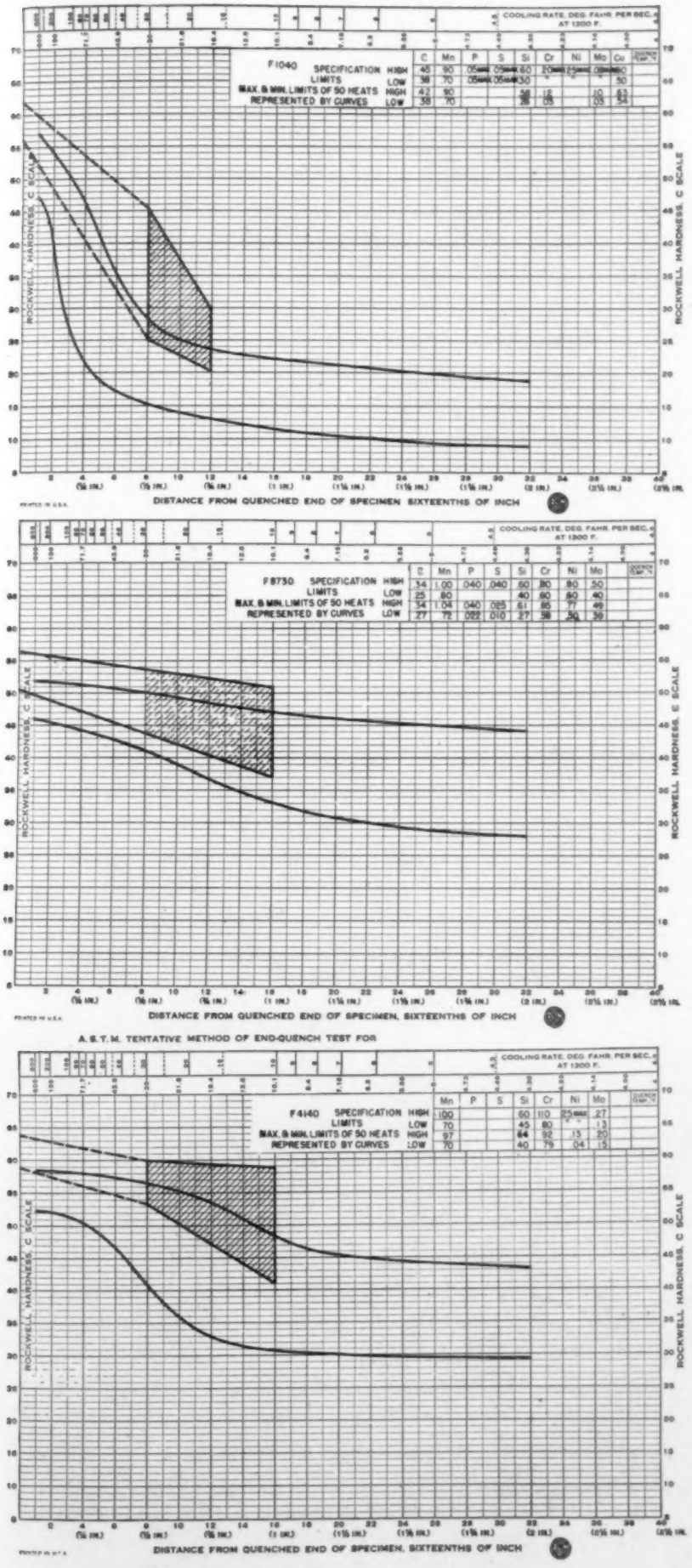


Fig. 8—Curves show excellent correlation existing between hardenability of cast Jominy end-quench bars with those of rolled or forged bars.

Fig. 9a (Below)—Showing removal of "slug" hardenability specimen from furnace with a pair of tongs fitted with three pins for grabbing same and spacers to the distance from the quench orifice. Fig. 9b (Right)—Showing tongs in use.







mold used for this purpose is shown in Fig. 7.

Extensive tests have been carried out to compare the results obtained from cast Jominy end-quench bars with those of rolled or forged bars. The excellent correlation existing between these various methods is indicated in Fig. 8.

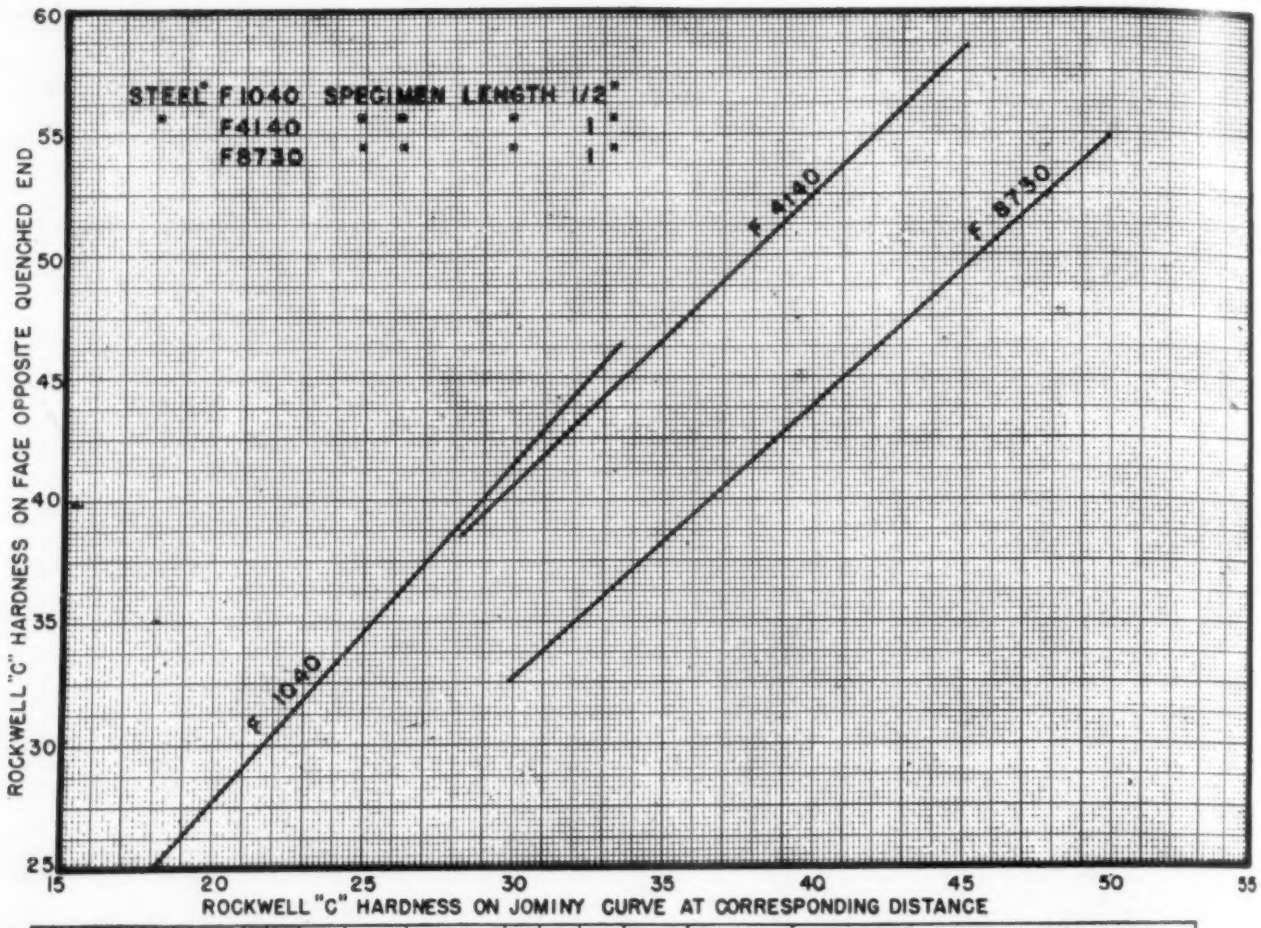
Preliminary Investigation of Hardenability Control in the Furnace. The desirability for a simplification of the hardenability test in the foundry is evident when we consider the time required to carry out a Jominy end-quench hardenability test even with a cast specimen. The fact that it is impossible to obtain results prior to the pouring is a much greater handicap in the foundry than is the case in the mill for reasons already discussed.

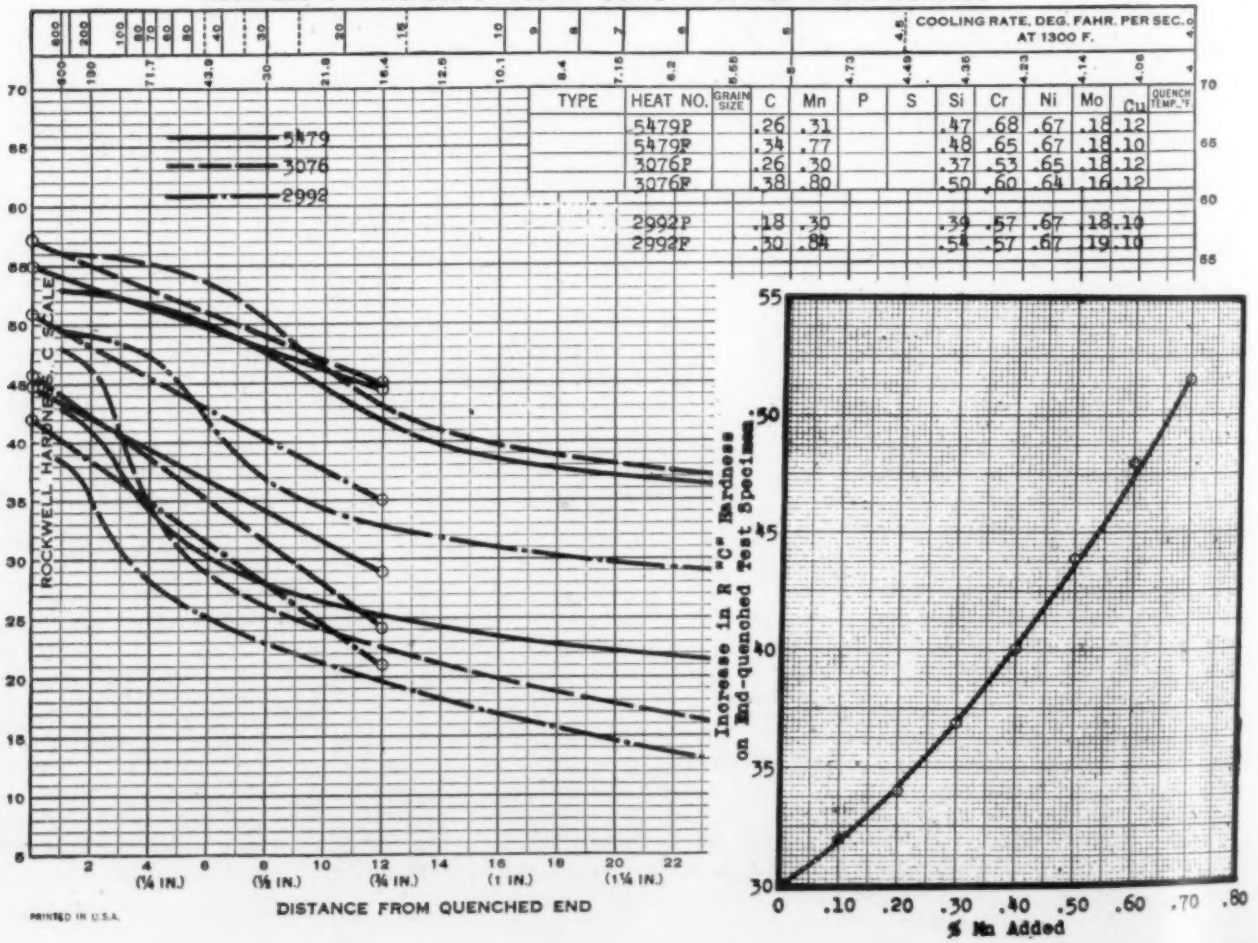
The ultimate aim must, therefore, be some method which permits the metallurgist to obtain the pertinent information relative to the hardenability characteristics of any given heat of steel in time to make the necessary corrections.

The avenue of approach selected for a preliminary investigation toward this end at the Ford Motor Company consisted of end-quenching 1-in. round specimens of predetermined length, depending upon the steel to be tested. The Rockwell hardness values obtained on both ends were used to evaluate the results, where the hardness at the quenched end was a measure of the carbon content and the hardness at the opposite end a measure of the susceptibility of the steel in question to harden.

Accordingly, a die or permanent mold was made which permitted adjustments in the sample length of one-half inch, three-quarters inch or one inch. These lengths were decided upon after preliminary tests indicated the same to be most satisfactory for the steels in question, namely modified S.A.E. 1040, 8730 and 4140 steels.

Fig. 10a (Above), 10b (Center), and 10c (Bottom)—Curves show max. and min. Jominy hardenability obtained and equivalent max. and min. Rockwell readings on the corresponding "slug" specimens representing approximately 50 heats from each of the steels tested.





TYPE	HEAT NO.	GRAIN SIZE	C	Mn	P	S	Si	Cr	Ni	Mo	Cu	QUENCH TEMP. °F
	5479P		.26	.31			.47	.68	.67	.18	.12	
	5479P		.34	.77			.48	.65	.67	.18	.10	
	3076P		.26	.30			.37	.53	.65	.18	.12	
	3076P		.38	.80			.50	.60	.64	.16	.12	
	2992P		.18	.30			.39	.57	.67	.18	.10	
	2992P		.30	.84			.54	.57	.67	.19	.10	

Captions for Figs. on previous page. Fig. 11 (Top)—Correlation of readings obtained on "slug" specimens with the corresponding Jominy readings at equal distances from the quenched end.

Fig. 12 (Bottom)—Curves representing some typical basic electric furnace heats showing hardenability before corrections were made to the liquid bath and after pouring. Included is a table indicating per cent in Mn addition required to obtain a given increase in as-quenched hardness.

The procedure consisted of removing the cast specimen as quickly as possible from the die, and after quenching the same, cutting off the gate. The reason for the rapid cooling from the die is to prevent the precipitation of carbide-forming elements which would increase the required soaking time on reheating.

A small electric furnace, or preferably an electric resistance unit, was used to obtain the proper austenitizing temperature. The quenching equipment used was the same as for standard Jominy end-quench hardenability bars, and the procedure identical. The sample was removed from the furnace or heater with a pair of tongs fitted with three pins for grabbing the same, and spacers to control the distance from the quench orifice as shown in Fig. 9.

The Rockwell hardness on each face was obtained by averaging four readings taken at a point midway between center and outside. Although the method in its preliminary form was crude, the results obtained correlate exceptionally well with Jominy end-quench hardenability bars cast at the same time representing approximately 50 heats from each of the steels tested and showing the maximum and minimum Jominy curves obtained and the equivalent maximum and minimum Rockwell readings on the corresponding so-called "slug" specimens (Fig. 10a, b, c).

In order to correlate the readings obtained on the specimens with the corresponding Jominy readings at equal distances from the quenched end, a large number of points were plotted and the following curves constructed representing the averages (Fig. 11). The results of these tests indicated that the method described was applicable as a rapid hardenability test.

The second part of the investigation consisted of an attempt to apply the method to the liquid bath and obtain hardenability readings prior to pouring making possible any necessary corrections.

It appeared that this procedure would be perfectly feasible for the acid electric furnace practice, as well as acid open hearth practice, where the alloying elements with the exception of the final manganese addition generally are added early enough to permit taking a reading. The final manganese addition in this case would be used as the correcting medium.

In the basic electric furnace the same procedure can be followed except that here addition of both manganese and silicon are usually delayed until just before the heat is ready to tap. As the effect of silicon, however, on the hardenability and the amounts added are rather minor, manganese again becomes the element to be used.

The results obtained to date show promise, but it is evident that much additional data must be collected before the suitability of this test can be finally evaluated. From the preliminary information at hand the curves (Fig. 12) representing some typical heats have been constructed showing the hardenability before corrections were made to the liquid bath and after pouring. Included is also a table indicating the per cent manganese additions required to obtain a given increase in the as quenched hardness.

The foregoing discussion has been offered in an attempt to analyze the need for hardenability requirements in the foundry and some of the pertinent factors relating to the hardenability of steel castings, in particular those that are directly affected by variations in the foundry practice.

A brief discussion of the proposed S.A.E. hardenability requirements has also been included, as well as the method for obtaining hardenability readings in the foundry. In addition, some results based on the preliminary investigation of a method applicable for obtaining hardenability readings of the molten metal have been presented.

It is our firm belief that hardenability requirements will find wider and wider application in the foundry, and that we, therefore, must concentrate our efforts in developing rapid methods for this determination whether they be along the lines described herein or by other means.

DISCUSSION

Chairman: C. H. LORIG, Battelle Memorial Institute, Columbus, Ohio.

Co-Chairman: J. A. RASSENFOSS, American Steel Foundries, East Chicago, Ind.

H. H. BLOSJO[1]: To make adjustments of hardenability, do you use a low or high carbon ferro-manganese?

MR. VENNERHOLM: The type of ferro-manganese used, whether low or high carbon, depends upon the requirements of the particular heat in question. Along with the hardness readings obtained on the test specimen for the purpose of evaluating the hardenability of the heat, we also take readings on the quenched end and by applying a carbon vs. hardness chart get an excellent indication of the carbon content of the heat at this point. This enables us to decide immediately whether to use high or low carbon ferro-manganese.

T. N. ARMSTRONG[2]: Mr. Vennerholm should be complimented upon his excellent paper and the manner in which the subject was presented. The comment that I have to make is in no way a criticism of Mr. Vennerholm's paper, but rather to point out a possible fallacy in the interpretation of the hardenability concept. The fact is too frequently lost sight of that steels of equal hardenability have equivalent properties *only* if the structures also are equivalent.

Grossman based his evaluation of critical diameters on 50 per cent martensite at the center. This means that at least 50 per cent of the structure at the center is comprised of transformation products other than martensite, and it has been rather definitely determined that the properties of intermediate transformation products differ with composition of the steel.*

As most heat treated steels do not contain 100 per cent martensite when quenched in the sizes used in engineering applications, differences in properties may result because of differences in the properties of intermediate transformation products.

There is currently being carried on some work which indicates that steels transformed at a given temperature do not necessarily have the same structures. High temperature transformation products of some steels are accompanied by brittleness and the brittleness is a function of the structure. The structures are dif-

*W. T. Griffiths, L. B. Pfeil, and N. P. Allen, *The Intermediate Transformation in Alloy Steels*, Second Report, Alloy Steel Research Committee, Iron and Steel Institute, (1939).

[1]Minneapolis Electric Steel Castings Co., Minneapolis.

[2]International Nickel Co., New York.

ferent for different steels transformed at the same temperature, consequently, the statement that properties are equivalent for equivalent hardenability is not always valid.

The work that was presented originally by E. J. Janitzky and his associates indicated that for steels of about equivalent carbon content, equivalent properties would be secured after heat treatment provided the hardnesses were of the same order. The sections investigated were of such dimensions that they probably developed nearly 100 per cent martensite on quenching.

This leads to the question as to whether you can entirely disregard composition. The steel industry originally attempted hardenability only, but early recognized to develop "H" steels on the basis of the error in trying to disregard chemical composition in promoting these steels and have now established rather wide limits of composition for each type, which permits the consumer some freedom in selecting a type analysis for a given hardenability.

MR. VENNERHOLM: The question as to whether reproducible properties can be obtained regardless of composition is debatable, and it may be necessary to introduce certain carbon limitations in order to arrive at the ultimate goal. As for the balance of the composition, I believe that we can leave the same to the option of the foundry.

It may at this point be in order to bring out the fact that the curves which have been presented are based on actual operation in the foundry covering a period of nearly a year during which period a large number of armor heats were accepted by the government on the listic performance requirements waived.

It may be that hardenability is not the ultimate solution in the foundry, but as far as our present knowledge goes, it appears that hardenability tests certainly have their place and serve a very definite purpose. We are all studying the problem trying to find better methods of evaluating the properties of cast steels, and for lack of anything better we are, at least for the time being, continuing our work on the present hardenability method as a means to this end, and the results so far are, indeed, encouraging.

MR. ARMSTRONG: I am not criticizing the test at all. I am merely pointing out that you cannot use the hardenability test alone as a criterion for measuring properties.

C. E. SIMS[2]: It seems that 50 per cent martensite is as good as getting 85 or 90 per cent martensite as far as determining the hardenability, but it is not a measure of the probable properties of that steel. It may be of interest to note that the Committee on Alloy Steel of the ASM has taken a stand that equivalent properties among alloy steels basis of hardenability only with all balare obtained only when quenched to at least 95 per cent martensite.

MR. VENNERHOLM: In answer to Mr. Sims' remarks I would like to state that our committee fully recognized this fact, and that, as you have noticed in Table 3 showing the hardenability table, the same is based on 80 per cent martensite.

Because of the present lack of experience and the additional problems to be anticipated in the foundry, we did not deem it advisable to go above this percentage for the present. Ultimately we hope to be able to adopt a percentage as near 100 per cent as possible.

[2]Battelle Memorial Institute, Columbus, Ohio.

THEORIES OF GRAY CAST IRON INOCULATION

• *Mechanics and theories of the gray iron inoculation process are presented in this paper, serving as an introduction to the several papers comprising the "Symposium on Inoculation of Gray Cast Iron" which are in course of preparation for later presentation.*

By H. W. Lownie, Jr.
Formerly Metallurgical Engineer
Westinghouse Electric Corp.
East Pittsburgh, Pa.

GRAY iron foundrymen during recent years have become quite interested in the process of inoculation. Many foundrymen responsible for the production of high quality cast iron have adopted an inoculation practice because they have found important product improvements to result.

In introducing a symposium on inoculation, it appears advisable to review in a general way some of the mechanics and theories of the process.

Rather than present throughout this discussion numerous references to various authorities, a general reference is made here to three publications[1-2-3] which discuss in more detail the present state of the theory and which present adequate bibliographies for the reader who wishes to study the subject further.

Inoculation may be defined as that process in which an addition is made to molten cast iron for the purpose of altering or modifying the microstructure of the iron and thereby improving the mechanical and physical properties to a degree not explainable on the basis of the change in composition.

To illustrate, iron analyzing, say, 3.20 per cent total carbon and 2.20 per cent silicon in the as-melted condition, may be inoculated with an addition of 0.20 per cent silicon. Some of the properties of the resulting 3.20 per cent total carbon, 2.40 per cent silicon iron will be considerably different from the properties of the original untreated iron, or from the properties of an iron melted directly (without inoculation) to the 3.20 per cent carbon, 2.40 per cent silicon composition.

Table 1
CLASSIFICATION OF TYPICAL LADLE INOCULANTS

Group 1		*Group 2*
Ca-Metal	Si-C	Cr-Si-Mn-Ti-Ca
Ca-Si	Si-Mn	Cr-Si-Mn-Zr
Ca-Si-Ti	Si-Mn-Zr	Mo-Si
Fe-Si	Si-Ti	Ni-Si
Graphite	Si-Zr	

The foregoing definition of inoculation is intended to exclude alloy additions made to the molten iron for the sole purpose of affecting the chemical composition of the melt (e. g., ferrochromium, ferromolybdenum, nickel, etc.), and to exclude materials added to the charge when those materials react with the metal before melting is completed.

These materials which are added to the iron to produce inoculation are called "inoculants," and the irons so treated are referred to as "inoculated" irons.

Inoculation Classification

One method of classification of inoculants is by placing them in two groups as follows:

Group 1—Inoculants which have the sole duty of producing inoculating effects (such as a changing of the graphite distribution and reduction of chill depth) to a degree considerably greater than can be explained on the basis of a change in composition. These inoculants usually contain one or more of the following elements: carbon, silicon, calcium, titanium, zirconium, and aluminum.

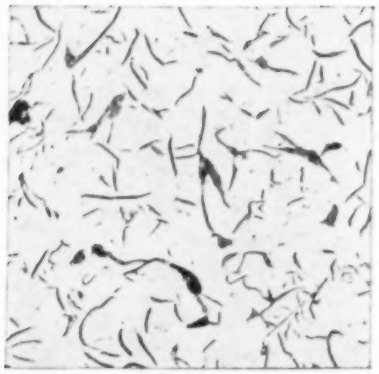
Group 2—Inoculants which, in addition to producing inoculating effects to a degree considerably greater than can be explained on the basis of a change in composition, also exert a supplementary effect which is explainable on the basis of a change in composition. In addition to one or more of the elements listed in Group 1, these inoculants also contain one or more additional elements which are not essential to the inoculating reaction but which do exert supplemental effects which are desirable in some applications.

The most common supplemental alloys used in this type of inoculant are chromium for the stabilization of carbide (pearlite), or nickel to promote carbide decomposition. Chromium and nickel are not considered as inoculating elements in themselves, as the effects of these elements are explainable on the basis of the change in chemical composition.

Typical ladle inoculants may be divided into these two groups as shown in Table 1.

The approximate analysis of each of these inoculants is given in one of the references[4] at the end of this paper. However, it should be added for the sake of completeness that, with the exception of metallic calcium and graphite, all of the inoculants shown in Table 1 probably also contain aluminum in amounts which are small but nevertheless may be quite significant.

Microconstituents Affected by Inoculation—Since the primary effect of an inoculant is to produce certain changes or alterations in the microstructure of the iron, it is necessary to define the microconstituents that



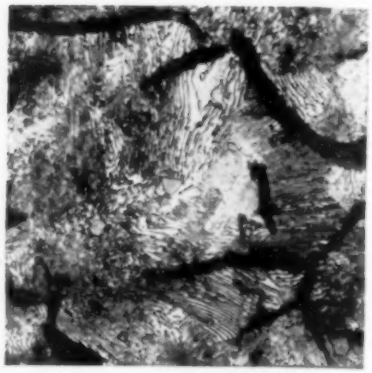


Fig. 1—Type A random distribution of graphite flakes in a matrix of pearlite. Graphite rating—5A. (Left) Unetched, X 100. (Right) Etched, X 750.

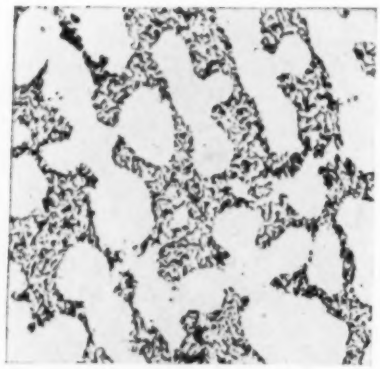
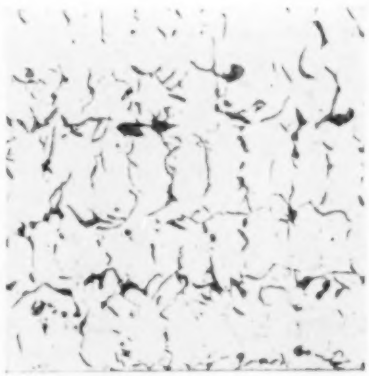


Fig. 2—Type D interdendritic (random orientation) graphite flakes. Etched structure shows tendency of massive ferrite (white) to be associated with this type of graphite distribution. Graphite rating—8D. (Above) Unetched, X 100. (Below) Etched, X 750.



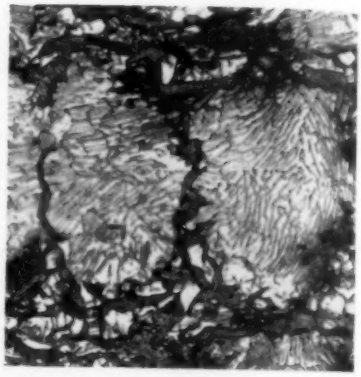
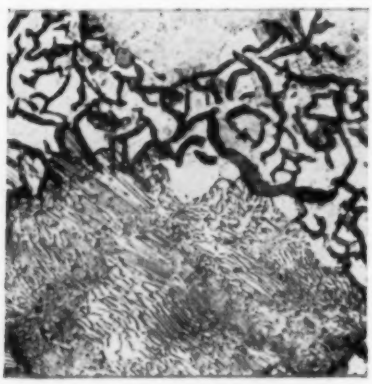


Fig. 3—Type E interdendritic (preferred orientation) graphite flakes in a matrix of pearlite. Graphite rating—5E. (Left) Unetched, X 100. (Right) Etched, X 750.





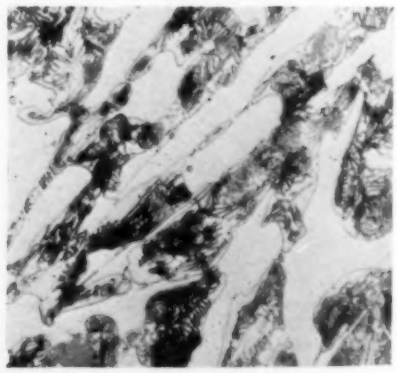


Fig. 4—White cast iron showing complete suppression of graphite flakes. White areas in etched structure are hard massive iron carbide (cementite). Graphite rating—none. (Left) Unetched, X 100. (Right) Etched, X 750.

are changed.

The most pronounced change resulting from inoculation is the change occurring in the distribution of the graphite flakes throughout the matrix or background metal. These graphite flakes may be large and coarse, fine and lacy, or even may be completely absent, depending upon the conditions of metal composition, cooling rate, and inoculation.

The A.F.A. and A.S.T.M. jointly have adopted a standard chart[5] for the classification of graphite flake size by number and graphite flake distribution patterns by type letter. This method of classification is used in this work.

First consideration will be given to illustrating the various microstructures involved. Subsequent paragraphs will illustrate when these particular structures are formed, and other articles in the inoculation symposium will discuss the effects of these changes in microstructure upon physical properties.

Graphite Flake Distribution

Figure 1 illustrates large graphite flakes dispersed in a random manner through a matrix of pearlite. Such a random distribution of flakes is designated as Type A. The flake size is identified by a size number. Inoculation is performed for the purpose of promoting this Type A graphite.

Type D graphite is shown in Fig. 2. This fine and lacy graphite pattern is described as "interdendritic

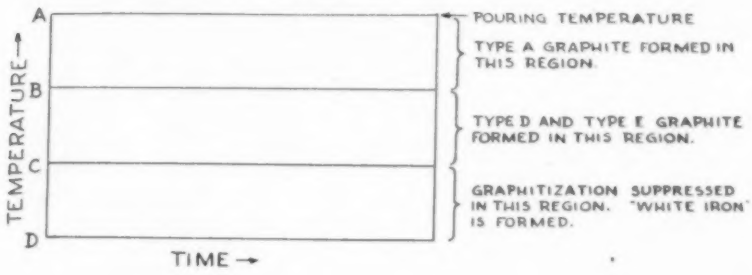


Fig. 5—Schematic representation of temperature regions of graphite formation.

segregation, random orientation." The continuous network of the fine graphite flakes tends to lower physical properties. In addition, free ferrite (as shown in the etched microstructure of Fig. 2) tends to be associated with this type of distribution and to lower the physical properties.

Inoculants are added to cast iron to attempt to eliminate this undesirable Type D graphite distribution and its associated ferrite masses.

Type E graphite (interdendritic segregation, preferred orientation, Fig. 3) also is considered as an undesirable form of graphite distribution. The statements regarding the undesirability of Type D graphite also refer to the Type E distribution.

"White Iron" Occurrence

Although Type D and Type E graphite generally are to be avoided, there are cases where the formation of any type of graphite is preferable to the occurrence of massive iron carbide ("white iron").

In such cases, although the inoculant may not be potent enough to produce the most desirable graphite distribution (Type A) by inoculation, it may, by the alloying effects of its constituent elements (e. g., silicon and nickel), result in the formation of one of the less desirable forms (Type D or Type E), which in turn may be preferable to the formation of massive iron carbide.

Figure 4 shows the case of an iron that has solidified completely "white," that is, there is no graphite present. The carbon in this iron which under some conditions would solidify as flake graphite has been retained in chemical combination with the iron and has formed the hard, white iron carbides that give "white iron" its name.

The A.S.T.M. - A.F.A. classification system also contains distribution ratings of Type B and Type C but, for the purposes of this discussion, these types may be considered as relatively unimportant.

Cooling Rate

Effect of Cooling Rate Upon Graphite Structure — The graphite structure of cast iron is determined primarily by its composition, cooling rate, and the degree of inoculation. Although the effect of chemical composition is of undeniable importance, the effect of cooling rate upon a given iron will be used to set up an arbitrary system for discussing the formation of flake graphite in cast iron.

This system then will be expanded to consider the effects of composition and inoculation. Although the system does make certain assumptions without confirmed metallurgical basis, it is believed that an attempt to make a graphical presentation will be helpful in discussing inoculation.

As an elemental hypothesis, consider that the temperature range from the pouring temperature down to "red heat" is divided into three regions, as shown in Fig. 5. In this schematic sketch the location of the dividing lines B and C is a subject of considerable experiment and debate over whether graphite can or cannot form directly from the melt or from austenite.

After an extensive review of the literature on the formation of graphite flakes, Epstein[1] stated: "It may be concluded that no clear-cut proof has been presented that an iron-graphite eutectic may form directly from the melt, neither is the contrary evidence which has been brought forth so positive as to preclude the possibility of the formation of such an eutectic."

With reference to our simplified diagram, this simply means that the solidification temperature (or temperature range) is not definitely located with respect to the lines (or regions) B and C. Although of considerable theoretical interest, the accurate determination of the location of these regions is not necessary to understand in a practical manner how inoculants work.

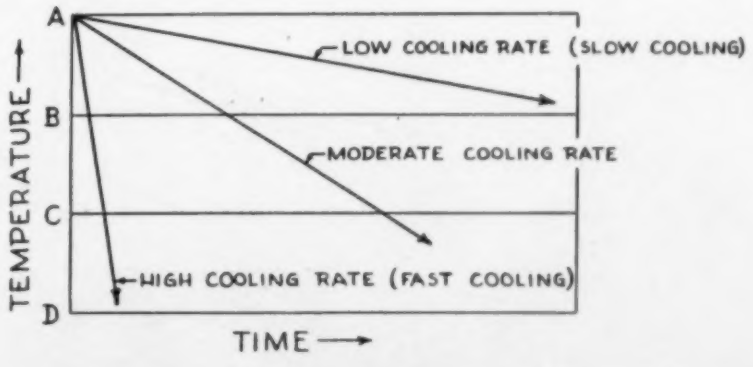
Graphitization Behavior

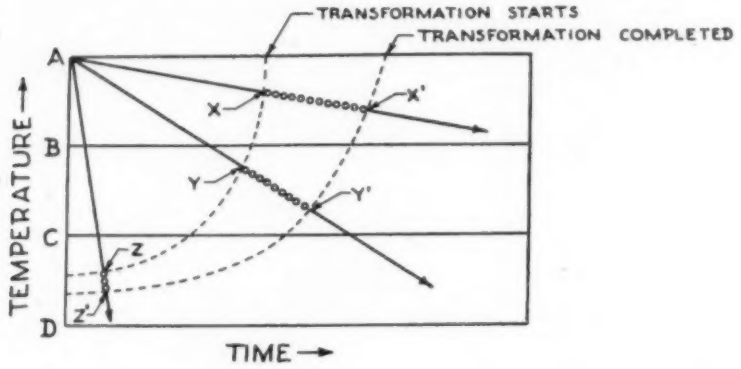
It is necessary to emphasize here that these schematic sketches are only approximate "behavior diagrams," and may not be interpreted too closely. The value of the diagrams lies in the fact that they do provide a simple manner of representing graphitization behavior.

The facts that the boundary lines A, B, C, and D are drawn at constant temperature and that the three regions are shown to be of equal temperature ranges should not be considered as significant.

Various cooling rates can be represented by lines of differing slope originating at the pouring temperature, as shown in Fig. 6. In ordinary

Fig. 6—Representation of cooling rate on schematic time-temperature diagram.





X, Y, Z—Points at which transformation starts with each of the indicated cooling rates.

X', Y', Z'—Points at which transformation stops with each of the indicated cooling rates.

Fig. 7—Representation of time required for graphitization or formation of white iron.

sand casting practice, low rates of cooling are obtained with thick casting sections, and high rates with thin castings. The use of chills also promotes high rates of cooling.

Physical transformations (such as the formation of graphite flakes) do not occur instantaneously. After the conditions for transformation are established, a time "lag" usually occurs before the transformation will commence, and then some finite period of time is required for reaction to be completed.

Such a transformation zone may be represented schematically on the diagram as shown in Fig. 7. An iron will not begin to form graphite flakes of the type indicated by a particular region until this minimum time is exceeded, and the graphitization that is going to occur will be complete when the cooling curve leaves this zone.

Time-Temperature Relation

Another concept is desirable. As the cooling rate becomes lower, the transformation temperature becomes higher and the time required for completion of the reaction becomes longer. This situation is represented in Fig. 7 by the increasing horizontal distance (time) as the cooling rate becomes lower.

The higher temperatures and longer transformation times at low cooling rates (slow cooling) cause the graphite flakes forming at low rates to become quite large. Rapid cooling and rapid formation of flakes result in flakes of a small size.

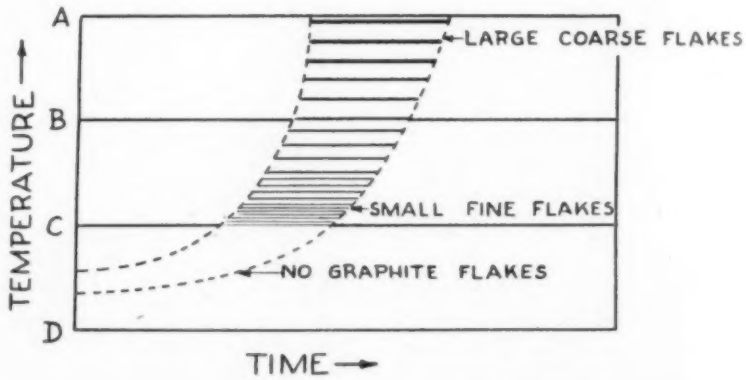


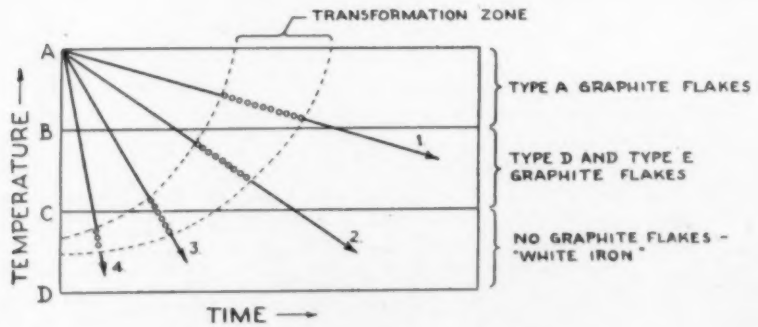
Fig. 8—Representation of effect of transformation time on graphite flake size.

Thus, Type A flakes (formed at low rates) usually are larger than Type D or Type E flakes (formed at higher rates). This situation is indicated in Fig. 8.

Bringing together the factors illustrated in Figs. 5 to 8, we now can draw a chart to illustrate the mechanics of graphite formation, as in Fig. 9. At Rate 1 (Fig. 9), graphite formation will begin and end in the Type A region.

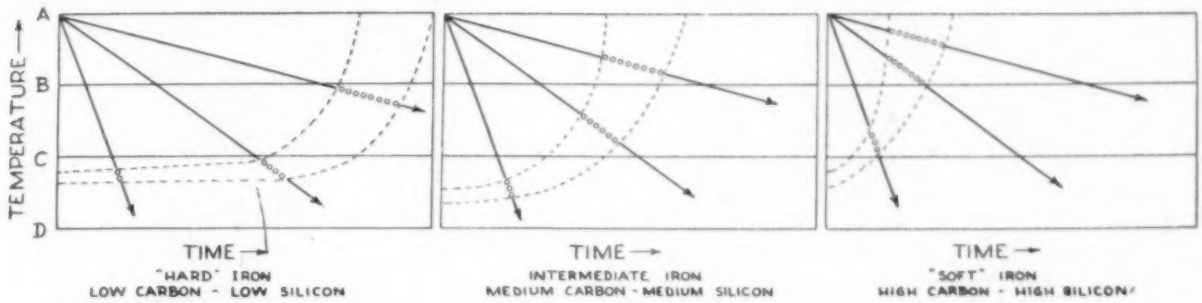
At a somewhat higher rate (2), graphite will form in the Type D and Type E region. At very high rates such as 4, graphite will be completely suppressed and white iron, free of graphite, will be formed.

Rates intermediate to these examples are also possible. For example, at Rate 3, some graphite will be formed in the Type D - Type E region, and the remainder of the car-



No.	*Rate*	*Product*
1	Low	Type A graphite.
2	Moderate	Type D or Type E graphite.
3	High	Type D or Type E graphite mixed with iron carbide (mottled iron).
4	Very high	No graphite (white iron).

Fig. 9—Solidification products formed at various cooling rates.



Forms Type D and Type E graphite on slow cooling; white iron with moderate or fast cooling. Very slow cooling required to form Type A graphite.

Forms Type A graphite with slow cooling, Type D or Type E with moderate cooling, and white iron with fast cooling.

Forms Type A graphite with slow and moderate cooling; Type D or Type E graphite at high rates. Requires very fast cooling to form white iron.

Fig. 10—Effect of composition upon graphite structure.

bon will form as white iron. It should be noted that transformations occurring in the white iron region result in the formation of "chill."

Irons which do not complete the formation of graphite in the Type A region (above boundary line B, Fig. 9) are referred to as undercooled or supercooled irons. It is the purpose of inoculation to prevent such undercooling and to cause all of the graphite to form in the Type A distribution.

Effect of Composition Upon Graphite Structure—Using the same type of diagram, the effect of changes in composition can be represented by a displacement of the "transformation zone." For a specific cooling rate there may exist an undercooled condition which produces Type D or Type E graphite flakes or massive cementite.

Low Carbon Iron

A "hard" iron (iron of low carbon content) will have a low transformation zone and will require very slow cooling to form Type A graphite. Even at moderate rates such an iron may form white iron. This situation is shown in Fig. 10.

A "soft" iron (high carbon, high silicon), on the other hand, will have a high transformation zone. Even moderate cooling rates will form Type A graphite flakes in such an iron; very high rates are required to form white iron ("chill").

One effect of the composition of the iron on the graphite structure, therefore, is to change the tendency of the iron to undercool. Low carbon, low silicon irons with a "low reaction zone" are quite prone to undercool, whereas high carbon, high silicon irons have less tendency to undercool.

Effect of Inoculation

Based on the hypothesis of the foregoing diagrams, the effect of inoculation may be stated quite simply — inoculation raises the "transformation zone." That is, inoculation may make an iron that would solidify (at a given cooling rate) as "white iron," or with Type D or Type E graphite, actually solidify with Type A graphite.

In this respect the inoculation effect is somewhat akin to "softening" the analysis of the iron, but instead of reducing tensile properties due to "softening" of the matrix, it actually improves properties due to the better distribution of the graphite flakes.

Stated in another way, inoculation permits the use of a stronger base iron in a given application with reduced tendencies to form chill (white iron) or undesirable graphite distribution patterns.

As low carbon, low silicon irons have considerable tendency to undercool, inoculation will be quite potent with such irons. However, high carbon, high silicon irons, which tend for the most part to solidify with Type A graphite anyway, benefit less from inoculation.

High carbon, high silicon irons are often poured into thin sections which introduce a high cooling rate which promotes undercooling. Therefore, inoculation is often as desirable for these irons as for the low carbon, low silicon types.

If a given inoculation treatment resulted in the formation of Type A graphite flakes exclusively at even the highest cooling rates, then such an iron could be called "completely inoculated." Present inoculation practices do not result in such "complete inoculation."

From the diagrams it can be seen that high cooling rates (fast cooling) promote undercooling. Therefore, with light sections the need for inoculation increases over the need in heavy castings (Fig. 11).

Inoculation Theories

Thus far in this discussion no attempt has been made to explain why inoculation has its peculiar effect upon the mode of formation of graphite in cast iron. Various investigators have proposed theories, but up to the present time no conclusive evidence has been obtained to establish any one theory.

The proposed theories overlap somewhat, but for the purpose of discussion may be designated as:

1. Graphite Nuclei Theory.
2. Degasification Theory.
3. Silicate Slime Theory.

According to the *graphite nuclei theory*, graphite particles in the melt act as nuclei to begin the graphitiza-

This paper was presented and discussed at a Gray Iron Session of the Fiftieth Annual Meeting, American Foundrymen's Association, at Cleveland, May 10, 1946.

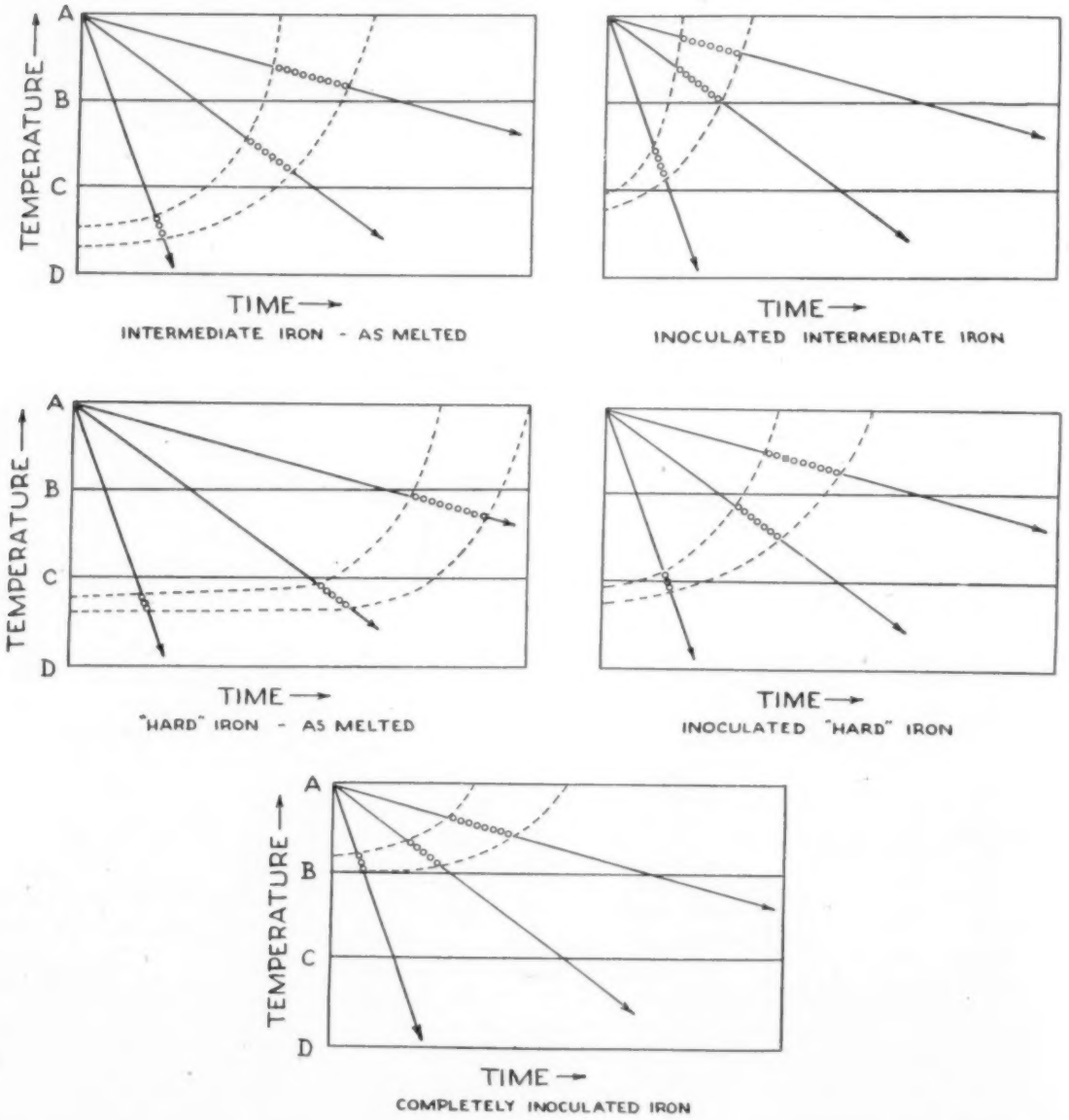


Fig. 11—Effect of inoculation on graphite formation in gray cast iron.

tion phenomenon. Such nuclei may be imagined as extremely small solid particles of graphite in the molten iron.

In order to form in the Type A distribution, graphite flakes must form on nuclei distributed through the melt. In the case of an iron which has solidified with Type D or Type E graphite (or as white iron), it has done so because it has passed through the temperature region of Type A graphite formation without forming any graphite flakes, and thus the graphite has formed in the lower temperature pattern.

Inoculation is considered here as the addition of effective graphite nuclei to the molten iron. The presence (or increase) of the number of effective nuclei makes the iron more likely to form Type A graphite at any given cooling rate.

The graphite nuclei theory is substantiated by the fact that the late addition of solid elemental carbon (granular graphite) to an iron normally solidifying with Type D or Type E graphite is known to change solidification to the Type A pattern.

Further evidence is the fact that certain alloys, which are efficient inoculants when added in the solid granular form, lose their inoculating ability when they are added as superheated molten alloys. Such superheating of the alloys is supposed to destroy the nuclei.

One investigator has attributed the potent effects of silicon-bearing inoculants to a local concentration of silicon; this concentration promoting a shower of fine graphite nuclei in the melt. It is interesting in this connection to note that most of the commercial inoculants contain appreciable amounts of silicon, a powerful "graphitizing alloy."

The tendency for superheated irons (irons heated considerably above the melting point) to form Type D and Type E graphite flakes in preference to Type A can be attributed (according to this theory) to the destruction of graphite nuclei

at high temperatures in the irons.

The *degasification theory* attempts to explain the mechanism of inoculation on the basis of reactions occurring between the inoculant and dissolved or chemically combined gases, such as oxygen, hydrogen, nitrogen, etc.

These "degasification" reactions may produce inoculation by (1) the elimination of "chill-forming" gases, and (2) the formation of inclusions in the melt which may act as effective nuclei.

It is known that inoculated irons show a "wearing off" effect. If molten iron is held for an appreciable time (of the order of 15 min.) between inoculation and pouring, the effect of inoculation gradually "wears off" and even may be eliminated.

Proponents of the degasification theory hold that this "wearing off" is due to reabsorption of the gases which were eliminated by the inoculation treatment.

It is a fact that most of the effective inoculants also are deoxidizers. On the other hand, other gases may be of more significance than oxygen.

The *silicate slime theory* provides for a submicroscopic "slime" or "pulp" of ferrous silicate inclusions which act as nuclei for the formation of coarse graphite. The theory is supported by the fact that a slag which removes silicon or ferrous oxide helps to promote the formation of fine graphite.

It has been pointed out that there is an analogy between this behavior and the popular belief that submicroscopic inclusions affect grain size in steel.

Nucleation Process

A certain similarity of these three theories is evident. The process of nucleation appears in each of them and, therefore, it may be supposed (for the present) that inoculation is at least partially the result of the addition of nuclei to the melt.

Numerous efforts have been made by various investigators to either prove or disprove each of the three theories but, as yet, no such effort has been wholly successful.

The microscopic or even submicroscopic size of the particles involved and the high temperatures at which the transformations occur impose major problems in the study of the solidification graphitization phenomena. The basic theory of inoculation is, therefore, still a debatable subject.

Conclusion

It is hoped that the foregoing discussion of inoculation will enable foundrymen to appreciate more readily the fundamental concepts behind the use of inoculants. The theory is not well-defined, as yet, but in its present state is sufficient to outline the graphitization transformations which occur when a cast iron is inoculated.

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DISCUSSION

Chairman: H. BORNSTEIN, Deere & Co., Moline, Ill.

Co-Chairman: V. A. CROSBY, Climax Molybdenum Co., Detroit.

DR. A. DI GIULIO[1] *(written discussion):* The author has presented a rather ingenious and over-simplified theory of the mechanism of inoculation. The hypothesis advanced must be accepted with reservations.

Purposely the author has avoided stating the relationship of the liquidus line in relation to the Zone Diagram illustrated in Fig. 5. Are we to understand (1) that the process of Graphitization is independent of the liquidus solidus range, (2) does Fig. 5 imply that Type A Graphite is always formed directly from the liquid?

In all diagrams the cooling rates have been represented as straight lines. In actual practice this is not the case, hence, how does the change in rate of cooling effect the graphite formation within any of the indicated zones.

The author states "It is the purpose of inoculation to prevent such undercooling and to cause all of the graphite to form in the Type A distribution." Does this imply then that inoculation is essentially the process by which all the graphite of iron is made to come out from the liquid, that is above the melting point?

It is known that the degree of undercooling in liquid-solid transformation depends largely on the rate of cooling to which the system is subjected. Also that the slower the rate of cooling, the less tendency for the system to exhibit the phenomenon of undercooling.

If we agree with the author that inoculation prevents undercooling, then why not make use of very slow rate of cooling to obtain the desirable type of graphite?

In the literature many cases have been reported of irons heated to excessive superheating temperatures, which have shown Type D Graphite regardless of the fact that they were allowed to cool and then were poured at the proper pouring temperature. How can this be explained on a basis of undercooling and does the author believe that inoculation will work under these conditions?

MR. LOWNIE *(answer to Dr. Di Giulio's written discussion):* Most of Dr. Di Giulio's comments are concerned with the controversial physical metallurgical issues of inoculation. As stated in the paper, the present-day theories have been previously discussed in considerable detail in Reference 1 and in a number of papers listed in the bibliography of Reference 3 of the present paper. The answers to many of Dr. Di Giulio's specific questions are contained in the paper.

The author agrees with Dr. Di Giulio that the mechanism of inoculation as presented in this paper is "over-simplified" and "must be accepted with reservation." Several sentences taken from the paper illustrate the author's feelings on this point:

1. "Although the system does make certain assumptions without confirmed metallurgical basis, it is believed that an attempt to make a graphical presentation will be helpful in discussing inoculation."

2. "It is necessary to emphasize here that these schematic sketches are only approximate 'behavior diagrams,' and may not be interpreted too closely. The value of the diagrams lies in the fact that they do provide a simple manner of representing graphitization behavior."

As stated by Dr. Di Giulio, the author *has purposely* avoided stating the relationship of the liquidus line to the Zone Diagram in Fig. 5. In connection with Fig. 5, it was explained that this relationship is still somewhat controversial and a quotation given from Epstein to summarize the present state of the controversy. It was stated in the paper that, although of considerable theoretical interest, accurate knowledge of the relationship of the liquidus line to the zones of Fig. 5 is not necessary to understand in a *practical* manner how inoculants work.

Quoting from the paper, "With reference to our simplified diagram, this simply means that the solidification temperature (or temperature range) is not definitely located with respect to the lines (or regions) B and C." This statement clearly avoids the implication that Type A graphite is always formed directly from the melt.

As the diagrams are purely schematic, cooling curves were plotted as straight lines in the interests of simplicity. Actual cooling curves are usually curved when

[1] Consulting Chemical and Metallurgical Engineer, Detroit, Mich.

plotted on rectangular coordinates and approximately straight when time is plotted logarithmically. As time units are not indicated on the diagrams, the abscissa may be considered as being plotted logarithmically when such a refinement is desired.

The very slow rate of cooling proposed by Dr. Di Giulio as a method for obtaining the desirable type of graphite certainly is not generally applicable to normal foundry procedures. Practically, the cooling rate is governed by the section thickness of the casting, the surface area of the casting, the pouring temperature of the iron, the heat capacity of the molding material, and other factors.

All of the diagrams contained in the paper show that if one *can* cool *slowly enough*, the desirable types of graphite will be formed. Usually, it is simpler to inoculate irons that cool at the faster rates rather than to provide elaborate and costly methods of changing the cooling rate.

In the last paragraph of his discussion, Dr. Di Giulio has overlooked the definition of undercooling as given in the paper. Certain types or grades of cast iron will undercool even when cooled at comparatively slow rates. The white irons used in the malleable iron industry are examples of such irons.

Whether inoculation will work under specific conditions depends upon the composition of the iron, the cooling rate, and the effectiveness of the inoculant and the inoculating practice. The general beneficial effects of inoculation are well known to engineers familiar with gray iron foundry practice.

MEMBER: How long will the ladle inoculant be effective?

MR. LOWNIE: A definite answer depends upon the temperature of the metal, the particular inoculant being used, and your measure of "effectiveness." Inoculants vary in their rate of deterioration. As a general figure, subject to many exceptions, I suggest 15 min.

MEMBER: I am concerned with the handling time, i.e., in handling the ladle from the bull ladle to the molding floor.

MR. LOWNIE: A definite answer of so many minutes will have to be worked out under your particular conditions. In general, you should handle the inoculated metal as rapidly as possible.

If you are pouring light castings you will want to handle the metal rapidly for two reasons, (1) to retain as much temperature as possible, and (2) to retain high effectiveness of the inoculant.

If you are pouring heavy castings and the ladle is to be filled from several taps from the cupola, hold all inoculation for the last tap. That gives you the maximum amount of time for handling the ladle.

MEMBER: What is the time limit on that? In other words, how cold can you pour the metal and still get the effect of the inoculant?

MR. LOWNIE: I do not want to venture into that at this time. It is an entirely different subject.

T. E. EAGAN[1]: It is always the best practice to pour inoculated irons as hot as possible. By doing so the best graphite distribution is obtained.

MR. LOWNIE: You usually want to do that because the mere fact that you are inoculating means that you are trying to obtain quality castings. Higher metal temperatures generally improve quality. Solution of the inoculant also is improved at higher temperatures. However, I would not want to specify a minimum temperature unless I was familiar with the particular job and its conditions.

MEMBER: But assuming that you have a production line and you are counting on the inoculant to enhance the value of your casting, at what temperature would you pour to get the full value? We are using a holding furnace and are going to pour at 2600° F., would that be sufficient?

MR. LOWNIE: That is a good temperature. I know of cases where excellent results are obtained below 2600° F.

Actually we still know very little about the mechanism of inoculation. For the present, specific problems must still be worked out individually. The tapping temperature, the pouring temperature, the thickness of the critical section of the casting, the design of the casting, the mold conditions, the particular inoculant being used, and the composition of the base metal all affect the results that you obtain. Theories and past experience point the way to good practice but answers as to specific times and definite temperatures are best worked out under the specific conditions involved.

J. H. SCHAUM[2]: A soft iron may have many of its graphite nuclei dissolved by holding it at relatively high temperature for a period of time in an electric or air furnace. This refinement causes the soft iron to precipitate Type D or E graphite when slowly cooled in the mold.

MR. LOWNIE: That is correct. Electric furnace irons are particularly prone to form modified graphite. Many such details are not included in the paper.

W. A. SPINDLER[3]: Do you feel that the graphite nuclei theory covers the theory adequately?

MR. LOWNIE: There are unexplained discrepancies in each of the proposed theories. I, personally, cannot offer one in preference to the others. However, I often use the graphite nuclei theory because it seems to be the most readily understood by many people; not because it has the most basis in fact.

MEMBER: At what minimum temperature can you inoculate satisfactorily? How low can you inoculate and still get satisfactory results?

MR. LOWNIE: Try to hold inoculating temperatures at 2600° F. or higher and you will probably be safe. However, some people are obtaining good results at lower temperatures. It depends upon the job.

MR. EAGAN: I have seen heavy sections poured at pretty low pouring temperatures. The idea is to pour as hot as you possibly can pour to get the best effect.

E. V. SOMERS[4]: We attempt to maintain our tapping temperature at 2700° F. and above by means of the optical pyrometer.

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[1]The Cooper-Bessemer Corp., Grove City, Pa.
[2]Naval Research Laboratory, Washington, D. C.
[3]University of Michigan, Ann Arbor, Mich.
[4]Westinghouse Electric Corp., Trafford Foundry, Trafford, Pa.

MOLDING SAND BINDERS

This paper deals with fundamental theory of molding sand binders; causes of bond properties and strength; forces active in the green state between sand and clay and between sand and organic binders; viscosity of binders and its influence on green compressive strength; and simultaneous use of various binders.

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PROPERTIES OF MOLDING SANDS which are of interest to the foundryman, as given by Dunbeck[1], are grain size, green, dry and hot compressive strengths respectively, permeability, durability, flowability, moisture content, resilience, expansion, contraction and sintering point.

These properties are needed in different degrees according to use. Naturally bonded sands are always sufficient if they have a certain grain size, permeability, expansion and refractoriness as initial properties. In case naturally bonded sands do not develop the other desired properties, a suitable admixture is made to the sand as required.

All admixtures, as well as the corresponding substances in naturally bonded sands, have one essential feature in common; they connect the sand grains into a system whose strength, produced by ramming, enables both the mold and the cooling casting to safely sustain the stresses to which they are subjected.

In testing practices these stresses are characterized by the green, dry and hot compressive strengths corresponding to the temperatures resulting from the cooling casting. At the same time these admixtures have to produce in the sands many other properties if the sands are to be regarded as suitable for foundry use. This applies to inoragnic materials such as clays, and to organic substances used for the production of cores.

Causes of Bond-Properties and Strength. To achieve a certain coherence among the sand grains, attractive forces have to interact. For the coherence of two bodies in the dry state the following conditions have to be assumed; that (1) the largest possible part of each body is in contact with the other and (2) at the same time, distances between their surfaces are of the same order as the distances between atoms or molecules of the bodies considered.

Van der Waal Forces

Under these conditions the forces which interact to hold the two bodies more tightly together are greater as the size of the bodies become smaller and as the proportion of surface and weight becomes greater. For inorganic substances these forces are most frequently represented by electrostatic forces between ions or dipols, and for organic materials, without dipolar moments, these forces are Van der Waal forces.

The forces producing coherence between silica sand grains are negligible and of no importance because the grains are too large and their surfaces too irregular. The specific task of every binder is to use most effectively the inherent surface forces present upon each sand grain. To fulfill this task every binder has to possess the following properties:

1. It has to wet the surface of the sand grain as much as possible.
2. The *adhering* forces between the binder and the sand grain have to be as intense as possible.
3. The *cohering* forces within the binder itself must be of the same order as the forces of (2).

In practice, property (1) is most frequently accomplished by adding the binder in the liquid state. To accomplish the other two properties the binder must *contain* or *develop* the maximum number of particles of colloidal dimensions. The surface-weight proportion is optimum when those colloidal particles take a plate- or fibre-like form.

Bond clays (inorganic materials) and all water-dispersed organic binders (such as dextrin and starch) used for bonding molding sands contain such colloidal particles. Such colloidal particles develop during the setting of cement and during the baking of organic binders, and especially during the polymerization of oils such as linseed.

For sands in the "green state" the adhering forces are interacting in the presence of a liquid. These forces function in the presence of water, but little, if any, in the presence of oils such as linseed.

Forces Active in the Green State. Between sand and clay. In this case we are concerned with the existence and interaction of forces between silica grains and clay, and between the particles of clay themselves in water. U. Hofmann and A. Hausdorf[2] have studied these forces of Na-bentonites (electrodialyzed Ca-bentonite) with various exchange cations.

In the presence of water the elementary particles of Na-bentonite are surrounded by a diffusion double-layer. In the presence of water the exchange cations (1) try to diffuse through the double-layer and

Presented at a Sand Research Session at the Fiftieth Annual Convention of the American Foundrymen's Association, May 8, 1946, at Cleveland.

(2) try to separate themselves, aided by an increase in thermic movement, from the negatively charged surfaces of the silica layers of the clay lattice to which they are attracted by electrostatic forces.

According to Debye - Huckel's theory, the extent of this diffusion double-layer, i.e., the average distance of the cations from the surface, is given by the quantity $1/k$ with the dimension of length (see Fig. 1). The distance depends upon the concentrations of the same cations in the solution, and it diminishes with increasing concentration of the cations.

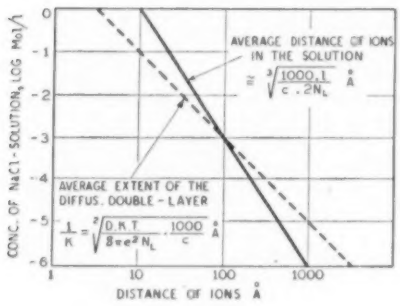


Fig. 1—Relation of the average distance between ions in solution and the relation of the average depth of the diffusion double-layer both with respect to the concentration of NaCl solution.

Also the average distance separating the ions in solution diminishes with increasing concentration of the cations, but at not so rapid a rate as $1/k$. With greater concentrations, two surfaces can approach so closely that an intermixing of the two diffusion double-layers results. The cations spread equally between both negatively charged surfaces and bind them as illustrated in Fig. 2.

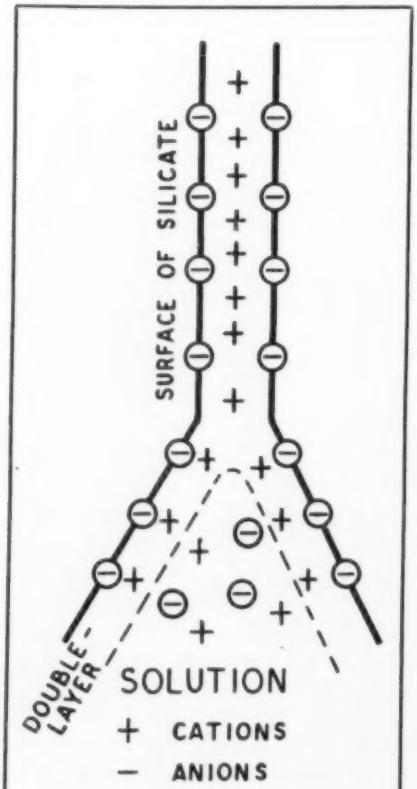


Fig. 2—The uniform distribution of the cations between two silicate surfaces of montmorillonite on the assumption that a sufficient concentration of the loss is present in solution.

Binding Action Explained

In order to explain the binding action between silica grains and Na-bentonite, we can assume that even the surface of the silica grain contains a small number of exchange cations, so that a mixed diffusion double-layer is formed in the presence of water. In any case, Si-O dipols are present on the silica surface which by reason of their electrostatic dipolar forces attract the cations of the diffusion double-layer.

Since only a relatively small amount of water is involved in molding sands, let us consider the extent to which the above described mechanism applies. In this water other cations usually Ca and Mg are present along with the Na cation of Na-bentonite. According to Debye-Huckel's theory the quantity $1/k$ is true only for monovalent cations when the distances involved are greater than 25 A. We, however, are interested in distances less than 25 A.

The above described mechanism is true for not only Na-bentonite but for all clays and is at least qualitatively supported by the following facts. According to Grim, Bray and Bradley[3], the green compressive strength of molding sands is dependent upon the amount of exchange cations and as this strength increases the number of cations increase.

With mixtures of constant Na-bentonite content at increasing moisture contents, the green compressive strength increases at first because the silica grains can be more easily moistened and diffusion double-layers can develop in the clay. As we continue to increase the amount of water, the green compressive strength attains a maximum value and a further increase in water content brings about a decrease in this strength because the diffusion double-layers are unfavorably changed due to a diminution in the cation concentration.

From this point of view there is no justification for the use of soft water, a practice sometimes recommended when bentonite is used. Laboratory tests (Fig. 3) prove that the green compressive strength of a mixture bonded by Na-bentonite increases slightly in relation to the hardness of water.

Consideration is seldom given to the influence of anions often regarded as insignificant but which nevertheless may be unfavorable. For example, an excessive admixture of 2.3 grams of Na_2SO_4 to 100 grams of Na-bentonite produces a decrease in green compressive strength from 300 g./cm². to 150 g./cm². (4.3 psi. to 2.15 psi.).

Forces Active in Green State

Between sand and organic binders. The green compressive strength of pure silica sands bonded with organic binders is generally lower than clay bonded sands and most frequently this strength cannot be increased by increasing the amount of binder.

Water-dispersed organic binders have less green bonding power, rarely exceeding 100 g./cm². or 1.4 psi., than clays, and oil bonded sands have the lowest green compressive strength (20-40 g./cm². or 0.29-0.58 psi.). When organic binders are used the interacting forces (Van der Waal's) are different; nevertheless, after drying, a high compressive strength is achieved by mixtures bonded with organic materials, and this strength is at least that produced by bond clays.

Experience has indicated that a certain increase in the green compressive strength can be achieved by increasing the viscosity of the sand mixture which is limited by the technical conditions of mixing the sand. Above a certain viscosity binders do not mix well with sand and consequently previously mentioned con-

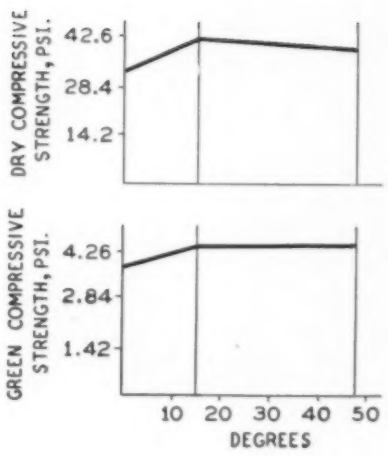


Fig. 3—Influence of the hardness of water on the green and dry compressive strengths for a mixture of 5 per cent Na-bentonite and 4 per cent water.

dition of perfect adhesion (No. 2 under "Causes of Bond-Properties and Strength" is not fulfilled.

The influence of adhesion, at a lower degree of viscosity, can be more readily observed in its effect upon the dry compressive strength, rather than the green compressive strength. Since various binders behave differently in this respect, it seems useful, therefore, to consider the problem of viscosity of organic binders and bond clays at the same time.

Viscosity of Binders and Its Influence on Green Compressive Strength. Assuming that the silica sand grains are uniformly wetted, each grain will be better enveloped by the binder at lower viscosities of the binder. Because of this, some organic binders of high viscosity at room temperature are usually admixed at a higher temperature. It is also well known from practice that the admixing of some binders to the sand depends upon the type of mixing equipment used.

There is naturally a great difference in mixing Na-bentonite in the presence of water by means of a shovel mixer and by means of the Eirich, Simpson or other muller-type mixers. Mixing in the muller-type mixers is accomplished under the influence of higher shearing stresses.

Influence of Shear Stresses

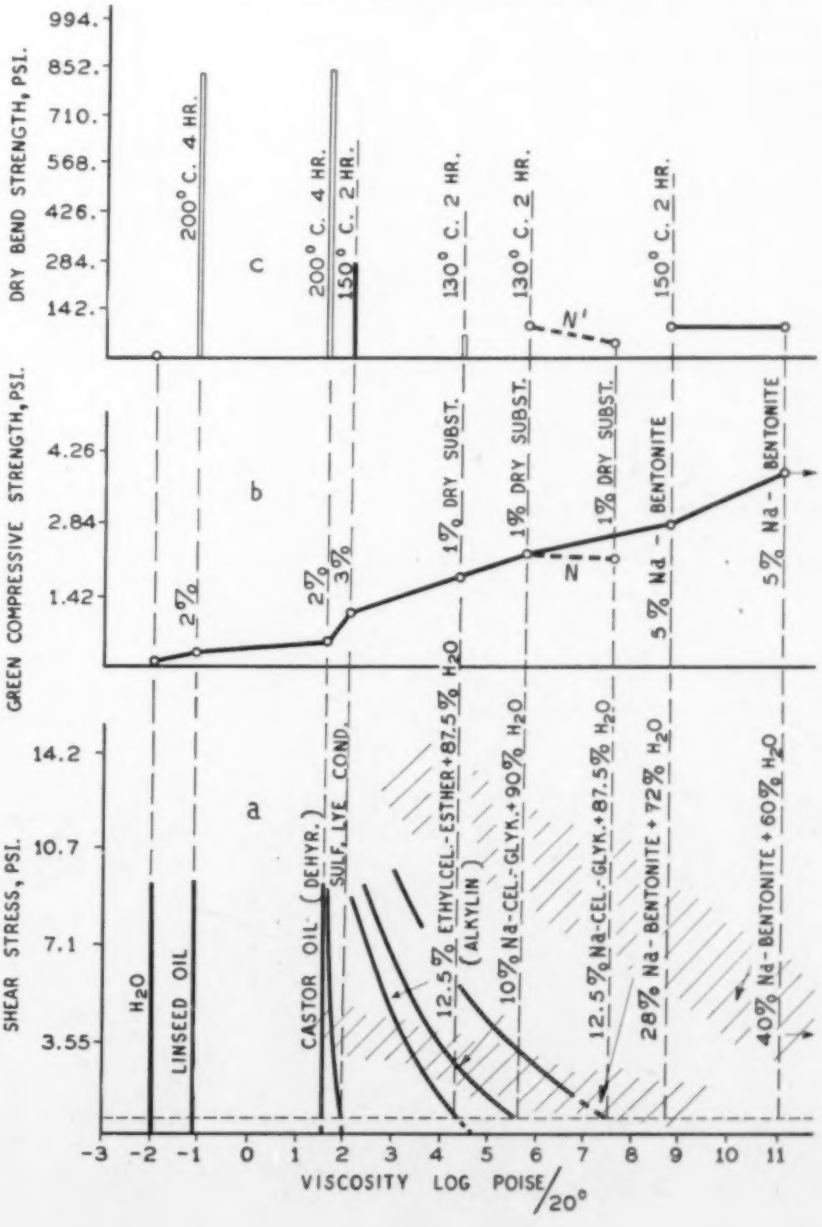
For comparison, Plesinger[4] shows that after mixing in a muller-type mixer, a sand containing the optimum admixtures of Na-bentonite and water gave a green compressive strength 20-60 per cent higher and a dry compressive strength 60-300 per cent higher than the same sand mixed by a paddle-type mixer. The influence of shear stresses is evident from the well-known viscosimetry of colloidal substances.

It appears difficult to follow the influence of shear stresses on the viscosity of some binders, especially for large particle clay suspensions at concentrations which occur in their practical applications in molding sands.

Up to now the consistometer of Hophler[5] has been used for this purpose. The suspension to be tested is enclosed in a hollow steel cylinder of known diameter. A body, generally a small ball, is pressed into the suspension under different pressures, and the time measured for the ball to traverse a given distance. Thus the velocity of flowing is measured.

In clay suspensions, especially under large pressures, water is

Fig. 4—The relation between viscosity and (a) shearing stress, (b) green compressive strength and (c) dry transverse strength for water, linseed oil, dehydrated castor oil, condensated sulfite lye, ethyl-cellulose-ester, sodium-cellulose-glycolate and Na-bentonite.



pressed out by the ball and the surface of the ball is lubricated by the water. The amount of water in the suspension changes with the distance traversed by the ball and, consequently, the measured time values change as well. Owing to war conditions it has not been possible to introduce a more satisfactory method and the dependence of velocity of flowing on shear stresses could be measured only for some suspensions and then only approximately.

Flow Values

The values of viscosity which may be obtained from the curve of flowing are, under these circumstances, significant only for general orientation. The differences between the flow values of the various binders in suspension are so large that they shift into the background the theoretical uncertainties and lack of exactness originating from the testing procedure. However, the flow values obtained do permit us to make some interesting conclusions.

Figure 4 shows a summary of results obtained. Figure 4a shows the relation of viscosity to shear stress for some well-known binders (linseed oil, Na-bentonite, etc.) and also for some materials that may come into consideration as binders (Na-cel-glyk and ethylcel-ester).

The viscosity of materials such as oils do not depend upon shear stresses. Na-bentonite is one of those substances which, when in suspension, exhibits diminishing viscosity with increasing shear stress. This phenomenon is described as structural viscosity or, more generally, as quasi-viscosity. At small shear stresses some binders behave almost like a solid and their viscosities approach infinity.

Figure 4b shows the relation between green compressive strength and the measured viscosity at its maximum value. Figure 4c shows the corresponding values for the dry transverse strength. In all tests the viscosities of the binders were measured for concentrations generally found in foundry sand mixtures and these concentrations were admixed with the sand. Because of previously mentioned experimental difficulties the testing of Na-bentonite suspensions required a relative increase in the amount of water.

In general, for all types of materials, we have to assume that their necessary uniform distribution can be achieved only when the viscosity does not exceed a certain limit related to any material in question. When a particular limit is exceeded, the uniformity of distribution necessary for green compressive strength is incomplete and green compressive strength is diminished.

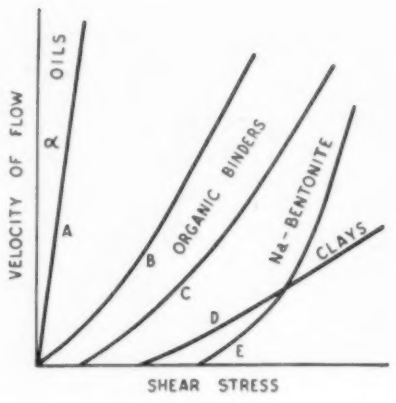


Fig. 5—Schematic representation of the relation of velocity of flowing to shearing stress for various types of binders.

For a binder which exhibits quasiviscosity the necessary uniformity of distribution of that binder can be achieved by increasing the shear stress if viscosity at small shear stresses is too large. If at a given shear stress, the limit of mixability for any mixer is exceeded by increasing the concentration of the binder, the resulting green compressive strength of the mixture decreases. This is shown in Fig. 4, with sodium-glycolate of cellulose (Curve N, Fig. 4b).

Due only to this influence of shear stress, it is possible to mix the Na-bentonite suspension and to explain the influence of different kinds of mixers upon the achieved strengths according to experience from practice. A lower viscosity achieved by warming, results in better mixing of all binders, especially viscous oils, resinous materials and pitch. The warming of certain binders is a general foundry practice.

Viscosity Relationship

Figure 4b shows that the relation between green compressive strength and viscosity has more general importance, since this strength usually increases as the viscosity increases. This relation is more evident when we consider the character of flowing given schematically in Fig. 5 for various types of binders.

For the coordinate system used in Fig. 5, viscosity is determined from the tangent to the curve in question at any particular shear stress. The curves are of different character and become straight lines for pure viscosity (after Newton) which occurs with water and oil. The curved lines in Fig. 5 indicate that for certain water-dispersed organic binders, viscosity varies with shear stress.

Influence of Thixotropy

When a curve passes through the origin of the coordinate system, a movement approaching that of flowing occurs at very low shear stresses. If the curve begins to rise from a certain shear stress, the binder up to this stress behaves as a solid. The position and character of the curves depend upon the temperature and on the duration of shear stress, as mentioned later in connection with the study of the influence of thixotropy.

From theoretical considerations, materials which behave as solids up to the greatest possible shear stresses will possess the best properties for the development of green compressive strength; and if at the same time these same materials have the lowest possible viscosity they will possess the best properties for admixing into the sand and for ramming the sand. Figures 4 and 5 show that of the materials mentioned, Na-bentonite more closely approaches the theoretical conditions.

Some macromolecular organic substances which exhibit swelling in the presence of water have qualities approaching those of Na-bentonite, and achieve excellent strength after drying. A suspension of Na-bentonite behaving as a quasisolid material at low shear stress produces a sufficient green compressive strength in the sand; however, at a higher shear stress the viscosity rapidly decreases and the Na-bentonite can be well mixed into the sand.

In this respect other bond clays are characterized by a less favorable behavior as shown in Fig. 5. It is not possible at present to directly compare the influence of shear stress on the binder alone and a mixture of the binder and silica sand. The sand mixture can be classed as a

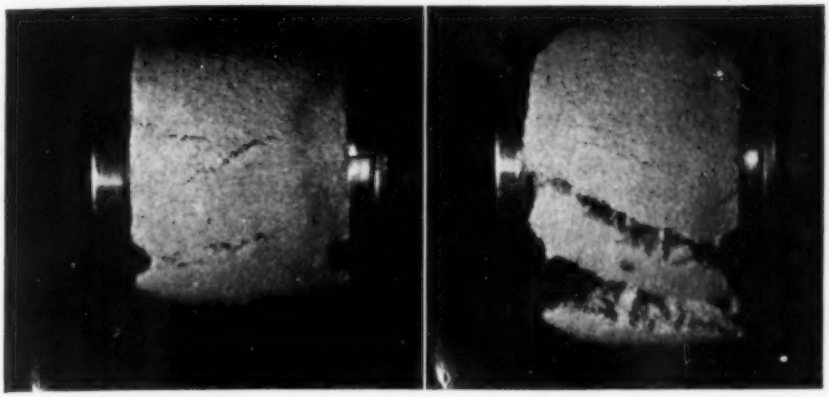


Fig. 6 (Left)—Compression test on "green" sand bonded with one per cent ethyl-cellulose-ester (starch). Green compressive strength 130 gm./cm.² (1.85 psi.). Deformation just before fracture 10 per cent.
Fig. 7 (Right)—Same as Fig. 6, but at instant of failure. Deformation 14 per cent.

very rough polydisperse system. The influence of the properties of the silica sand, such as grain size, grain distribution, and form and quality of the grain surface is evident.

A pure silica sand mixed with a purely viscous binder (such as shown by curve A, Fig. 5) shows a certain green compressive strength, however small. The sand mixture behaves as one system according to curve B or curve C of Fig. 5, providing the concentration of binder does not exceed a certain limit above which the sand mixture flows.

The individual characteristics of the binder can best be observed by watching the movement of grains in "green sand" during a pressure test. The movement of the grains would be recorded best by filming. Figures 6 to 9 show the more viscous behavior of sand-Alkylin system (mixture) as compared to the sand-(Na-bentonite) system.

The influence of various amounts of different clays upon silica flour relative to the coefficient of internal friction has been studied by Endell[6] and collaborators. *Although* a different physical quantity is concerned, the results are instructive from the point of view considered. Na-bentonite more than other binders, reduces the coefficient of internal friction. A very fine silica flour can serve as an example to show how the properties of binders can be influenced.

Concerning their mutual influence as to mechanical properties such as permeability and dry compressive strength, we may refer to the work of Morey and Taylor[7] and to Fig. 11. The causes of the different characters of the curves of Fig. 11 will be sought in the simultaneous presence of different clays and in the influence of water upon their swelling abilities with the object of explaining the green compressive strength curves and especially the different positions of maximum permeability.

The swelling of binders from the point of view of molding sands has been studied by H. Reininger and H. Jentsch[10], K. Endell and Collaborators[9], and Piper[11]. The swelling mechanism cannot be regarded as the principal cause of the binding properties of materials found with or admixed with molding sands. This is indicated by the circumstance that there does not exist any principal difference in the manner of admixing the binder as is shown in Fig. 12 for the addition of Na-bentonite and water to sand.

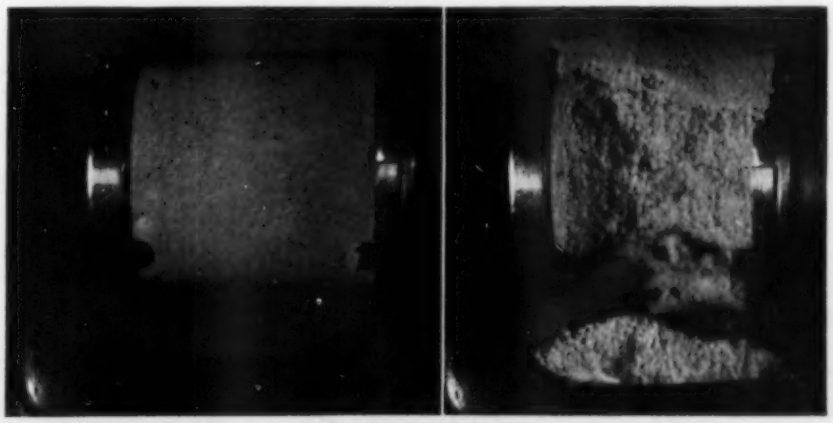
The swelling of the binder and the certain character of viscosity are but different manifestations of the binder concerned with regard to the size of the elementary particles, the shape of the particles and the action of specific surface forces, etc. Thixotropy, one of many other properties of this type, has been found very useful for explaining the behavior of colloidal binders in sand mixtures.

Thixotropy and Green Sand. Thixotropy is most frequently defined as the process describing the ability of a colloidal gel to change, upon shaking or mixing, into a relatively mobile mass, which reverts slowly again to the gel condition when in a state of rest. Thixotropy is an especially clear illustration of structural viscosity, because in the case of small shear stresses, they will be found to have no effect upon the gel but, in the case of high shear stresses (for example, produced by shaking), the gel becomes liquid.

Thixotropy Defined

According to Winkler[12] thixotropy is well defined as the proportion of the volumes of liquid and solid matters which is required, one minute after shaking, for the suspension to just flow out of an 8 mm. diameter test tube, when the test tube is

Fig. 8 (Left)—Compression test on "green" sand bonded with 5 per cent Na-bentonite. Green compressive strength 260 gm./cm.² (3.7 psi.). Deformation just before fracture 0.0 per cent.
Fig. 9 (Right)—Same as Fig. 8, but at instant of failure. Deformation 18 per cent.

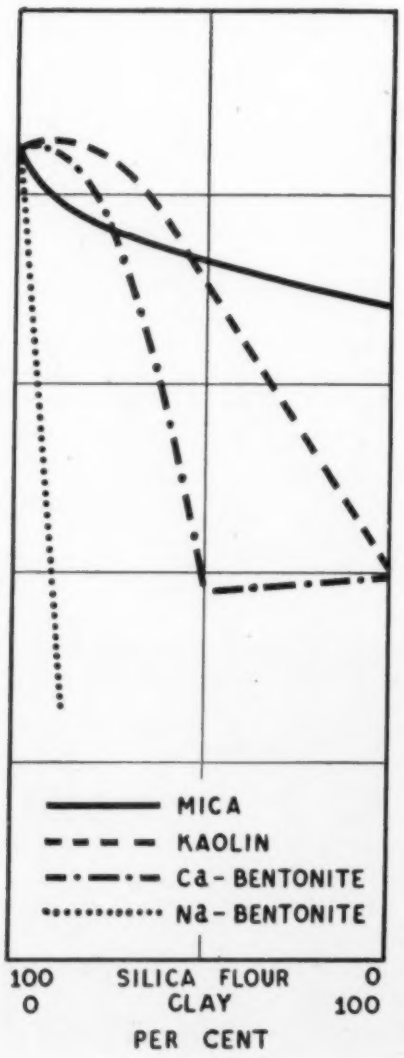


changed from a vertical to a horizontal position.

U. Hofmann[13] has conceived a model which is useful in explaining thixotropy of Na-bentonite. Hofmann imagines that under certain circumstances the layers of Na-bentonite in water are arranged as a house built of cards. Thixotropy, therefore, attains its maximum value when the particles of Na-bentonite, without losing contact with one another, occupy the greatest possible volume.

Figure 13 shows the influence of various conditions of mixing Na-bentonite with sand upon the value of thixotropy, measured according to the method of Winkler. It will be clearly seen that mechanical action such as mixing Na-bentonite with sand in the presence of water produces very favorable results due to the swelling and distribution of the binder.

Fig. 10—Relation of different clays admixed with silica flour and the coefficient of internal friction.



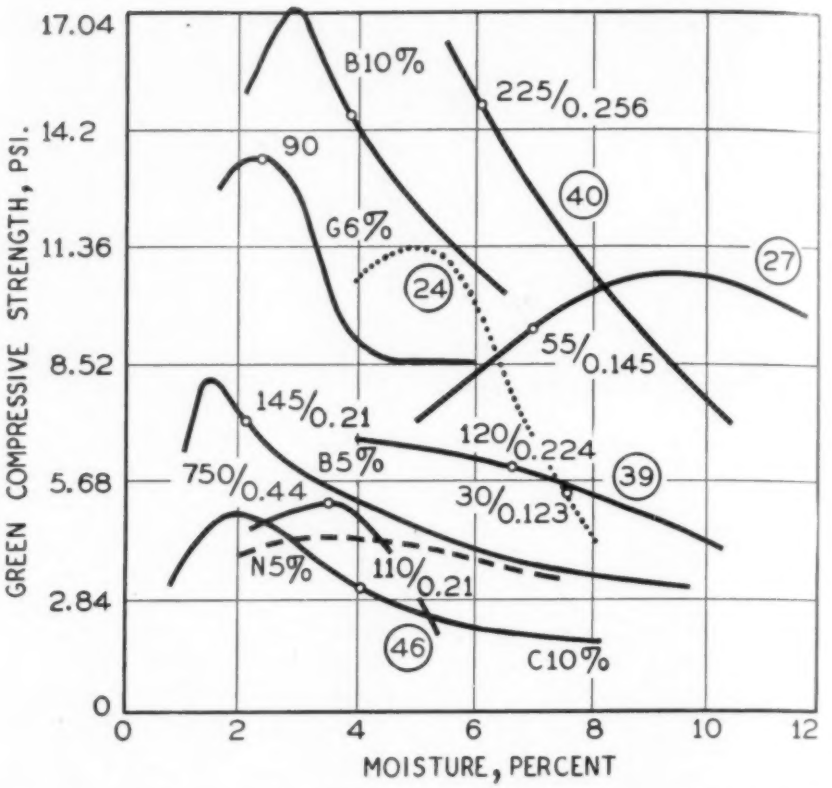


Fig. 11—Green compressive strength of different sands and the point of maximum permeability.

Curve B—Na-bentonite according to Morey and Taylor.[7]
Curve C—Clay according to Morey and Taylor.[7]
Curve G—Activated Na-bentonite from Geisenheim according to Endell.[7]
Curve N—Current activated German Na-bentonite.
Czechoslovakian naturally bonded sands according to Pisek and Holman.[8]

	Clay Content Per cent
Curve 24—Rudice	15.9
Curve 27—Stepanov	14.5
Curve 39—Blansko	10.6
Curve 40—Rosice	14.1
Curve 46—Kobylisy	9.2

Note: In a designation such as 90/0, 145 the first number gives the point of maximum permeability, the second number the average grain size for the sand used.

Thixotropy has also been measured for the same mixture for various orders of admixing the ingredients. Thixotropy changes with the resulting space distribution of Na-bentonite and parallels the course of green compressive strength. At the same time it can be demonstrated that thixotropy indicates very sensitively the *quality* of Na-bentonite, and that it can serve in practice for the evaluation of various deliveries.

Thixotropy can exert considerable influence upon the green compressive strength because of the dependence of this strength upon the time factor. In the case of Na-bentonite, green compressive strength measured immediately after mixing can be one-third less than it will be some hours later. In conducting such tests all outside influences such as drying-out have to be eliminated.

Some macromolecular organic materials behave similarly. These materials are always sensitive, in varying degrees, to the time of mixing, because their molecules can be destroyed by the grinding effect of sand grains. This can be demonstrated experimentally by centrifuging a suspension of such substances through a layer of sand resulting in a decrease of viscosity even though the composition remains the same.

Strength After Drying. The term "drying" has different physical, chemical and mechanical interpreta-

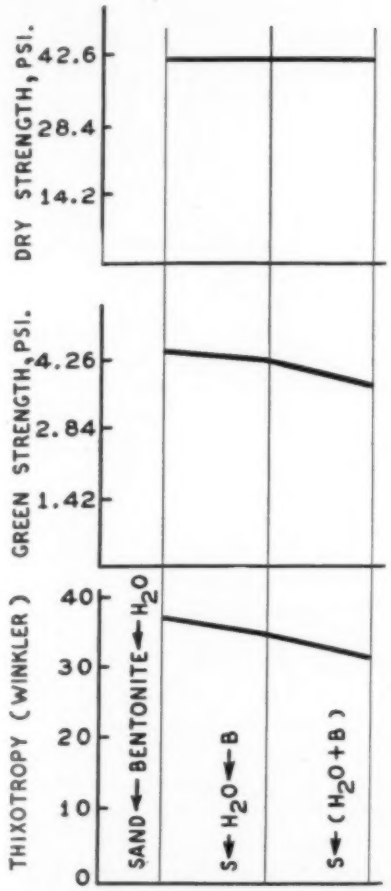


Fig. 12—Influence of the order of admixing Na-bentonite and water to sand on the strength properties.

tions with respect to the particular binder being considered. It is necessary to distinguish between the processes by which a binder develops its strength after drying. The processes are

1. Simple mechanical desiccation during which the binder does not change chemically and,
2. Chemical change in the binder and,
3. Physical change marked by losing or binding of water.

In the first case capillary pore water is always removed and it does not matter if this is done in vacuum at room temperature or by the application of temperatures up to the boiling point of water at normal barometric pressures, this latter method being the only practical way. All clays and some water-organic binders are dried this way.

In the second case the binder changes chemically, the action being accelerated by the application of temperatures above that of the boiling point of water. A classical example is the polymerization of linseed oil. Under the influence of varying higher temperatures different reactions of the binder will take place.

For example, the condensation product of sulfite lyes with formaldehyde and urea can be completed during the drying according to curve A, Fig. 14. The influence of higher temperatures upon the "dry" transverse strength after 2-hour drying period is indicated by curve B of Fig. 14, for the precondensed products of sulfite lyes with formaldehyde and urea.

The third process of drying involves (1) the loss of water of crystallization accomplished above the boiling point of water, for example in clays, and (2) the binding of water (hydration) to the binder particles, for example, portland cement.

Drying Phases

When chemical reaction is necessary to produce dry strength the drying requires two phases. During the first phase the mechanically held water is removed and in the second phase the chemical reaction occurring results in an increase in strength, providing the binder is not destroyed because of too high a temperature. Figure 4 shows the difference between binders requiring only the removal of mechanically held moisture and those binders requiring a chemical reaction for the development of dry strengths.

Upon desiccation the particles of the binder are enabled to approach one another and the previously mentioned attractive forces grow more effective than when water is present. It is very important (1) how the binder particles *cohere* at their surfaces and (2) how the binder particles *adhere* to the surface of the sand grains. Research upon various kinds of bentonite by the electron microscope has shown that Wyoming bentonite is probably more adaptable since after being dried it forms a stronger, more cohesive skeleton.

When more water than that required to produce the optimum green compressive strength is admixed, the dry compressive strength will be found to increase. This is common for all organic binders and is caused by better distribution of the elementary particles. Viscosity and physical-chemical surface effects must be considered for studying organic binders.

For example, on experiment with an asphalt emulsion (35 per cent asphalt, 8 per cent Tall oil, 3 per cent NaOH 38°Be and remainder water) has shown that the emulsion is completely separated into constituents upon contact with pure silica sand and the dry transverse strength is zero. This behavior follows experience with highway-emulsions of *oil-in-water* type, for distribution in pure silica sand is much worse than distribution in crushed stone which is produced from minerals containing compounds of bi- and trivalent metals.

It has been possible to stabilize the tested emulsion partly by (1) admixing 5 per cent Na-bentonite which results in a dry transverse strength of 4.5 kg./cm.², or completely by (2) immersing the sand in a 0.5 per cent soap solution which results in a dry transverse strength of 8 kg./cm.². Foundries *generally*

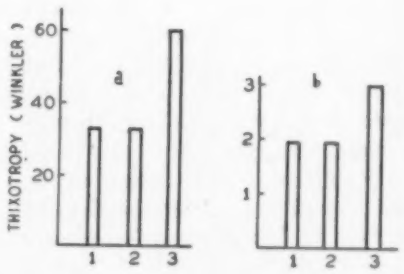


Fig. 13—Values of thixotropy (according to Winkler) for sand and Na-bentonite mixtures reported as follows: (a) Ratio of volumes of water to one unit volume of Na-bentonite and (b) Ratio of volumes of water to volumes of sand plus Na-bentonite.

Column (1) Sand and 5 per cent Na-bentonite as "dry" materials hand-mixed for 20 min.,

Column (2) Same proportion of materials mixed "dry" for 20 min. in a muller-type mixer and,

Column (3) Five per cent water added and "wet" mixed for 20 min. For the third condition, thixotropy was determined after the addition of 33 volumes of water, agitating, and a setting period of 15 hr.

use emulsions of the *water-in-oil* type and the previously mentioned separation into constituents does not occur with this type of emulsion.

Of all the factors involved in the attainment of dry strength, viscosity is the most evident, and one general statement pertaining to this production of dry strength by organic binders is as follows: When the viscosity exceeds a certain upper limit, depending upon the binder and mechanical mixing equipment the dry strength continues to decrease at constant amounts of binder. This is well shown in Fig. 14c in the case of sodium-glycolate of cellulose and in Fig. 15 for the condensate of sulfite lye.

Factor w/b. For the example cited in Fig. 15 the optimum values of dry transverse strength can be achieved for each amount of binder at a constant proportion of total water in the mixture and binder. This relation is given by the proportionality factor w/b.* For the example cited, w/b = 0.4 and the binder itself had a water content.

If the green compressive strength and the dry transverse strength of any binder was dependent solely upon viscosity, it would have to be expected that for various amounts of the binder the maximum strengths would be obtained at constant value of w/b. This has been found to be true for the green compressive strength.

With some small deviations the factor w/b is 0.26 for the data reported by Morey and Taylor[7] who studied the properties of Na-bentonite bonded mixtures. From the dry compressive strength data of Morey and Taylor, w/b was calculated as 0.54 when the relative of some certain per cent of binder with water ceases to be linear. If we desire an exact expression for w/b it is always well to note the strength and permeability values at definite amounts of binder or water in relation to the factor w/b.

Relationships which otherwise would have to be expressed by a three-dimensional diagram, are thus quite easy to comprehend. In practice it is not feasible to compare several three-dimensional diagrams, which would be necessary when one is obliged to seek a compromise percentage of water sufficient for dry transverse strength but not too high for ample green compressive strength. The comparison of two-dimensional diagrams (simple coordinate systems) giving the relation for the factor w/b is very easy.

From these two-dimensional diagrams it is always evident that for a given amount of binder the value of the factor w/b is always much lower and the viscosity higher after obtaining the maximum green compressive strength in comparison with a lower viscosity and higher value of the factor w/b after obtaining the maximum dry transverse strength.

Uniform Binder Distribution

Frequently the latter strength value cannot be attained because in the "green" condition the sand flows. As a result we are primarily concerned with increase of internal friction within the binder for the production of green compressive strength. The binder must be uniformly distributed between the grains. On the other hand, to achieve the highest possible dry transverse strength, the grains must be perfectly enveloped, for only then can the binder achieve a strong continuous binding of the grains.

Simultaneous Use of Various Binders. The simultaneous admixing of two or more binders provides properties not attainable from the use of a single binder. Most frequently, practice in Czechoslovakia has found it necessary to increase the strength properties in the "green" state of core mixtures binded by organic materials such as oils (which produce high strength properties upon drying).

Mere experience leads most frequently to quite negative results, proven by the example given in Fig. 16, where an emulsion oil (water-in-oil) has been literally spoiled by admixing a naturally bonded sand. Figure 17 indicates much better results can be obtained by the simultaneous use of Na-bentonite with sulfite lyes, or even better, condensation products derived from them. Such mixtures have given satisfactory results for the production of blown cores. The use of Na-bentonite with a cereal binder such as corn flour belongs in this class.

In all cases of the simultaneous use of a clay and a water-dispersed-organic binder, the clay is the decisive factor for green compressive strength. The eventual increase in green compressive strength upon the addition of an organic binder is fictitious for the organic binder absorbs a part of the water. The reduced amount of water available to the Na-bentonite naturally produces a higher green compressive strength.

In all cases of simultaneous use of a clay and a water-dispersed-organic binder, the dry strength properties are always higher than when the clay is used alone and generally lower than when the organic binder

*The proportionality factor w/b can be considered in various ways.

Let w = amount of water added to mixture in per cent.
b = amount of binder "as supplied" by producer, in per cent.
b' = amount of binder less any water content, in per cent.
w' = amount of water in "as supplied" binder, in per cent.

Therefore $b' + w' = b$

or $\dfrac{w}{b' + w'} = w/b$

Another proportionality factor such as $\dfrac{w + w'}{b'}$ can be used.

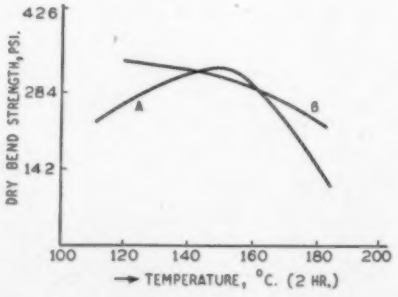


Fig. 14—Relation of the condensation degree of the sulfite lye binder to dry transverse strength.

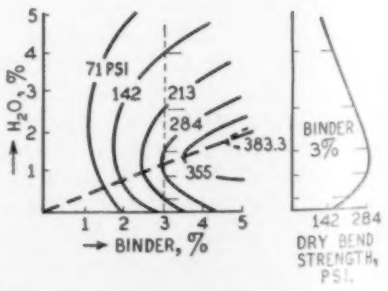


Fig. 15—Dry transverse strength of condensated sulfite lye binder at various percentages of water.

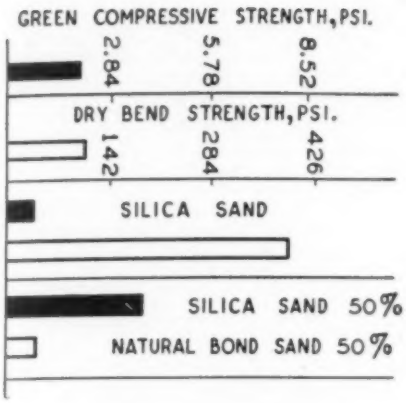
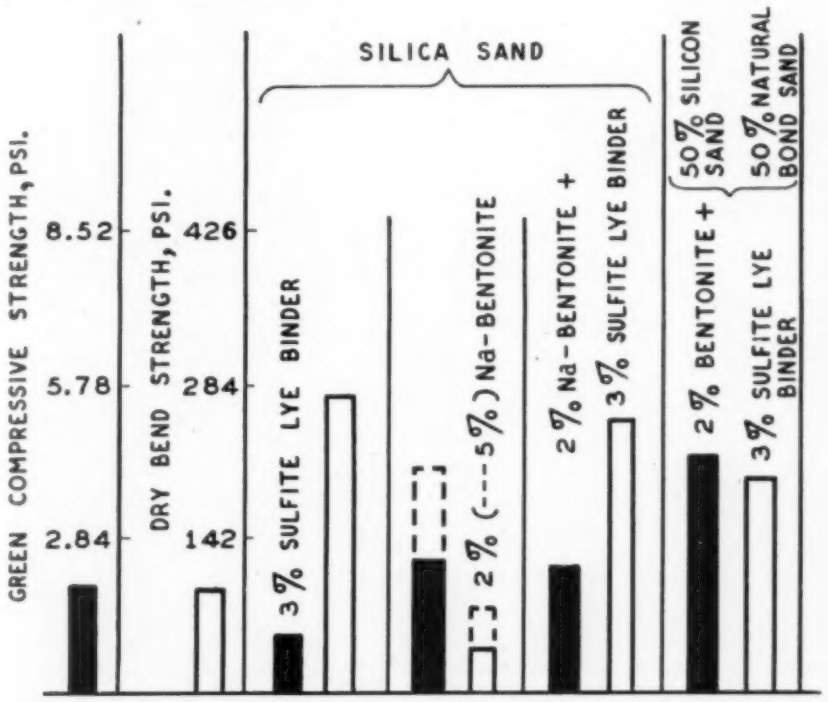


Fig. 16—Relation between the simultaneous addition of an emulsion oil (oil-in-water) and naturally bonded sand to silica sand, with respect to the developed green compressive strength and dry transverse strength.

Fig. 17—Relation between the simultaneous addition of Na-bentonite and a condensated sulfite lye binder to silica sand with respect to the developed green compressive and dry transverse strengths.



is used alone. Water-soluble-organic binders are influenced less by the presence of clays than water-dispersed-organic binders.

Other properties such as reduced drying on the molded surface or change of strength at high temperatures as well as "green" properties are involved in the simultaneous use of inorganic and organic binders.

The rather widespread use of cement-sand mixtures has led to experiments with various combinations which should have offered a mixture of certain green compressive strength needed for the production of cores. For this purpose water-dispersed starch or ethylester of cellulose have been used. They should have influenced the factor w/b of the cement-water system.

At first the strongly swelling substance should deprive the cement of water. During the setting of the cement the organic binder should permit the returning of the water necessary for the hydration of the cement. Na-bentonite, or even the admixture of a naturally bonded sand has been used with some success, but at a reduction in the final strength of the cores.

None of these methods is a final solution and we can but hope that the results of the theoretical study of binders in the indicated directions will bring improvements, either with an organic binder offering at the same time green and dry strengths, low gas content, and suitable strength at higher temperatures, or by a combination of binders influencing one another in a positive sense.

Sand at Higher Temperatures. Up to the present, very little attention has been given in Europe to strength of sand mixtures at elevated temperatures. The study of the influence of elevated temperatures has been limited to observation tests.

This test leads to the knowledge that sands bonded by clays increase in strength properties up to a certain temperature at which no liquid phase occurs. The causes of this phenomenon should be determined with regard to the extent of crystalline transformations, to reactions in the solid state, and to the tighter binding of the exchange cations which on warming lose their exchange ability.

Much attention has been given to the behavior of sand mixtures at temperatures approaching the melting points of iron and steel. Ardenne and Endell[14] especially have studied, under the electron microscope, bonding materials at fusion temperatures and the following conclusions have been made: The fusion temperature of binders which form a continuous peel of sand facilitating cleaning is relatively low. Up to the present the only experiment regarding the viscosity of the bonding material of a naturally bonded sand, from the Rosenthal district, has been conducted in the temperature range 1200°-1450° C (2192°-2642° F).

According to Endell[15] viscosity at elevated temperatures can be calculated from chemical composition. The results for certain available sands are interesting even though not reliable at this time. However, these results do support Endell's conclusion regarding the calculation of viscosity and this field of investigation deserves more attention, for it may lead to improvements in molding sand.

Conclusion. The description of experimental details has been limited as much as possible. The methods used to determine the strength properties do not differ essentially from the Standard A.F.A. procedures[16]. The A.F.A. procedures are by far the most frequently used in Europe even though objections have been raised regarding the method of ramming the test specimens.

It has to be admitted that to some extent these objections are justified, especially by those foundries that desire to study more thoroughly the mechanism of bonding. It is evident that this subject deserves to be ex-

amined and discussed. *Current* foundry practice will not be influenced by any discussions concerning ramming.

Besides his own results and experiences the author quotes those of other authors for the purpose of more completely informing the American foundryman upon the European trends of investigation during the long interval of the war. Unfortunately American publications so important for comparison were inaccessible until recently.

Summary. For the explanation of the bonding mechanism resulting in green compressive and dry transverse strengths of molding sands we have to assume the interaction of certain surface forces never before described in foundry literature. Attention has been given to the viscous properties of binders, with special emphasis upon thixotropy, regarding their influence upon the observed green compressive and dry transverse strengths. The importance of the factor w/b has been shown. Strength at elevated temperatures has been briefly mentioned.

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DISCUSSION

Chairman: H. F. TAYLOR, Massachusetts Institute of Technology, Cambridge, Mass.

Co-Chairman: J. A. RASSENFOSS, American Steel Foundries, Indiana Harbor, Ind.

D. C. WILLIAMS[1]: I understand there is a shortage of cereal binders. I think if you will take advantage of the fundamentals Dr. Jenicek brought out in his paper, you can perhaps use a clay in your core sand mix. However, you are going to run into difficulty. You may have to take into account the shear stresses that Dr. Jenicek mentioned. Bentonite will probably be the best one. You can use it in small quantities, but increase the shear stresses in your mixture. You will have to increase the spread of revolution of your mixer.

He also brought up the point which I think is important in mixing two clays together in the same mixer. Be sure that those two clays, in their fundamental properties, complement one another. Some of you mix fire clay and bentonite in the same mixture. They, however, do not have properties which complement each other, and one subtracts from the other. It will take a lot of study to appreciate this paper.

He also mentioned the fact that the order of mixing the ingredients is important. You can change the properties of any new mixes by the way you add the ingredients to the mixers. If you would understand the fundamentals that Dr. Jenicek has presented you can get a more economic use from binders that you are going to use, and particularly some of the organic ones of which there may be a shortage. You will use less of them and get better properties out of your mixture.

[1]Cornell University, Ithaca, N. Y.

CRITIQUE OF MAURER DIAGRAM FOR CAST IRON*

Marcel Remy
de l'Association Technique de Fonderie de Belgique

TWENTY-TWO YEARS AGO Maurer published his famous diagram in *Kruppsche Monatshefte* July, 1924. Since then, founders have been faced with this diagram (Fig. 1) in every foundry technical publication.

The meaning of the diagram may be briefly recalled: Total carbon percentages are scaled along the ordinate of the diagram, silicon contents along the abscissa; each point of the diagram corresponds to a possible composition of a Fe-C-Si alloy. According to Maurer, the structure of a cast iron would be adequately determined by its carbon and silicon contents, and the various possible structures would be realized by alloys, the composition of which would correspond to certain zones of the diagram, these zones being contained between straight lines converging at a point on the ordinate corresponding to 4.3 per cent total C.

From the start this diagram proved to be at fault and Maurer himself provided a small zone around point 4.3 of total C, which would correspond to what he called "meliertes gusseisen" or mottled cast irons. Then there was the question of cooling rate; as far back as 1927 Maurer published a paper, in collaboration with Holtzhaussen, and from that time it was recognized that the original diagram was only correct for test pieces of 30 mm. whereas, originally, Maurer's pretensions were unlimited. Reading the 1924 paper, one is astounded at the alleged scientific laws propounded by the author.

In his 1927 paper Maurer corrects the mistakes made in his diagram in the zones corresponding to high C content by bringing higher up point *A* to which the straight lines converge.

In a recent paper by Uhlizsch, a new diagram is shown with a different set of lines for each rate of cooling. These new diagrams definitely condemn the original hypotheses of Maurer.

The eminent French scientist, Prof. A. Portevin, commenting on a study of the matter which the writer sent to him, says: "Maurer has made an error of conception and a number of errors of method or of drawing." Portevin refers to experiments made by Mr. Laplanche on cylindrical test bars of 50 mm. diameter cast in green sand. These experiments have enabled a series of diagrams to be drawn, corresponding to various values of a coefficient *K* which characterizes the aptitude of cast iron to graphitization. It should be noted that the curves of these diagrams diverge in the direction of the higher carbon contents and converge in the direction of the higher silicon contents. That is precisely in contradiction to the Maurer diagram in which the straight lines converge toward the higher C and diverge toward the higher Si. No greater opposition could be found between the representation of results.

The reader will also find in *La Fonderie Belge* of the Belgium tech-

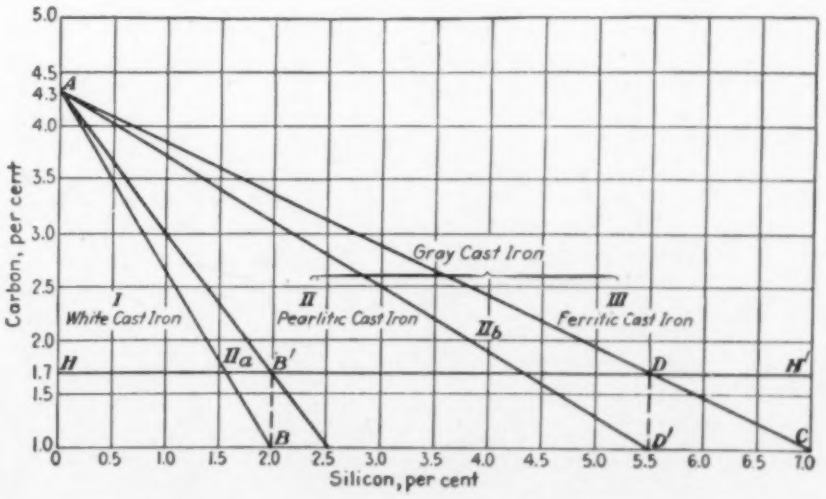


Fig. 1—Maurer Diagram for Cast Iron.

*Presented informally at a Gray Iron Committee Luncheon May 8, 1946, at the Fiftieth Annual Convention of the American Foundrymen's Association in Cleveland, Ohio, by Vincent Delport, European Manager, Penton Publishing Co., for Marcel Remy of de l'Association Technique De Fonderie De Belgique. It is an abstract of three papers on the same subject published in *la Fonderie Belge*. These papers were translated from the original Belgian by the Staff of Battelle Memorial Institute and are herein included in abstract form.

nical association the writer's own opinion as a foundryman: I ask myself how it is possible to clarify all the structures in one diagram having total C as an ordinate, while the phenomena of solidification alone are influenced by total C, and this factor (of total C) does not intervene, or hardly, in the phenomena of transformation in the solid state (formation of pearlite and ferrite).

By what should the Maurer diagram be replaced? In any case, as far as Belgian founders are concerned, the Maurer diagram is definitely condemned to eternal oblivion.

FALLACY OF THE MAURER DIAGRAM*

M. Georges Halbart[1]

FOUNDRYMEN ARE FAMILIAR with the Maurer diagram for cast iron (Fig. 1). The writer shall recall its meaning to you.

If the total carbon contents are represented as ordinates and the silicon contents as abscissae, each point on the diagram corresponds to a possible composition of an Fe-C-Si alloy.

The structure of a cast iron, according to Maurer, would be determined by its contents of C and of Si and the different possible structures would be realized by alloys corresponding to the zones of the diagram delineated by the straight lines converging at a point corresponding to 4.3 per cent carbon, 0 per cent silicon.

Under these conditions, Zone I would be the zone of white cast irons; Zone II that of pearlitic cast irons, and Zone III that of ferritic cast irons; Zones IIa and IIb being intermediate Zones, the first that of mottled cast irons containing both primary cementite and graphite, and the second that of pearlite-ferrite-graphite cast irons.

What is it that impresses us in this diagram? It is first of all its simple form. Theoretical one might say, in order not to say "scientific." There are straight lines which converge at a point, which is known, on the iron-carbon diagram, to be a singular point.

But, from the beginning, the Maurer diagram, in practice, has been at fault. This was not the first time, nor the last, that an experimental study has been vitiated by error. In science, nothing is ever definitive. One approximation replaces another, and therein lies progress. But, besides modification and improvements, there are changes, particularly difficult to accomplish, such as those which represent the abandonment, not of a simple collection of experimental results, but of a law or theory.

There was first the point 4.3 per cent carbon, 0 per cent silicon which, in the diagram in question, was a point. An alloy of this composition would have an indeterminate structure, since this point was common to all zones. Also, Maurer traced, around this point, a boundary delineating a reserved zone: that of "mottled cast irons" (meliertes gusseisen).

Next, and above all, there was the question of rate of cooling. Admitting that the diagram may be valid for "average" pieces, it was certainly not so for pieces of very different thickness. Foundrymen did not rely on Maurer to know that a composition, cast in thin section, would give a white cast iron, but if cast in a heavy section would give a pearlitic or even ferritic cast iron.

In 1927, Maurer published a new report in which he studied the influence of the rate of cooling on the structures of cast irons. In his first study, Maurer had specified that his diagram was only valid for "normal conditions."

In his first study, Maurer found a small piece of cast iron which persisted in not solidifying in an orthodox manner. But it had, citing Maurer, solidified under "unfavorable" conditions.

Maurer published in collaboration with Holtzhaussen the initial diagram for various conditions of cooling. Here it is a question of systematic tests. These tests were made on bars of uniform dimension (Θ=30 mm.); the rate of cooling was modified by preliminary heating of the mold, or by the use of a metallic mold.

From this epoch dates experimentation with 30-mm. test bars, and from this time, it was admitted that the original diagram was valid only—and had been set up only—for this dimension of test bars. Now this is an error. In considering the Maurer diagram, it is well to recall that initially its pretensions were unlimited: there was a single diagram for all cases or, at least, for all "normal" cases.

But, if he deemed it necessary to restrict the field of application of each of the diagrams, each, however, kept the same form as the initial diagram; only the position of the different lines was modified. The authors who, in the second memoir, took the trouble to do serious experimental work, committed the sin of the preconceived idea.

They decided in advance that it would be the form of the diagram in which they would, willy-nilly, enter their results. At the most, they subordinated to these experimental results, the value of the numerical coefficients, that is to say, the position of the lines of the diagram. But, in its essence, the "law" remains the same. The theoretical foundations are not shaken. There is no revolution.

In fact, what then were the theoretical foundation: For our part, we do not know. Borgerhoff had the curiosity to seek them in the sources. Here is what he found: here, in their eloquent reality, are *the scientific bases of the Maurer diagram,* extracted from the memoir of 1924.

"In his diagrams for special steels, Guillet ran the lines together at carbon content of about 1.65 per cent. An analogous point should be determined on the carbon axis of the diagram of cast irons and, accordingly, I have chosen the eutectic point at 4.3 per cent. On the silicon axis, I placed the point "B" at 2 per cent Si because, as a result of my

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*Abstract of translation of original article appearing in *la Fonderie Belge*, p. 7, Special No. (1940-1945).

[1]Ingenieur A. I. Lg. Secretaire General de l'A.T.F.B.

tests on black fracture in steel, I recognized that steels with 2 per cent Si could in all safety be cast up to a content of 1 per cent C, without there being any separation of graphite.

I have taken the point "C" with 7 per cent Si from the Guillet diagram for silicon steel. By joining points "B" and "C" to point "A", there are obtained the three fields sought.

In the following way, there were obtained zones of transition IIa and IIb between the three fields: Point "B" was projected toward the top, on the horizontal which passes through the point 1.7 per cent C, because this C content should be considered as the limit point between steel and cast iron. Inversely, point "D" was projected to D^1 on the X-axis; then "B^1" and "D^1" are joined to point "A" so as to obtain the fields of transition."

In the memoir of 1927, Maurer discreetly goes a step toward rectifying another glaring error. His diagram was especially erroneous in the zone of highly carburized cast irons; thus, if one believes his first diagram, every eutectic cast iron should be ferritic. Maurer corrects the convergent lines, which, at the same time, avoid henceforth (it is very certain) the critical point "A". The diagram is very much changed. Frankly, there remains nothing of the harmonious initial law.

The law of Maurer is useless to the foundryman as a diagram which indicates the analyses to use in each case, and implies a law of geometrically simple form. Moreover, such a law would be useful to the foundryman, if the steps that should be taken to verify it by experiment were not indecently limited.

If Maurer had begun by doing experiments and if he had sincerely reported his results, it would have given foundrymen a serious tool; it would have been proof of scientific honesty—and would have remained unknown. In place of that, he tranquilly designed a fantastic diagram —and has become famous.

In Use 20 Years

For 20 years, this diagram has been reproduced even in university courses, frequently in its initial form and without any indication that its theoretical value is null and its practical value around zero, and perhaps negative because of mishaps which it may cause. It is time that this mistake, as absurd as it is injurious, cease. *We condemn Maurer and his diagram to perpetual oblivion.*

But, in all equity, can we not find in Maurer certain attenuating circumstances? Perhaps. First this: his diagram had qualities from the mnemonic point of view, a feeble compensation, obviously, to the major vice of inexactitude.

But Maurer undoubtedly had a greater merit: In 1924, he had insisted on the fact that, in discussing cast iron, the greatest simplification possible was to consider it as a ternary iron-carbon-silicon alloy, with two independent variables, and, in 1927, he added to these factors, the speed of cooling.

In the writer's opinion likewise, it is fallacious to try, as in the case of steel, to report properties of cast iron merely in terms of carbon content.

Replacement Diagram

We have condemned Maurer. Then the question is: By what will we replace his diagram, because it must be admitted, in all equity, that for 20 years it has been the answer—a bad one, it is true—to pressing questions? It was a bad tool, but a necessary tool.

Certainly, from a practical point of view, there is no difficulty. A two-dimensional, purely experimental diagram, established and used for a fixed rate of cooling will advantageously replace the Maurer diagram. There are such diagrams and perhaps they will be used when Maurer is forgotten.

Likewise, in place of establishing a diagram for each rate of cooling, one can be established for each structure. Such diagrams give the necessary relations between rates of cooling, and C and Si contents. A diagram (which even takes into account P contents) exists for pearlitic cast irons and is known as the Osann diagram. This indicates, for example, at about what silicon content a pearlitic cast iron is obtained when the thickness of the casting, as well as the carbon and phosphorus contents, is known.

Mr. Remy has evolved the basis of a new practical study of the question. He would classify cast irons, and especially cast irons poor in phosphorus, according to their combined carbon content, and not according to more or less uncertain metallographic structures.

Graphite and Carbon Content

The writer is personally of his opinion, insisting, however, on the fact that at least two figures are necessary to characterize approximately a cast iron: it would be necessary, aside from the combined carbon content, to take into account the graphite content. Moreover, it would be indicated to report the combined carbon content in the matrix, as 98 per cent matrix, for example, for a cast iron with 2 per cent graphite.

Under these conditions, a cast iron would be considered constituted of a steel matrix more or less in carbon (combined) and graphite. The two figures: combined carbon of the matrix and graphite content would be the two most logical values to characterize a cast iron from the structural point of view.

Let us recall that in an ordinary steel, a single figure, carbon content, as a first approximation, characterizes the steel. The work to be done would consist of establishing the relations between these two figures naturally characterizing a cast iron as a first approximation, and the independent variables, knowing the thickness of the piece, the total carbon content and silicon content.

Here is work for the future. The object of this work would be to convert the three independent variables (thickness, total carbon and silicon) to two variables only (graphite content and carbon content of the matrix). All the other properties of ordinary cast irons could thus be classified in a double entry table, instead of depending on three parameters.

From the theoretical point of view, now, it is necessary likewise to replace the Maurer diagram which palliates the insufficiency, for the foundryman, of the iron-carbon diagram. It has been proved, in fact, that the iron-carbon diagram cannot, by itself, explain the metallurgical phenomena which are encountered in the production of cast iron and the practitioners who are always remaining on the single variable car-

bon, are singularly destitute.

There exist now the ternary diagrams iron-carbon-silicon and iron-carbon-phosphorus. The use of these two diagrams, even in a very simplified form, permits an understanding of the process of solidification which the use of the three basic binary diagrams cannot explain. Moreover, by a particular chance, not at all evident at the beginning, we can reason soundly on the two ternary diagrams and thus avoid the use of a quaternary iron-carbon-silicon-phosphorus diagram.

In correlation with these diagrams, we can state the two following laws:

1. Crystals of primary austenite, is an ordinary hypoeutectic cast iron, always have a higher silicon content than the mean content of the alloy. They are, on the other hand, poor in phosphorus. Under these conditions, the "eutectic" liquid will be poorer in silicon and richer in phosphorus than the initial alloy. The difference will be greater as the alloy is poorer in carbon.

2. In a ternary alloy, and even more so in a quaternary iron-carbon-silicon-phosphorus alloy, there is no longer a deposit of a binary eutectic at a fixed temperature, but a deposit of "eutectic" crystals in a temperature interval which can attain 210° C.

In other words, graphite or cementite "eutectics" are deposited in this temperature interval. The only true eutectic, from the structural point of view, is the phosphorus eutectic.

Knowledge of the first of these laws can replace advantageously, for all theoretical purposes, the Maurer diagram. As for the second, it permits the explanation of the fact that the structure of normal cast irons is not, in general, that of a eutectic comprising primary crystals.

Indeed, these laws are deduced logically from the ternary diagram. However, since it is prudent to be distrustful, if not of logic, at least of its application elsewhere than in mathematics, we prefer to consider these two laws as hypotheses to be submitted to experimental verification. To our knowledge, they agree well with experiments.

In connection with logical deduction, we will remark that it would be possible likewise, based on the ternary diagram, to "demonstrate the Maurer diagram" because it is always possible to demonstrate a simple thing, starting from sufficiently complex data. It is sufficient to choose the "important" factors and to declare appropriate postulates. This is then only an intellectual game that is played afterward, to "explain" an experiment.

Thus, the wise man allows history to unfold as it pleases, save for, afterward, explaining logically the events that happened, if it amuses him. If it is possible for a man of genius to foresee history, it is not by logic, but by intuition, even though it is only intuition in the choice of the premises to which the logical reasoning will apply.

In the same way, the savant examines nature and explains it afterward, or at the time. If he makes more or less intuitive "logical" conjectures, he confronts them with experiments and and gives a reason if there is disagreement.

To return to the ternary diagrams, they allow an explanation and perhaps a forecast of the structures of cast irons; but their intricacy will require a constant appeal to experiment.

REMARKS ON THE SUBJECT OF THE MAURER DIAGRAM OF CAST IRONS*

A. Portevin

ESTABLISHMENT OF THE MAURER DIAGRAM for cast irons is cited because, it admits of, in its initial form, a double error against which we must be on guard: an error of conception, in confusing a problem of chemical kinetics with a problem of chemical statics, and an error of method, in excessive extrapolation of lines drawn previously.

1. *Error of conception*—The cementite of white cast iron is in a metastable state with relation to graphite, the stable form in gray cast irons, and the passage of one to the other as well as their temporary coexistence requires a consideration of states other than equilibrium. Consequently, the study is a problem of chemical kinetics for isothermal conditions, each state *being a function of time dependent on the temperature.*

One can then, in all strictness, remove the temperature variable only for times sufficiently long (the speed being small at regular temperatures of observation) or the time variable only for temperatures sufficiently high (where the speed or reaction is great) which results, moreover, in equilibrium conditions—a problem of chemical statics.

Among the many variables upon which the problem depends, there should be designated time t and temperature Θ, or at least a function binding these two variables under particular conditions.

In every first approximation, we can limit the chemical factors to two—carbon and silicon—thus removing the other chemical variables S, Mn, Cr, gas, etc., but, even with this simplification, the problem remains with four variables—C, Si, t, Θ—which can be reduced to three if we allow, as we have just said, t and Θ to be linked by a relation such as the law of cooling.

A novel scheme consists in conceding that this law of cooling is simple, and capable of being characterized by a single parameter, such as the mean speed V, which depends in turn on all the numerous factors of casting, notably the size of the piece, with definite conditions for the other factors of casting.

But we cannot go further in this process of simplification and trace a general diagram of state as a function of C and Si without committing a fundamental error of conception. It is necessary to add a third variable, as we have just said, which

*la *Fonderie Belge*, p. 5, no. 1, Jan.-Feb. (1946). Abstract of translation of original article.

may be the size of the piece, and then we ought always to bear in mind the series of simplified hypotheses on which it is based, so as not to diverge in the conclusions that are drawn from the representation by a diagram.

An entirely analogous problem of kinetics is furnished by the representation of the states of steel after cooling, states which result from the passage of the metastable austenitic state to the pearlitic state. It is the subject of the curves called "S curves" of Davenport and Bain, in which the variables, temperature Θ and time t, occur, the latter in logarithmic coordinates. Here then, there are simplifications which have not been remembered or which have escaped the authors: the curves correspond only to a definite time of austenitic decomposition, and, drawing them in a continuous manner, there are joined three branches of curves corresponding to three distinct phenomena or modes of decomposition Ar^{1}, Ar^{11}, Ar^{111}. In reality, the representation should be in three dimensions by "intrinsic surfaces".

If one associates the temperature Θ to time t by a relation representing a simple law of cooling defined by the speed V of cooling, there are obtained "characteristic curves of tempering" representing the various states of steel at cooling as a function of Θ_c the temperature of heating and V the speed of cooling, the general form of which is portrayed in Figure 2; but it is more a representative diagram for instruction than a utilizable idea, currently in practice by reason of simplifications introduced and of the intervention and the superposition of kinetics of cooling to kinetics of decomposition.

These characteristic curves can be generalized and extended to the representation of the state of steel as a function of any two of the variables, determining the final state of the steel and presupposing the equality of all the others. In particular, one might take as variables the content of associated carbon to the content of another element, Ni, Mn, Si, and one would have in principle, separating the various domains, curves of shape analogous to that of the characteristic curves, such as those resulting from the studies of Tutom Kase.

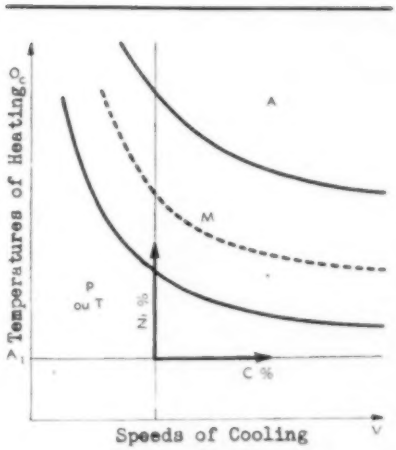


Fig. 2—See Text

2. *Error of method or of drawing*—One can, with a view to simplification, replace a curve by its cord in a not too extended interval; but it would obviously be a profound and inadmissible error to extrapolate this straight line by prolonging it outside the limits which have been adopted.

Even if a curve, determined by experimental methods, appears rectilinear, it is necessary to be careful in this extrapolation. Examples are found in which certain phenomena whose speed decreases with time, such as reactivity, relaxation, contraction of tempered alloys are represented. The experimental results appear, in fact, to follow a straight line, as a logarithmic function of time, but extrapolation would lend to the conclusion that the speed becomes null for sufficiently long times, which is inadmissible and proves the inexactitude of the linear law (Chevenard).

In the case of the Maurer diagram of cast irons, there could have been adopted, as a primary hypothesis, for lines separating the various domains, curves of a form analogous to the characteristic curves of steels cited above, because it should have been understood that the rectilinear line was not admissible since its extrapolation to the carbon axis led to a content of 4.3 per cent, above which cast irons would be invariably ferritic, and below which invariably white, with indetermination at the eutectic content.

The explanation, rather the excuse, invoked by Maurer in his first work of the diagrams of special steels by Guillet, is not an explanation, because these diagrams call for the same observations. The content of 1.65 per cent C, to which converge the straight lines of these latter diagrams, does not have any more significance with respect to the appearance of austenite in special annealed steels, than the eutectic content 4.3 per cent C of the Maurer diagrams has with regard to the appearance of graphite in cast irons.

Norbury, and then Ulitzsch and Weichelt, have drawn diagrams for

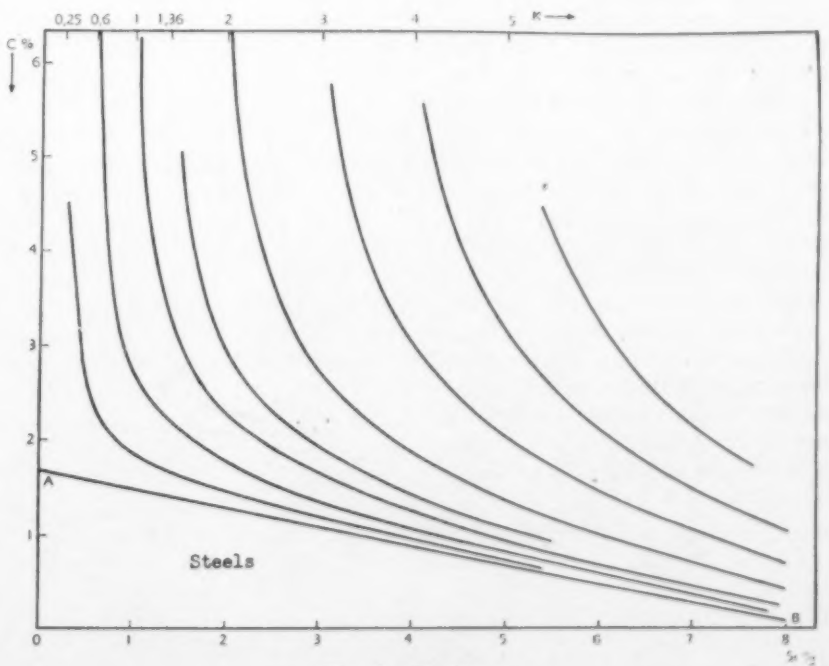


Fig. 3—See Text

cast irons with curves of a form analogous to the characteristic curves of steels to which we have just referred.

These diagrams are in reality only sections of a diagram with the three dimensions C, Si, and thickness of pieces—and each diagram with two dimensions corresponds, consequently, to a determined thickness for a piece cast under specified conditions.

Experiments of Mr. Laplanche permitted him to establish, for *cylindrical bars of 50-mm. diameter, cast in green sand,* families of curves corresponding to various values of a coefficient.

$$K = 4.3\ Si\left\{1 - \frac{25}{15\,C + 3\,Si}\right\}$$

which characterized the speed of graphitization.

Figure 3 gives these curves of equal graphitization, or of equal graphitization, or of equal proportion of graphite after cooling (which we propose to call "isographitization") for values

$K = 0.25\text{-}0.6\text{-}1\text{-}1.36\text{-}2\text{-}3\text{-}4\text{-}5.$

These curves have a remarkable similarity of form to those of Ulitzsch.

As a first approximation, the various categories of industrial cast irons can be classified on these curves. For example, white cast irons for malleabilization by graphitization in annealed nodules are situated on curves $K \leqslant 0.6$; pearlitic cast irons lie between $K = 1$ and $K = 2$; and completely ferritic cast irons to the right of $K = 3$.

In this diagram, the influence of the elements S, Mn, Cr, and Mo which, in their usual contents, displace the curves toward the right just as the increase of the speed *V* of cooling (by diminution of thickness or by chill casting), or even the influence of hydrogen are not taken into account.

The group of these latter curves diverges toward the high carbon contents and converges toward the high silicon contents, that is to say, precisely the inverse of Maurer's group of straight lines, which converge toward the high carbon contents, and diverge toward the high silicon contents. One could not find more complete opposition in the representation of results.

The Foundryman Corner*

WITH REGARD to the Maurer diagram, foundrymen are well off because they never favored this "fantasy."

When it is indicated that cast iron is white or mottled, such classification is made on the basis of the solidification phenomenon disregarding the transformation in the solid state. On the other hand, when cast iron is considered from the point of view of its crystal structure, the classification is made on the basis of the transformation phenomena in the solid state. It is well known that neither pearlite nor ferrite is present in cast iron at a temperature of only 250° C. below its solidification point.

The attempt to classify the cast iron structure, induced by the phenomena of solidification and transformation by a single diagram having as ordinate the carbon content (which does not, or only slightly influences the transformation), has always seemed to us a scientific insanity.

During the famous investigation of the Maurer diagram, the investigating commission appended to its file data obtained by Thyssen-Dawans in an investigation of cast-iron structure. Here, these data are plotted on the Maurer diagram. It may be observed clearly that these curves are in complete disagreement with the straight lines of Maurer.

There is a basic reason why the time-temperature relation influences the cast-iron structure during solidification and transformation. But when this relation is considered as a function of the casting thickness, such an assumption should not be generalized.

Even an apprentice molder, who has made only a few castings, knows that castings of the same thickness do not necessarily have the same structure, and that the phenomena of solidification are influenced by several different factors. Often the parts of the casting farthest removed from the feeding head are white, while the remainder of the casting, although of the same thickness, is gray. All of us remember quite well the experiment with a small cast iron bar of regular thickness, about 100 mm. in length, which stubbornly solidifies into white cast iron at its ends. Two solutions have been proposed to eliminate this behavior: the first is to lengthen the bar by two false ends (thus removing the defect outside the casting); the second is to increase the thickness of the bar by 60 per cent at the ends (assuring a sufficient coefficient of safety).

Conclusion

The present classification of cast iron, on the basis of its structure, we consider to be a purely commercial classification.

The data of experimental investigations will induce two separate classifications based respectively on data resulting from solidification analyses and on data resulting from the investigation of solid transformations.

In applying research data to commercial castings, we should not forget that the production of commercial castings will never take place under conditions of laboratory experiments, and a series of other factors will interfere. Such factors always must be considered.

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